THE JOURNAL OF THE INSTITUTE OF Metals

WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents:

PRICE: TEN SHILLINGS AND SIXPENCE

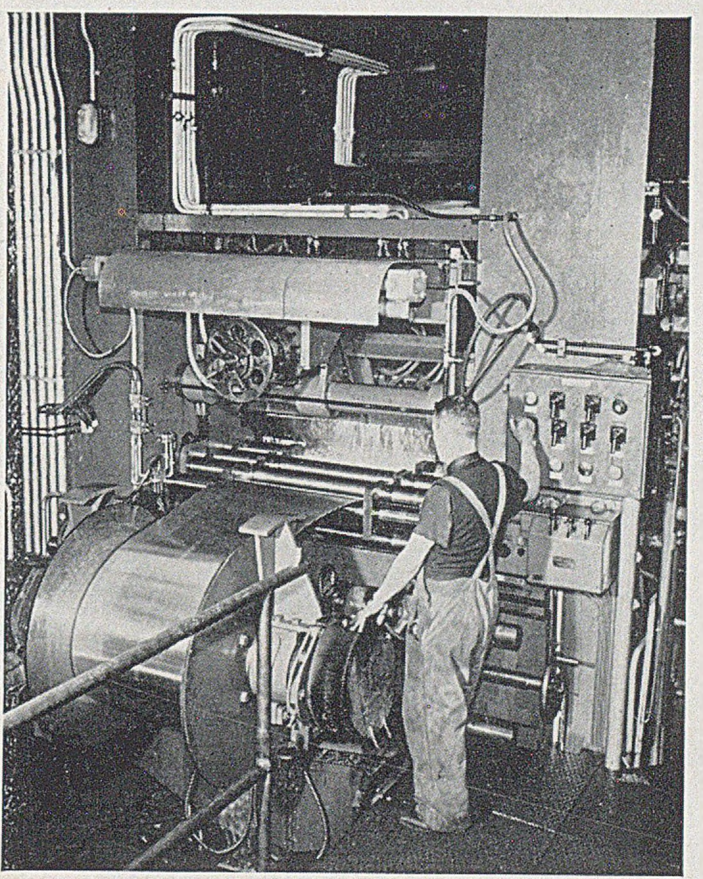
FEBRUARY 1953

DAVY-UNITED — DEVELOPMENTS IN STRIP HANDLING EQUIPMENT

The combined idle cone uncoiler and sticker roller bridle fitted to the entry side of the 66 in. wide Davy-United 3-Stand Tandem Cold Strip Mill at the Rogerstone Works of Northern Aluminium Company Ltd., represents a decided improvement in non-ferrous strip feeding equipment, particularly with the heavier coils which are now coming into greater use in the aluminium industry.

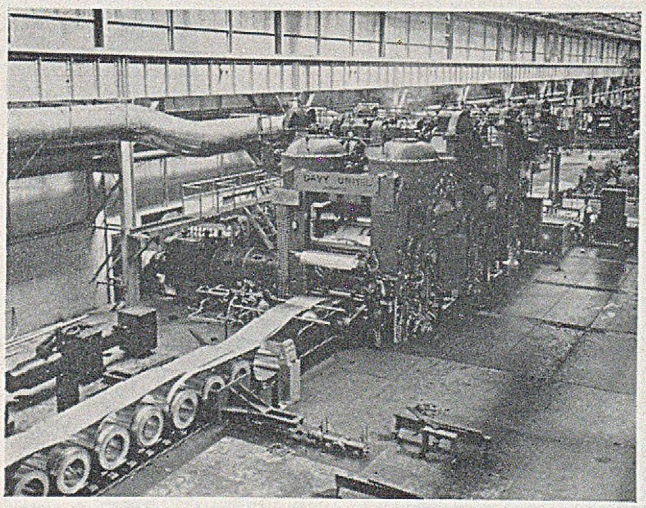
Combined Uncoiler and Sticker Roller Bridle

for aluminium strip



Close-up on the entry side of the tandem mill; the combined uncoiler and sticker roller bridle in the rolling position.

Coils queue up for cold rolling on the 66-in. wide tandem mill, a production-time-saving advantage made possible by the smooth and rapid operation of the combined uncoiler and roller bridle unit.



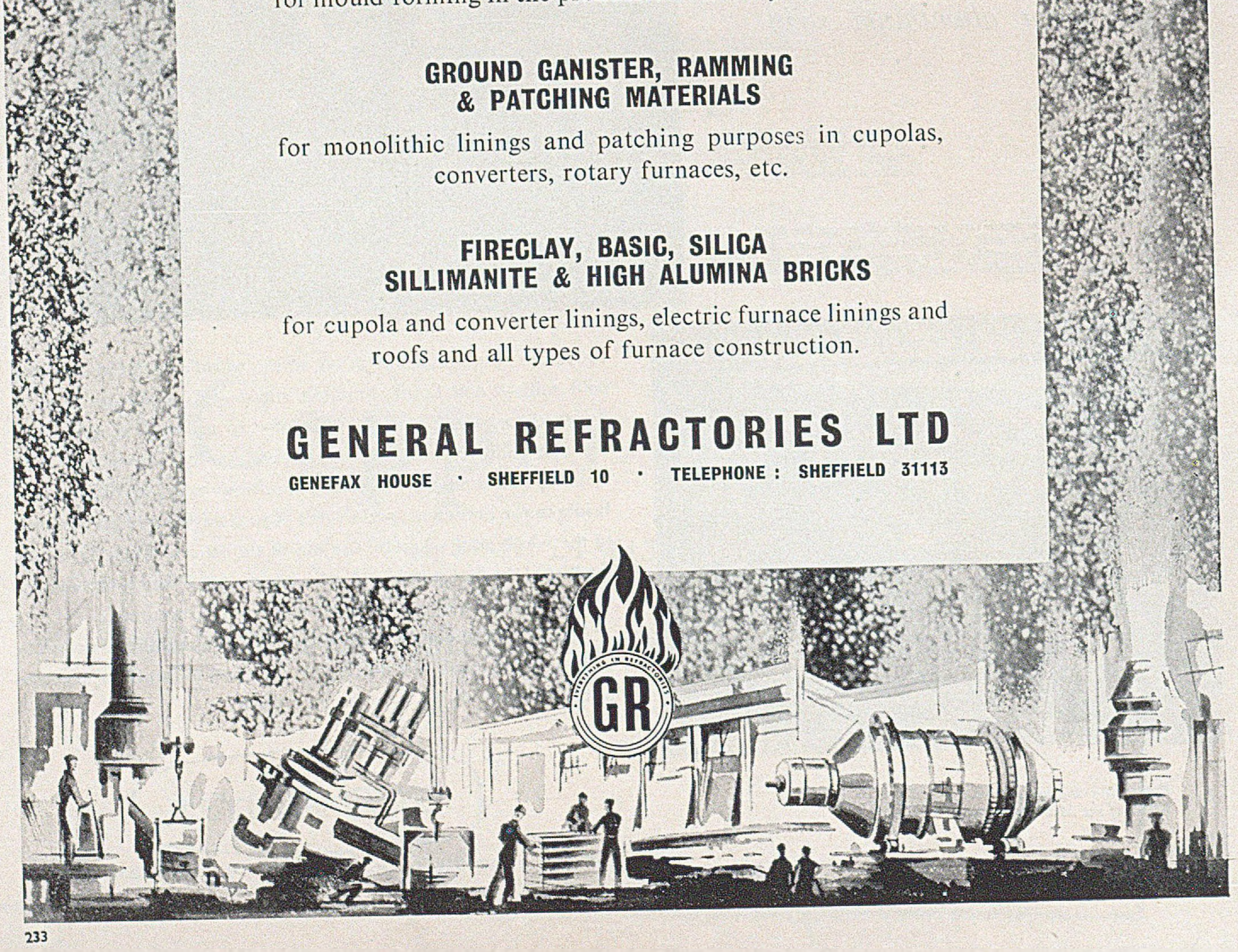
Proved in continuous operation on the world's fastest aluminium cold mill—2,000 f.p.m. finishing speed—this combined unit is designed to handle 5,000 lb. coils of strip up to 56 in. wide, the heaviest coils yet envisaged in British non-ferrous rolling. The opening and closing of the cones, the clamping action of the 7-roll bridle to the predetermined degree of penetration and the traverse of the whole unit whereby the tail of the strip is stuck right into the bite of the first stand work rolls, are all performed hydraulically under interlock control. One operator only is required.

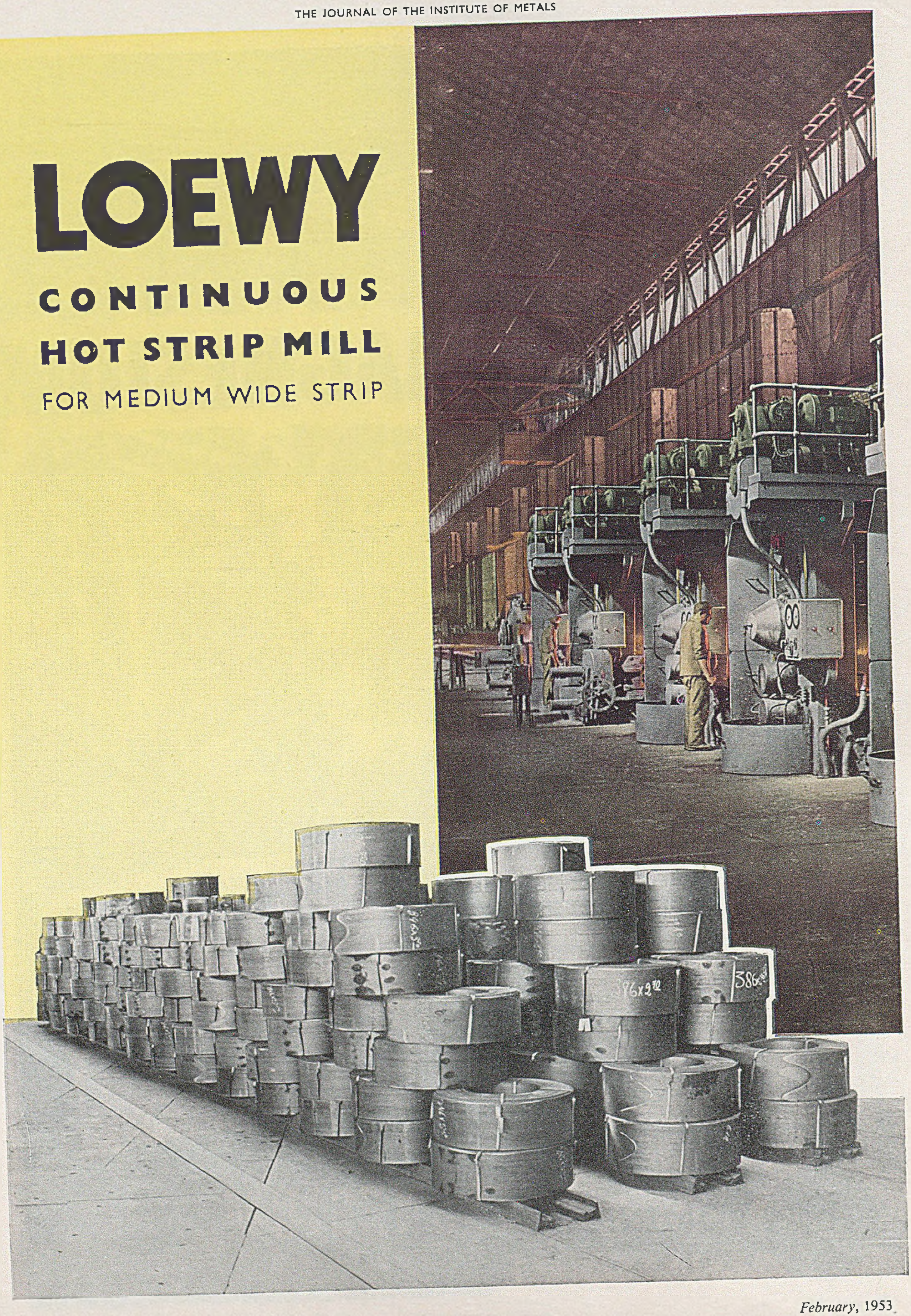
Apart from quickening and simplifying the motions of accurate coil entry the combined unit is also designed to impart a measure of back tension to the strip during rolling.

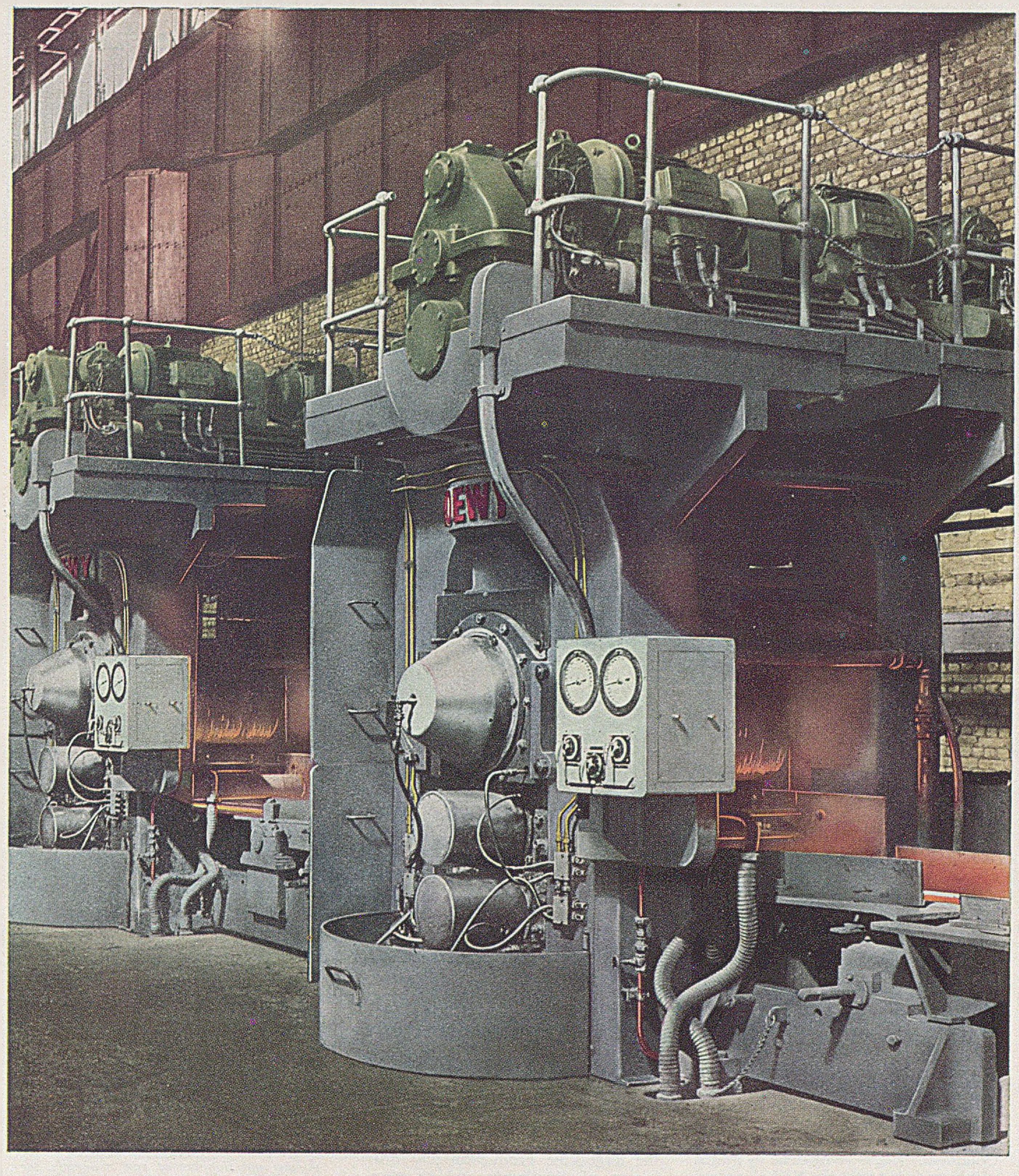
Progressive engineering at the service of the Metals Industries.

DAVY-UNITED

SHEFFIELD · MIDDLESBROUGH · GLASGOW


GR
EVERYTHING IN REFRACTORIES





aluminium adventure

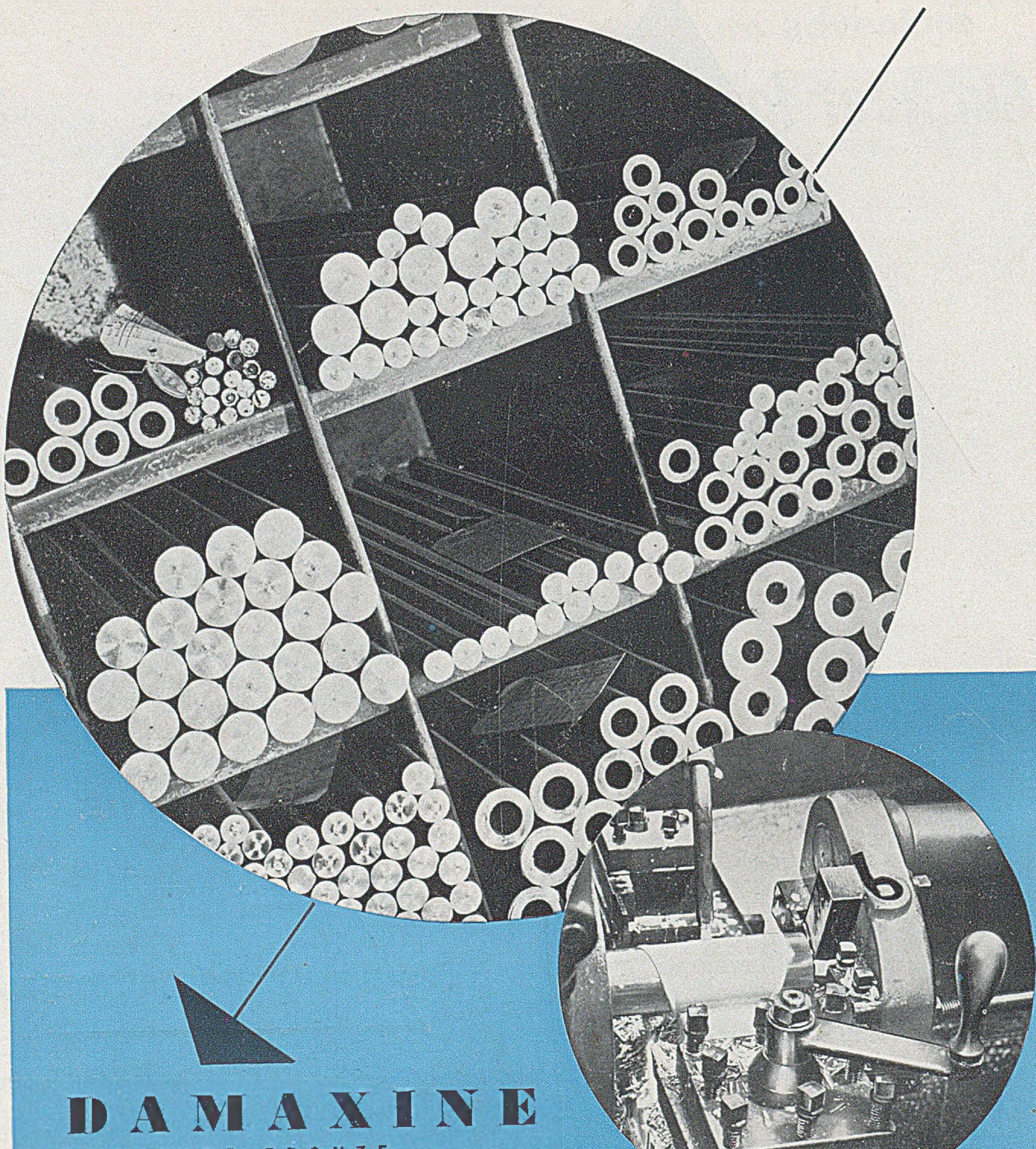
reversing a RIVER SYSTEM

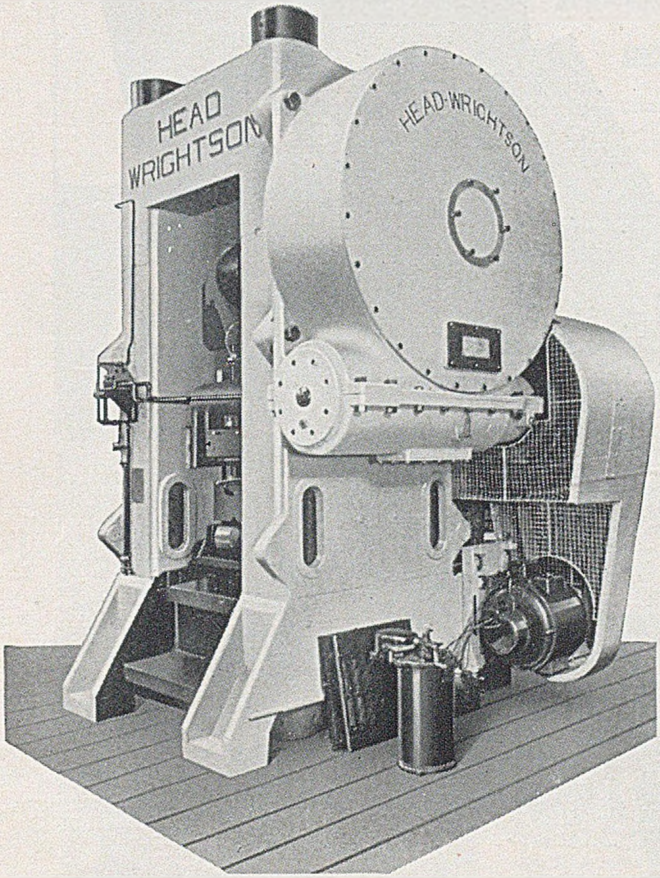
Reversing the course of a river system was one of the first problems to be overcome by the Aluminum Company of Canada Ltd., in their new hydro-electric project now taking shape in British Columbia. This is being done by means of a huge dam across the Nechako river, which, by the skill and ingenuity of modern engineering, will convert the river system into a huge reservoir with a surface area of 350 square miles. The water in this reservoir has only one escape route — through a ten-mile tunnel drilled through the coastal mountain range. At the farther end of this tunnel, half-a-mile lower than its intake, the water will become power. Here, in a cavern carved out of the mountain, will be the largest underground power station in the world. When fully developed, it will generate sufficient power for the smelting of nearly half a million tons of aluminium a year at the great new smelting plant of Kitimat. All this to attain one end — the large-scale production and distribution of aluminium and its alloys, from mine to market, by members of the Aluminium Limited Group of Companies.

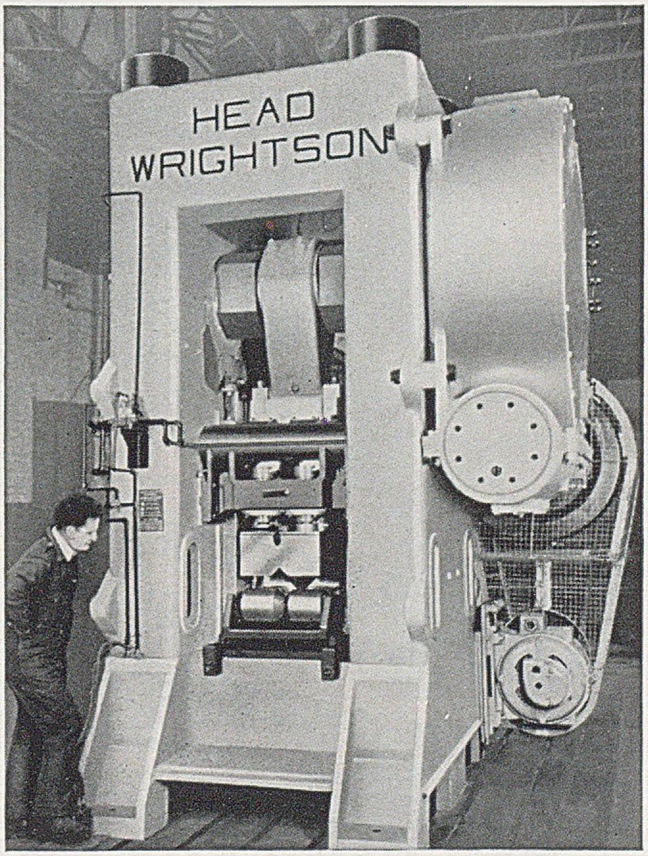
As demand for aluminium increases, and its usefulness as a major raw material becomes more widely recognised, so must production be expanded. One of the leading organisations engaged in this task is the Aluminium Limited Group of Companies whose resources encompass many widespread activities. These cover every aspect of the Industry — the mining and shipping of raw materials, the generating of hydro-electric power and the ultimate extraction and fabrication of the metal. To these must be added world-wide selling services and a programme of continuous research designed to improve production methods and to find new alloys.

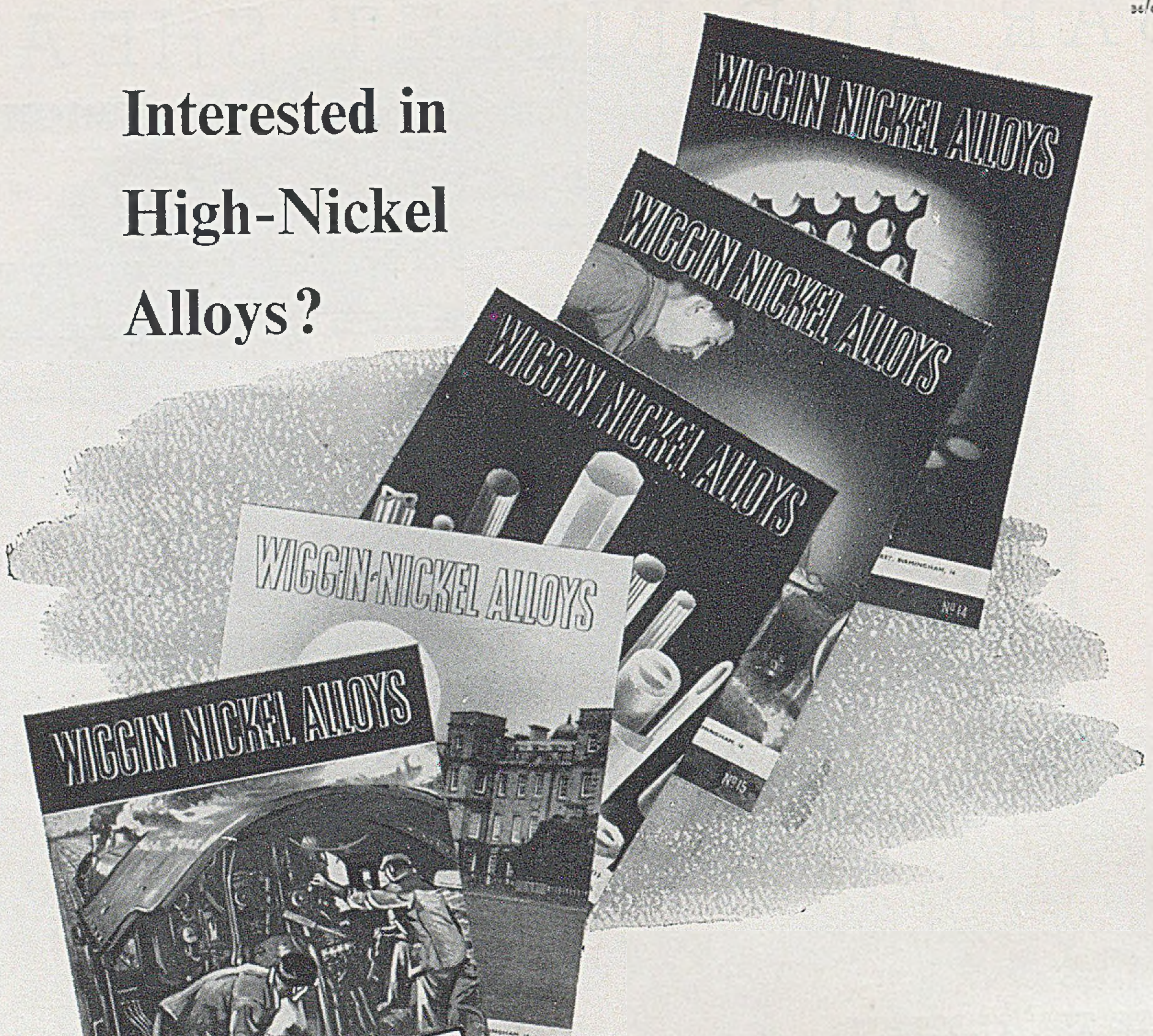
Aluminium Union Limited

THE ADELPHI, STRAND, LONDON, W.C.2. AN ALUMINIUM LIMITED COMPANY
PRINCIPAL BRITISH COMMONWEALTH DISTRIBUTOR OF ALUMINIUM




HEAD WRIGHTSON
HEAD-WRIGHTSON


HEAD WRIGHTSON

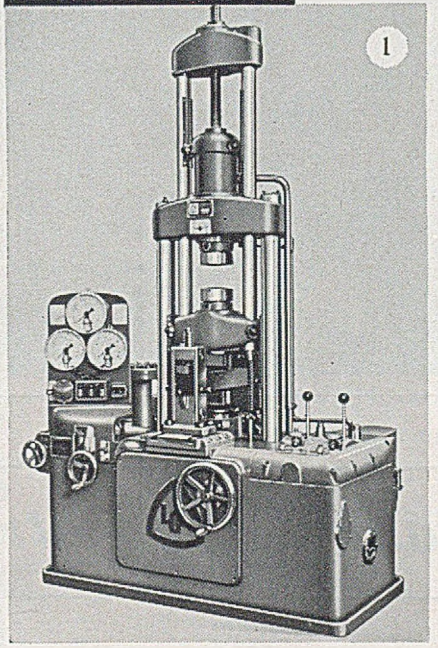


Is your TESTING EQUIPMENT keeping pace with material development?

The strength of materials under normal temperatures has to-day already been examined to a far reaching extent, but the behaviour of stressed materials at high and maximum temperatures is rapidly becoming the subject for a wider field of investigation.

LOS

LOSENHAUSENWERK of Dusseldorf have a long established reputation for the development of material testing equipment, and the machines shown here have been designed to meet many of the requirements of research workers and development engineers concerned with the manufacture and application of materials including the new heat-resisting alloys.



UNIVERSAL MATERIAL TESTING (1)

An outstanding feature of Losenhausenwerk machines for static and dynamic testing of materials is that actual components and complete assemblies can be tested under conditions as near as possible to those to which the parts are likely to be subjected while in use.

FOR STATIC TESTS

Tensile, compressive, collapsing, bending, folding, shearing and Brinell ball pressure hardness tests can be carried out on this equipment. There are machines for loads from 2,000 to 200,000 kg. The larger types can be fitted with a Pulsator for dynamic tests with fluctuating loads in tension *or* compression.

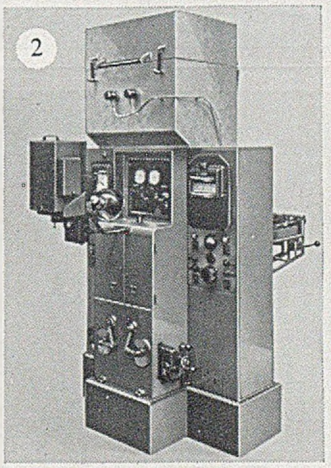
FOR STATIC AND DYNAMIC TESTS

The fatigue life of a component as designed can be determined on these machines in addition to standard fatigue tests on the material.

Three types of fatigue tests can be carried out :—

- (*a*) With fluctuating loads in tension.
- (*b*) With fluctuating loads in compression.
- (*c*) With alternating loads from tension to compression.

Available with dynamic capacities from 1,000 to 10,000 kg. The smaller machines may be fitted with furnaces for fatigue tests at *high temperatures.*



CREEP OR FATIGUE TESTING (2)

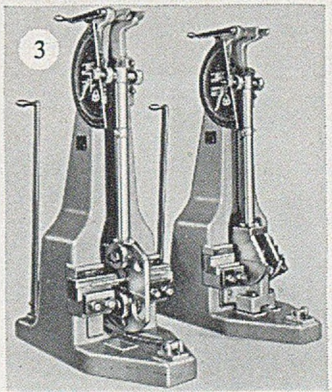
In addition to fatigue testing at high temperatures there is a range of Losenhausenwerk machines for examining the creep of various materials and alloys at *high temperatures* under constant load, constant elongation and variable elongation.

This plant is made in types for single piece tests or for testing up to 48 specimens simultaneously. Experiments can be carried out for the normal period of life characteristic of the part to be manufactured from the material under test. (For instance 10,000 hours or longer.) The attention given to this equipment need not exceed that of occasional supervision.

PENDULUM IMPACT TESTING (3)

Allied with the fatigue testing of material at low and high temperatures is the increasing demand for Impact tests following the Charpy method as distinct from, and in addition to, the Izod method which has long been standard in this country. The advantage of the Charpy method is the speed at which tests can be made so that temperature losses are reduced to a minimum.

Losenhausenwerk supply a range of Charpy type Impact Testers with striking energy 30, 75 or 250 mkg. The normal capacity 30 mkg. machine being capable, if required, of testing by both the Charpy and Izod methods.



GENERAL MATERIAL TESTING

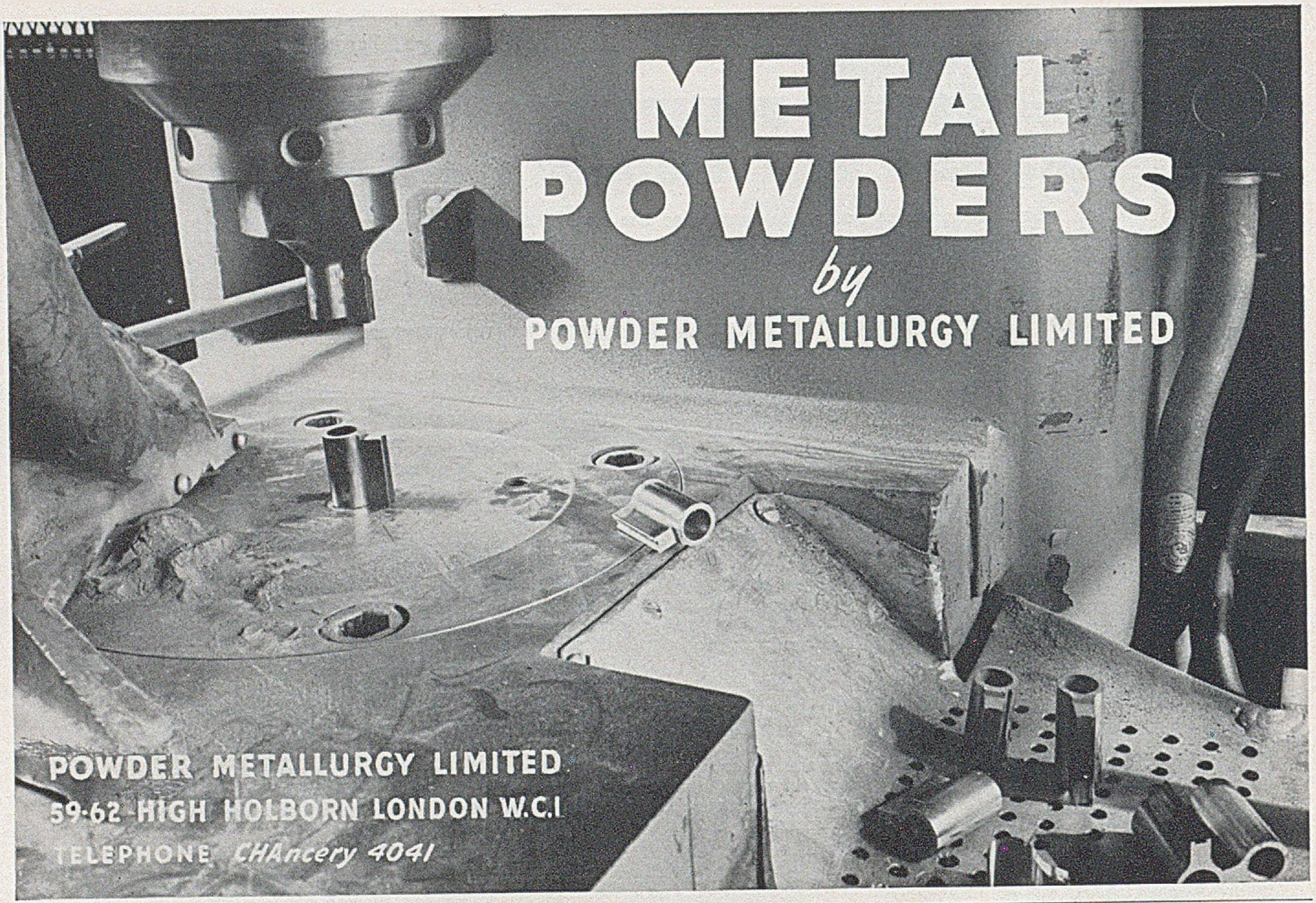
Equipment is available for Brinell Hardness Tests ; Spring Testing, Vibration Testing, Chain and Cable Testing, Drop Testing, etc. Dynamic Balancing Machines, Vibration Tables and Measuring Boxes can also be supplied.

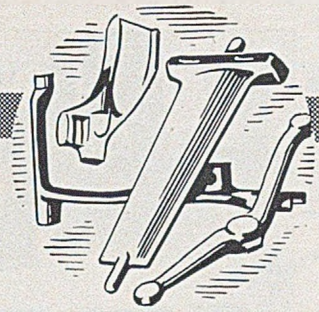
WICKMAN *of* COVENTRY

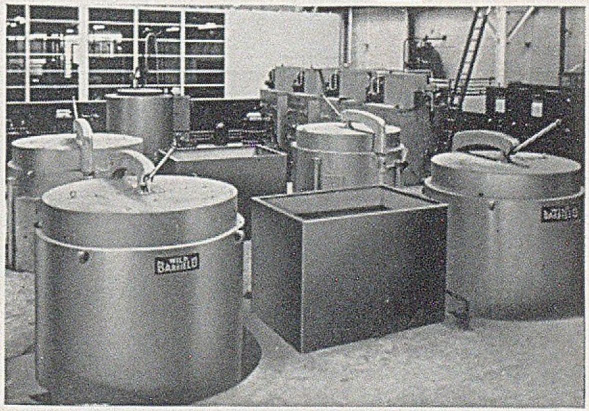
LONDON · BRISTOL · BIRMINGHAM · MANCHESTER
LEEDS · GLASGOW · NEWCASTLE · BELFAST

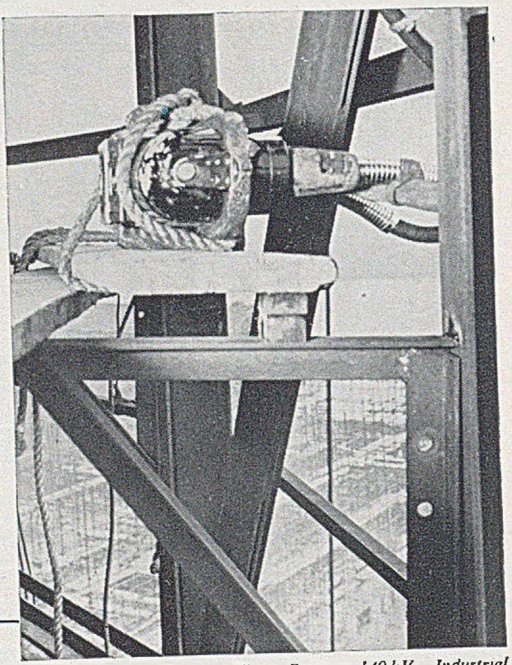
Wickman

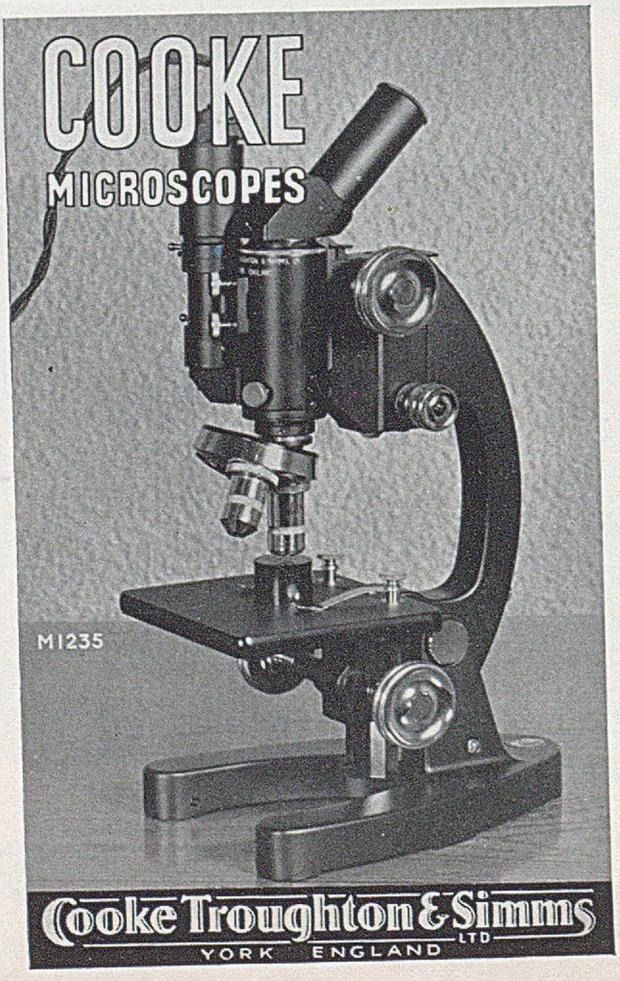
228F85



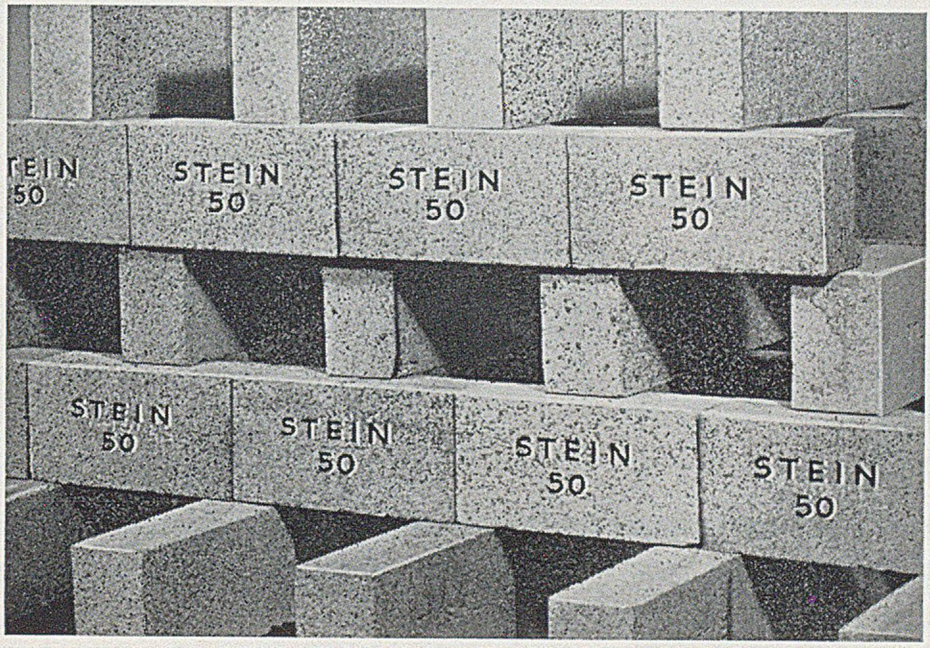







COOKE
MICROSCOPES
M1235
Cooke Troughton & Simms LTD
YORK ENGLAND

STEIN *Refractories*



STEIN 50

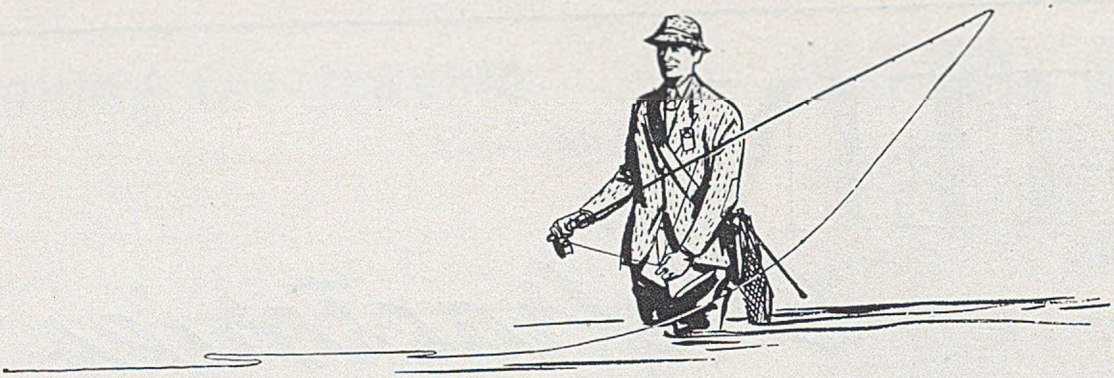
APPROXIMATE TECHNICAL DATA

Alumina	51-52%
Refractoriness	Cone 35/36 1780°C
Refractoriness-under-load 28 lbs./in^2	5% Deformation at 1595°C
After-Contraction 2 hrs. at 1500°C	Nil

For the top courses of Open Hearth furnace checkers where strong resistance to corrosion at high temperatures is vitally necessary we recommend Stein 50. This brand is of high refractoriness and is very constant in volume even after long periods of high operating temperatures. Stein 50 is specially recommended for furnaces burning liquid fuel.

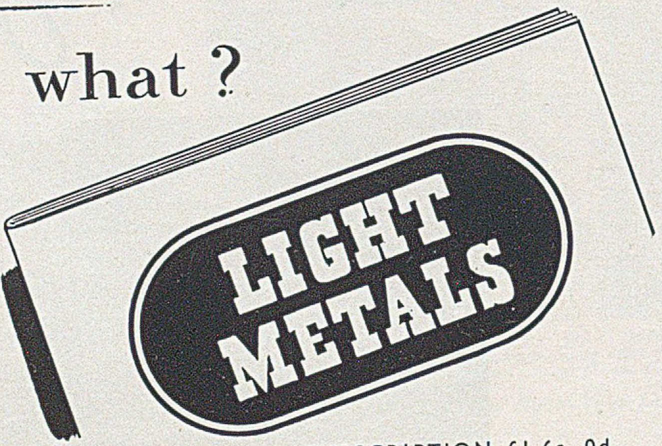
JOHN G. STEIN & C^{o} L^{TD} Bonnybridge, Scotland

TEL: BANKNOCK 255 (3 LINES)



Casting about for what ?

What are you angling for ? Ideas ? Facts and figures ? Current trends ? Or just snags ? Whatever you're after in the realm of light metals and their alloys, there's bound to be something useful on the subject in—



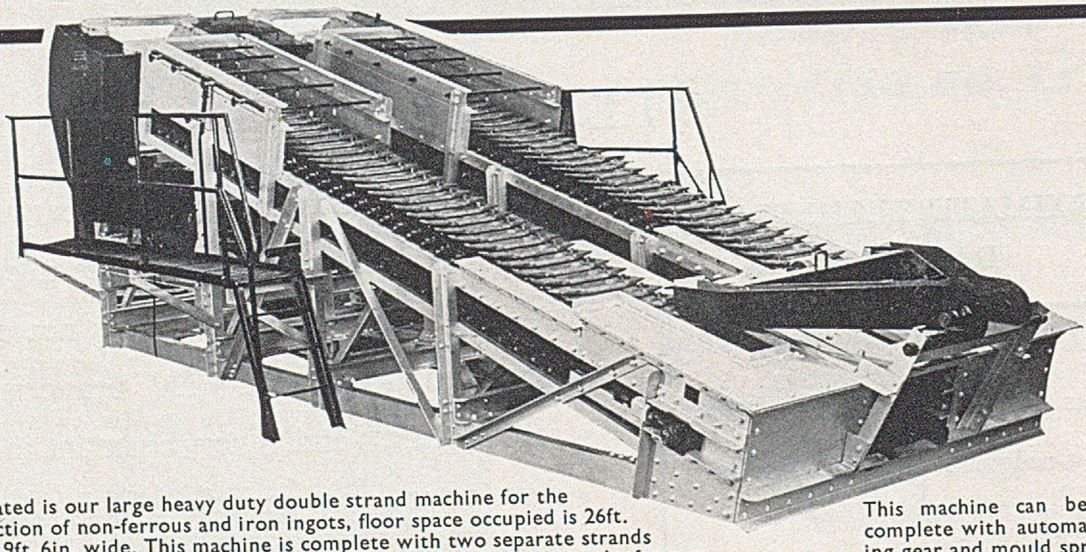
MONTHLY 2s. 0d. ANNUAL SUBSCRIPTION £1 6s. 0d.

TEMPLE
PRESS

TEMPLE PRESS LIMITED

BOWLING GREEN LANE, LONDON, E.C.1. TERMINUS 3636

DOUBLE STRAND HEAVY DUTY INGOT CASTING MACHINE



Illustrated is our large heavy duty double strand machine for the production of non-ferrous and iron ingots, floor space occupied is 26ft. long x 9ft. 6in. wide. This machine is complete with two separate strands each with 100 moulds and receive the molten metal direct from the furnace with integral dual channel launder. The machine is of mobile or static type and each strand is independently driven. It is fitted with water cooling Ingot Sprays and return water drip trays.

This machine can be supplied complete with automatic stamping gear and mould spray coating device for the return empty moulds.

Write for our Ingot Casting Literature

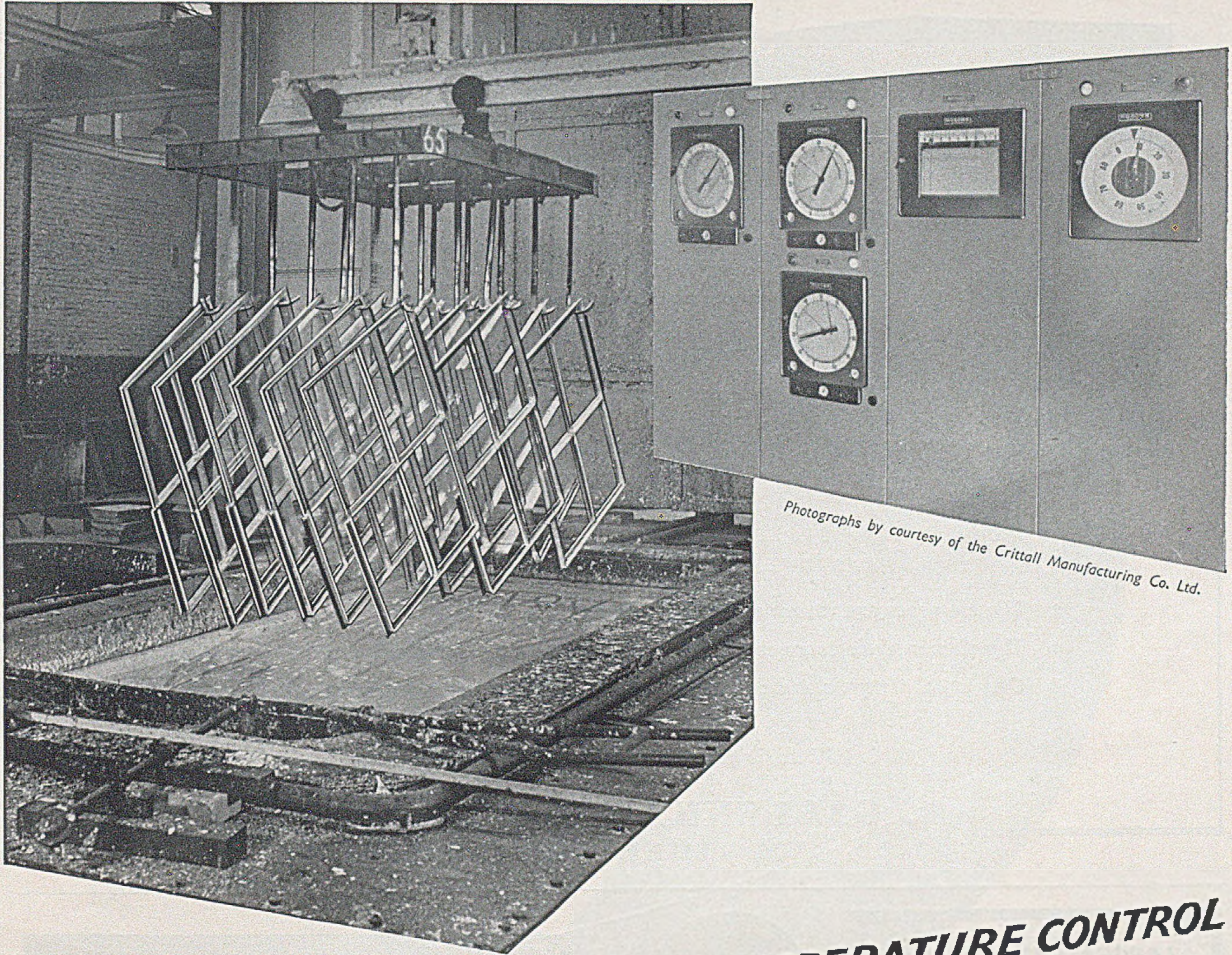
SHEPPARD & SONS LTD.
BRIDGEND
GLAMORGAN SOUTH WALES

Telephone:
BRIDGEND 567 (5 lines)

ESTABLISHED 1862

Telegrams:
"SHEPPARD, BRIDGEND"

Ironfounders, Heavy General, Structural & Mechanical Engineers



Photographs by courtesy of the Crittall Manufacturing Co. Ltd.

BROWN *ElectroniK* TEMPERATURE CONTROL FOR AUTOMATIC METAL PROCESSING

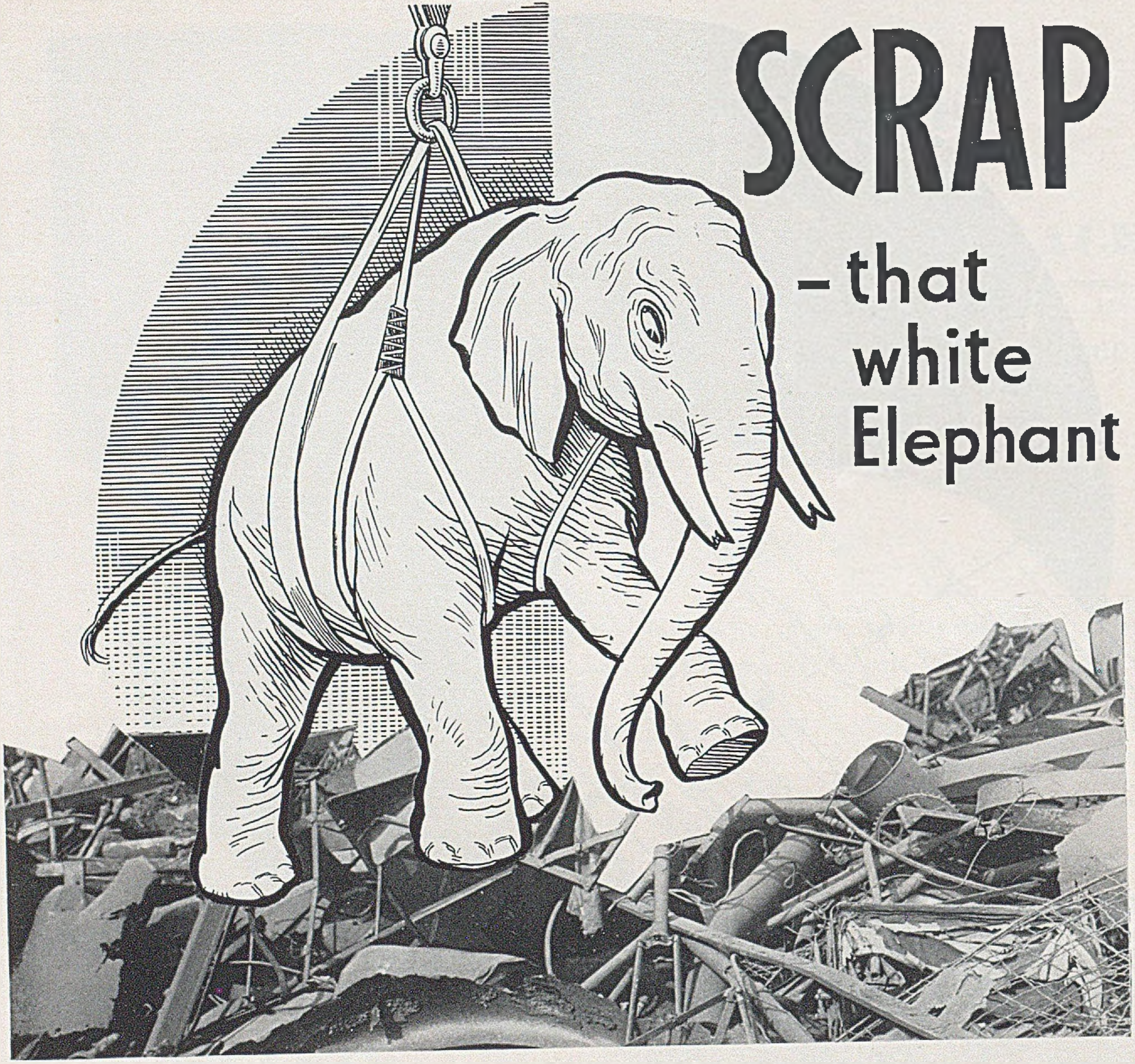
THE AUTOMATIC production lines at the Crittall factory at Braintree, Essex, pass window frames through the various stages of the galvanising plant every three minutes. Vital to the quality of the finished product during this process, is accurate temperature control of the pickling, fluxing, galvanising and etching baths. This control is maintained on each line by five Brown ElectroniK Potentiometers, for indicating, recording and controlling. Three of the instruments are fitted with Air-O-Line pneumatic control. The 'continuous-balance' principle of these instruments means that they balance *directly* from input changes and therefore have greater sensitivity, speed of response and freedom from inaccuracies caused by vibration. Available with either electric or pneumatic control forms, these robust instruments need the minimum of maintenance. Please write for Application and Specification Bulletins.

THE AIR-O-LINE control unit incorporating adjustable Proportional Band, Integral and/or Derivative Action, recognises, analyses and corrects for load changes without cycling, drifting or permanent departure of the pen from the control point.

HONEYWELL-BROWN

ADVANCED INSTRUMENTATION

HONEYWELL-BROWN LIMITED: 1 Wadsworth Road · Perivale · Greenford · Middlesex · Works and Scottish Sales Office: Blantyre Lanark · Scotland AFFILIATED COMPANIES: Amsterdam · Brussels · Paris · Stockholm · Zurich British Made: Patent Specification 436,098


SCRAP
– that
white
Elephant

THE INSTITUTE OF METALS

4 Grosvenor Gardens, London, S.W.1

COUNCIL AND OFFICERS

PRESIDENT

C. J. SMITHELLS, M.C., D.Sc.

PAST-PRESIDENTS

Professor A. J. MURPHY, M.Sc.
Sir Arthur SMOUT, J.P.
H. S. TASKER, B.A.

VICE-PRESIDENTS

G. L. BAILEY, C.B.E., M.Sc.
S. F. DOREY, C.B.E., D.Sc., F.R.S.
A. B. GRAHAM
P. V. HUNTER, C.B.E.
Professor H. O'NEILL, D.Sc., M.Met.
Professor F. C. THOMPSON, D.Met., M.Sc.

HON. TREASURER

E. H. JONES

ORDINARY MEMBERS OF COUNCIL

Alfred BAER, B.A.
E. A. BOLTON, M.Sc.
N. I. BOND-WILLIAMS, B.Sc.
K. W. CLARKE
C. H. DAVY
N. P. INGLIS, Ph.D., M.Eng.
Ivor JENKINS, D.Sc.
L. B. PFEIL, O.B.E., D.Sc., A.R.S.M., F.R.S.
Professor A. G. QUARRELL, D.Sc., Ph.D., A.R.C.S.
A. G. RAMSAY, Ph.D., B.Sc.
Professor G. V. RAYNOR, M.A., D.Phil., D.Sc.
Christopher SMITH
H. SUTTON, C.B.E., D.Sc.
Major P. L. TEED, A.R.S.M.
W. J. THOMAS

EX-OFFICIO MEMBERS OF COUNCIL

E. A. FOWLER, B.Sc., A.R.T.C.
(Scottish Local Section)
C. E. RANSLEY, Ph.D., M.Sc.
(London Local Section)
H. M. FINNISTON, B.Sc., Ph.D., A.R.T.C.
(Oxford Local Section)
K. M. SPRING
(South Wales Local Section)
M. M. HALLETT, M.Sc.
(Sheffield Local Section)
H. H. SYMONDS
(Birmingham Local Section)

REPRESENTATIVES OF OTHER BODIES

The following, in accordance with Article 32, represent Government Departments and allied societies at Council meetings, for purposes of liaison

Admiralty	Capt. (E.) H. J. B. GRYLLS, R.N.
War Office	Major-General S. W. JOSLIN, C.B.E., M.A.
Iron and Steel Institute	Capt. H. LEIGHTON DAVIES, C.B.E., J.P.
Institution of Metallurgists	MAURICE COOK, D.Sc., Ph.D.; E. G. WEST, B.Sc., Ph.D.

SECRETARY

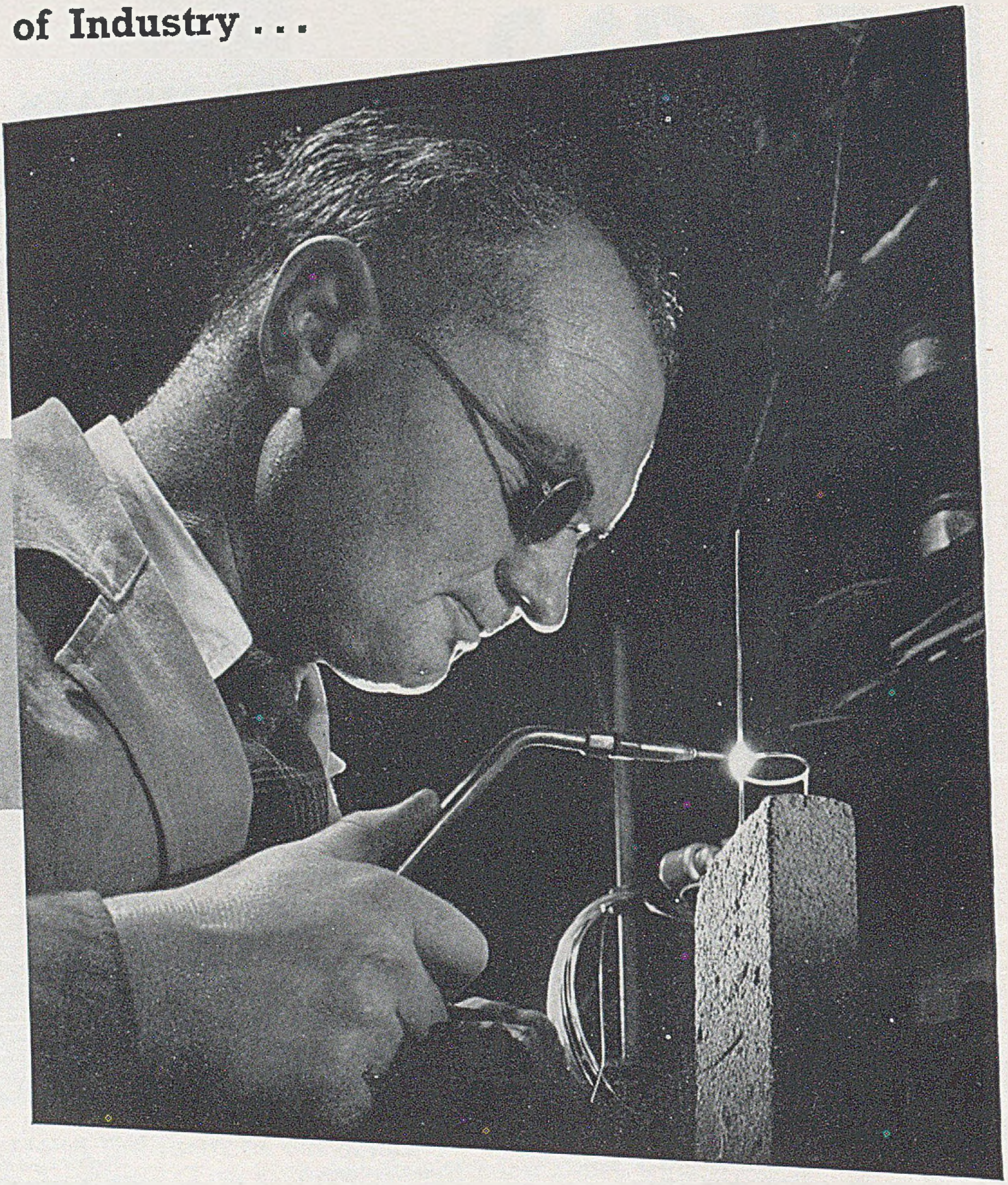
Lieut.-Colonel S. C. GUILLAN, T.D.

ASSISTANT SECRETARY

Major R. E. MOORE

EDITOR OF PUBLICATIONS

N. B. VAUGHAN, M.Sc.



JOURNAL OF THE INSTITUTE OF
METALS
WITH THE BULLETIN AND METALLURGICAL ABSTRACTS

SUBSCRIPTIONS
Members: Free

Non-Members: £6 0 0 ($17.25) per annum, post free

Single Copies 10s. 6d. ($2.00) each, post free

Secretary
Lieut.-Colonel S. C. GUILLAN, T.D.

EDITORIAL AND ADVERTISEMENT OFFICES
4 Grosvenor Gardens, London, S.W.1
Telephone and Telegrams: SLOane 5928 & 6233

Editor of Publications
N. B. VAUGHAN, M.Sc., F.I.M.

Manager, Advertisement Department
E. R. MORT, B.Sc., F.I.M.

Entered at Stationers' Hall

1953, PART 6 | PUBLISHED MONTHLY | FEBRUARY

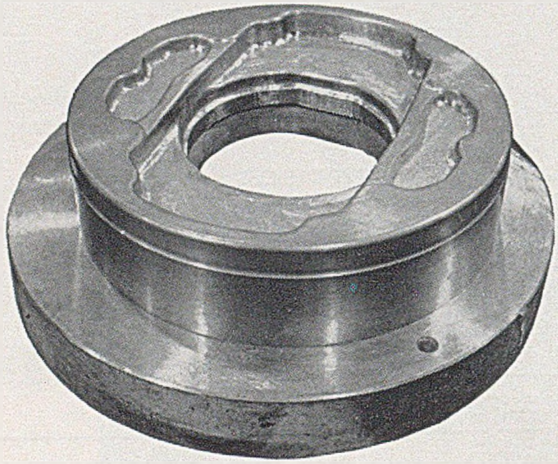
CONTENTS

BULLETIN

METALLURGICAL ABSTRACTS

JOURNAL

THE INSTITUTE OF METALS, 4 GROSVENOR GARDENS, LONDON, S.W.1



BULLETIN

OF THE INSTITUTE OF METALS

VOLUME I FEBRUARY 1953 PART 18

INSTITUTE NEWS

The Institute of Metals (Platinum) Medal

The Institute of Metals (Platinum) Medal for 1953 has been awarded by the Council to Professor GEORG MASING, of the Institut für allgemeine Metallkunde, Universität Göttingen, Germany, in recognition of his outstanding contributions in the field of metallography.

The medal, which is put at the Council's disposal by The Mond Nickel Company, Ltd., is awarded annually in recognition of outstanding services to the non-ferrous metal industries, whether on the industrial or scientific side, without distinction of race or country.

The Rosenhain Medal

The Rosenhain Medal for 1953 has been awarded by the Council to Dr. CHARLES ERIC RANSLEY, of the Research Laboratories, The British Aluminium Company, Ltd., Chalfont Park, Gerrards Cross, in recognition of his outstanding experimental and theoretical work on gas-metal equilibria.

The medal, which is put at the Council's disposal by Imperial Chemical Industries, Ltd., is awarded annually in recognition of outstanding contributions in the field of physical metallurgy made by persons under the age of 45 years, irrespective of nationality, sex, or membership of the Institute.

Capper Pass Awards

The Capper Pass Awards, which are made annually from a sum of £200 placed each year at the disposal of the Councils of the Institution of Mining and Metallurgy and of the Institute of Metals, by the Directors of Capper Pass and Son, Ltd., Bristol, are intended to encourage the publication of papers on processes and plant used in extraction metallurgy and on the subject of assaying in the *Transactions of the Institution of Mining and Metallurgy*, and of papers on processes and plant used in the fabrication of non-ferrous metals in the *Journal of the Institute of Metals*.

Provided that the Adjudicating Committee is satisfied that the quality and numbers of papers submitted, of the types for which the awards are made, reach a suitable standard:

(*a*) £100 per annum is available for one or more Awards to the authors of papers on some aspect of non-ferrous extraction metallurgy, including assaying;

(*b*) £100 per annum is available for one or more Awards to the authors of papers relating to some process or plant used in the extraction or fabrication of non-ferrous metals.

The Adjudicating Committee meets each year to consider papers published during the calendar year in the two journals referred to above.

The MSS. of suitable papers should be sent for consideration of the respective Publication Committees, addressed to the Editor of one or other of the two journals in the normal way.

1953 Spring Meeting: Visits to Works

Below are given some details of The Pyrene Co., Ltd., to whose works a visit is to be paid on Thursday, 26 March.

The Pyrene Company, Ltd., Brentford

The Pyrene Company was founded in 1914 to manufacture the well-known Pyrene fire extinguisher. The Company rapidly expanded its interests in fire and safety engineering, and has pioneered equipment and methods which have revolutionized fire-fighting technique in many fields; the development of foam branch-pipes and generators and proteinous foam-making compounds for oil-tank fires and air-crash fires, is one example. Besides the main factory at Brentford, there are now three subsidiary factories, one situated locally and the other two in South Wales. The central laboratories are at the main factory in Brentford, and it is there that the greater part of the research and development are carried out in all the Company's activities.

New fire hazards, necessitating the development of completely new extinguishants and techniques, are continually arising in modern industry. One recent example is the development of equipment and chemicals for the extinction of metal fires such as those due to magnesium, aluminium, sodium, &c. Much new equipment in the specialized field of aircraft fire safety, such as smoke and flame detectors, can be seen in the Aircraft Research Laboratory.

Apart from its fire and safety activities, the Company early developed other purely engineering interests and, among many other products, has manufactured motor-car bumpers since their adoption in this country many years ago. A feature of the Brentford factory is the quantity production of high-quality electroplated articles of relatively large size from steel through all the stages of pressing, pretreatment, electroplating, polishing, and assembly.

The Metal Finishing Division of the Company deals with a wide range of pretreatment and finishing processes for many metals. The best known of these are the Parkerizing and Bonderizing ranges, which the Company introduced to this country in the late nineteen twenties. Originally the Bonderizing processes were designed to provide corrosion-resistance and paint adhesion. Now, in addition, specially developed variations of them are also used to assist in the cold deformation of metals; e.g. in tube-drawing, wire-drawing, deep-drawing, and cold extrusion. The most recent addition to the Bonderizing processes for cold-forming is one for stainless steel. Besides their extensive use for protection against corrosion, some Parkerizing processes have also found application in anti-scuffing treatment on gears, tappets, valve guides, camshafts, &c., and a special process developed for this purpose, Parco-Lubrizing, is now in wide use for reducing wear on working parts, in the automotive industries. Examples of all these applications will be seen in the Company's laboratories.

Election of Members

The following 11 Ordinary Members, 1 Junior Member, and 22 Student Members were elected on 31 December 1952:

As Ordinary Members

Abdo, Ahmed Fathi, Head, Mechanical Department, Technical Secondary School, Cairo, Egypt.

Anstey, John Henry, B.Sc., Patent Department, The Mond Nickel Co., Ltd., Sunderland House, Curzon Street, London, W.1.

Bligh, Ernest W., Metallurgical Engineer, S. E. Opperman, Ltd., Boreham Wood, Herts.

Booth, Ernest K., B.Sc., Materials Engineer, A. V. Roe (Canada), Ltd., Gas Turbine Division, Box 430, Terminal " A ", Toronto, Ont., Canada.

Fast, Johan D., Chief Metallurgist, Research Laboratories, N. V. Philips' Gloeilampenfabrieken, Eindhoven, Holland.

George, Wilbert R., Student of Metallurgy, University College, Swansea.

Golda, Edward W., B.Sc., Technical Supervisor, Los Angeles Tube Division, Phelps Dodge Copper Products Corporation, 6100 Garfield Avenue, Los Angeles 22, Calif., U.S.A.

Johansson, Klas-Erik, Chief Metallurgist, Uddeholms Aktiebolag, Hagfors Järnverk, Hagfors, Sweden.

Lambilliotte, André Alphonse Joseph Emile, Administrateur-Directeur, Société Anonyme des Usines Gilson, La Croyère, Belgium.

Warland, Eugène, Ing. Civil des Mines, Administrateur-Directeur, Société Anonyme Métallurgique de Prayon, Trooz, Belgium.

Weldon, James W., 2315 Harrison Street, Kansas City 8, Mo., U.S.A.

As Junior Member

Skinner, Colin, B.Sc., Student of Metallurgy, University College, Cardiff.

As Student Members

Al-Samarrai, Shakir Haj Khalaf, Student of Metallurgy, University College, Swansea.

Brown, Darryl Anthony, Student of Metallurgy, Cambridge University.

Davies, Peter Malcolm, Student of Metallurgy, University College, Swansea.

Emmerson, Leslie Gordon, Student Metallurgist, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Fisher, Raymond John, Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Goodchild, Keith Trevor, Metallurgical Trainee, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Grant, John Michael Seafield, Student of Metallurgy, Cambridge University.

Heaton, Herbert Barrie, Student of Metallurgy, The Technical College, Bradford.

Hume, Gerald James Thomas, Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Lloyd, Brian Arthur, Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Markowicz, Jan Henryk, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Modlen, Geoffrey Frank, Student of Metallurgy, Cambridge University.

Parker, Robert, Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Pinney, Victor, Student of Metallurgy, Liverpool University.

Smith, Michael Duncan, B.Met., Student of Metallurgy, Sheffield University.

Smith, Robert, B.Sc., Spectrographer, Johnson, Matthey and Co., Ltd., 78 Hatton Garden, London, E.C.1.

Stolarczyk, Janusz Eugeniusz, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Szkopiak, Zygmunt Czeslaw, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Taberner, James Norman, Student of Metallurgy, College of Technology, Manchester.

Thomas, Raymond William, Foundry Metallurgist, Dartmouth Auto Castings, Ltd., Dartmouth Road, Smethwick 40, Staffs.

Thompson, Stewart Philip, Student of Metallurgy, Cambridge University.

Westrope, Alan Richard, Student of Metallurgy, University College, Swansea.

PERSONAL NOTES

Mr. W. Barr has been elected Honorary Treasurer of the Iron and Steel Institute.

Mr. J. Bateman has left Hard Metal Tools, Ltd., Coventry, and taken a post with Firth-Sterling, Inc., Pittsburgh, Pa., U.S.A.

Mr. L. J. Cartmell has joined the staff of Henry Wiggin and Co., Ltd., Birmingham.

Mr. G. J. Catterall has been awarded the degree of B.Sc. with Second Class Honours of London University and the A.R.S.M. He has now taken up an appointment with Group Four Metals, Ltd., London, E.14.

Mr. J. B. Chappell has taken a post at the Waunarlwydd Works of Imperial Chemical Industries, Ltd., Metals Division, South Wales.

Dr. W. F. Coxon has been appointed Managing Director of Arrow Press, Ltd., Watford, publishers of a number of technical journals.

Mr. N. H. G. Daniels has left England for the United States, where he is to undertake research at the Institute of Engineering Research, University of California, Berkeley, Calif.

Dr. C. H. Desch has been elected an Honorary Member of the American Society for Metals.

Mr. P. G. Dodd has been appointed Foundry Technical Officer to Austral Bronze Pty., Ltd., Alexandria, N.S.W.

Mr. C. R. Faulkner has left Birmingham University to take up an appointment in the Ministry of Supply at Culcheth, near Warrington.

Dr. G. E. Gardam, who until recently was Director of Research of the Design and Research Centre for the Gold, Silver, and Jewellery Industries, has taken up an appointment with Aluminium Laboratories Limited, Banbury.

Mr. R. H. Hannaford has taken up an appointment in the Research Laboratories of Fairey Aviation Co., Ltd.

Dr. Harold Hartley, Chairman of Radiation, Ltd., Birmingham, has been awarded the Melchett Medal of the Institute of Fuel.

Mr. T. L. Johnston has left Liverpool University, where he was recently awarded the degree of Ph.D., and is now at the Atomic Energy Research Establishment, Harwell.

Professor P. J. Lacombe has left the Laboratoire Central des Traitements Chimiques at Vitry (Seine) and has been appointed Professor of General Metallurgy and Director of the Centre de Recherches Métallurgiques in the Ecole Nationale Supérieure des Mines, Paris, in succession to Professor Crussard.

Professor R. H. Myers has been elected to the Council of the Australian Institute of Metals (Sydney Branch).

Mr. J. R. Parish has left Elliott Brothers, Ltd., Boreham Wood, and is now with Cathodic Corrosion Control, South Audley Street, London.

Mr. R. Parker, of Aluminium Laboratories, Ltd., has been transferred from Geneva to Banbury.

Señor Jorge A. Sabato has now returned to the Argentine after a visit to Great Britain. He wishes to thank the many members of the Institute who assisted in making his visit successful.

Professor Cyril Stanley Smith, Director of the Institute for the Study of Metals at the University of Chicago, is planning to visit various metallurgical research centres in England and Western Europe in February and March.

Death

The Editor regrets to announce the death of:

Mr. John Stanley Bowden, B.Met., on 28 November. He had been a Senior Metallurgist on the staff of the Tin Research Institute since the end of 1946.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

London Local Section

At a meeting of the Section held at the Royal School of Mines, London, S.W.7, on 6 November, Professor G. V. Raynor, M.A., D.Sc., Professor of Metal Physics at Birmingham University, gave a lecture on:

An Approach to the Theory of Ternary Alloys

The lecturer said that though further work was necessary to extend the present theory of binary alloys to solvents other than those of the first two groups of the Periodic Table, sufficient was known about alloy formation in copper and silver to justify an approach to the theory of ternary alloys based on these metals. The influence of known factors upon the details of ternary equilibrium diagrams could be traced, with particular reference to the shapes of the primary solid-solubility isothermals, and to the extent to which binary electron compounds projected into the ternary equilibrium model. In particular, a simple thermodynamical analysis was possible for solubility isothermals at which the primary solid solution was in equilibrium with a stable intermetallic compound. It could be demonstrated that there was a close similarity between the influence of the atomic size-factor on the crystal structures of binary 3/2 electron compounds, and its influence on the structures of the corresponding ternary phases.

These considerations could be used to interpret, and in favourable cases to predict, the homogeneity ranges of close-packed hexagonal and body-centred cubic 3/2 electron compounds in ternary systems, provided that no very stable intermetallic compounds between the two solute metals interfered. The influence of stable compound formation was illustrated by a comparison of the isothermal diagrams for the Ag–Mg–Zn, Ag–Mg–Sn, and Ag–Mg–Sb alloys. In the first case, the Mg–Zn compounds were of low heat of formation and exerted no influence on the equilibria in silver-rich alloys. The compound Mg_2Sn, of higher heat of formation, entered into equilibrium with the 3/2 electron compounds in the system Ag–Mg–Sn, and Mg_3Sb_2, of still higher heat of formation, prevented equilibrium between the 3/2 electron compounds in the system Ag–Mg–Sb, and itself entered into equilibrium with the primary silver-rich solid solution. The existence of ternary compounds had to be allowed for in developing the theory, and work was in progress to define the conditions under which they might be expected. Investigation of the Cu–Al–Sn and Cu–Al–In systems had indicated that such compounds might occur where any tendency to ordering in a binary intermediate phase was enhanced by the inclusion of a third metal. This led to interesting structural relationships between binary and ternary phases.

Scottish Local Section

At a meeting of the Section held in Glasgow on 10 November, Mr. R. A. F. Hammond lectured on:

The Properties and Engineering Applications of Electrodeposits

The lecture was primarily concerned with heavy nickel and hard chromium deposits as used in the engineering industries for salvaging worn or over-machined parts or for hard-surfacing.

After briefly reviewing the history of the subject and its present-day applications, the lecturer gave some information on the economics of the process. The techniques and equipment are similar to those employed in decorative electroplating, but the great importance of obtaining very strong adhesion and good mechanical properties of the deposits calls for conscientious and skilful workmanship. Much ingenuity is frequently required to secure reasonably uniform distribution of the thick deposits commonly employed in this class of work, for example, by the use of auxiliary anodes or cathodes.

Strong adhesion of the electrodeposit to the basis metal is of paramount importance. Given suitable cleaning and depositing conditions, the bond between the deposit and the base is atomic in nature, and slides were shown of photomicrographs of sections of electrodeposited metals in which the grain boundaries of the base were continued into the electrodeposit. This is possible for pairs of metals belonging to different crystallographic systems and for lattice spacings differing by as much as 15%. In such circumstances the adhesion of the deposit may equal or even exceed the tensile strength of the component metals.

Adhesion of, or approaching, this order is possible on most low-alloy structural steels and on many other metals and alloys used in engineering. For other materials, e.g. certain light alloys, special cleaning methods are necessary, and the adhesion is liable to be somewhat weaker; on a few materials only very weak adhesion is obtainable.

In addition to being strongly adherent, the electrodeposit must have suitable mechanical properties. The hardness of electrodeposits commonly exceeds that of the corresponding metal in the as-cast or rolled condition, and moreover the mechanical properties, including the hardness, can be controlled over a considerable range by adjusting the depositing conditions. The diamond pyramid hardness of electrodeposited nickel as normally applied ranges from 180 to 400 with U.T.S. values of 28 and 55 tons/in.^{2} and elongations of 35 and 6% respectively. Electrodeposited chromium has a diamond pyramid hardness usually ranging from 750 to 950 and virtually no ductility.

Most electrodeposited metals undergo a change in volume after deposition which sets up an internal stress in the deposit. In general, the stress is contractile, but some deposits, notably zinc and lead, show compressive stress. Nickel and chromium deposits are normally strongly contractile, but certain bright nickel deposits exhibit a slight compressive stress. The residual stress in these electrodeposits does not usually lead to practical difficulties, but may cause a marked reduction in the fatigue limit of the steel base, which could be serious in those limited applications in which alternately stressed components operate near the critical loading.

Heat-treatment of chromium-plated components between 150° and 300° C., as commonly employed for relieving hydrogen embrittlement, has been shown to cause a still further reduction in the fatigue limit of the steel, and the advantage of using a higher temperature, e.g. 440° C., as described by Logan, and also methods of minimizing the effect of fatigue-limit reduction by shot-peening and other means, were discussed.

Finally, a brief account was given of the behaviour of electrodeposited nickel and chromium as bearing surfaces.

OTHER NEWS

Conferences on " The Deep Drawing of Metals " and " Continuous Casting of Ingots "

The Department of Industrial Metallurgy, University of Birmingham, is arranging its annual two-day Conference this year on Monday and Tuesday, 16 and 17 March. The subjects will be " The Deep Drawing of Metals ", on 16 March, and " Continuous Casting of Ingots ", on 17 March. Full details of the programme may be obtained from the Secretary, Department of Industrial Metallurgy, The University, Edgbaston, Birmingham 15. No fees are chargeable for the attendance, but registration is required.

DIARY

Local Sections and Associated Societies

10 February. South Wales Local Section. " Dislocations in Crystals ", by Dr. B. A. Bilby (University College, Metallurgy Department, Singleton Park, Swansea, at 6.30 p.m.)

12 February. Liverpool Metallurgical Society. " Nucleation in Metals and Alloys ", by J. H. O. Varley. (Liverpool Engineering Society, The Temple, Dale Street, Liverpool, at 7.0 p.m.)

16 February. Sheffield Local Section. " The Constitution and Properties of Some Titanium-Base Alloys ", by W. A. Baker. Joint meeting with the Sheffield Society of Engineers and Metallurgists and the Sheffield Metallurgical Association. (The University, St. George's Square, Sheffield 1, at 7.30 p.m.)

18 February. Manchester Metallurgical Society. " Metallurgical Aspects of Lubricating Problems ", by A. L. H. Perry. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

25 February. Liverpool Metallurgical Society. Visit to The British Aluminium Co., Ltd., Warrington.

27 February. Birmingham Local Section. All-Day Symposium on " Making the Best of Metals ". Opening address by R. Lewis Stubbs. (College of Technology, Suffolk Street, Birmingham, at 10.15 a.m.)

27 February. North-East Metallurgical Society. " Steel Foundry Radiographic Practice ", by G. M. Michie. (Cleveland Scientific and Technical Institute, Middlesbrough, at 7.15 p.m.)

2 March. Sheffield Local Section. Annual General Meeting, followed by " The Economic Use of Non-Ferrous Metals ", by F. Hudson. Joint meeting with the Institute of British Foundrymen, Sheffield Branch. (Sheffield College of Commerce and Technology, Department of Engineering, Pond Street, Sheffield 1, at 7.30 p.m.)

3 March. Oxford Local Section. " John Dalton and Modern Thought ", by Stephen Toulmin. (Black Hall, St. Giles, Oxford, at 7.0 p.m.)

3 March. South Wales Local Section. " Recent Advances in Furnace Design ", by E. S. W. Eardley. (University College, Metallurgy Department, Singleton Park, Swansea, at 6.30 p.m.)

4 March. Manchester Metallurgical Society. " Precipitation-Hardening ", by Dr. H. K. Hardy. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

5 March. Birmingham Local Section. " Copper and Copper-Alloy Development ", by Dr. E. Voce. (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

5 March. Leeds Metallurgical Society. " Lubrication in Cold-Working Processes ", by A. E. McAulay. (Chemistry Department, The University, Leeds 2, at 7.15 p.m.)

5 March. London Local Section. " Dislocations in Crystals " by Dr. B. A. Bilby. (4 Grosvenor Gardens, London, S.W.1, at 7.0 p.m.)

Other Societies

12 February. Institute of Metal Finishing, North-West Branch. " The Electrodeposition of Tin and Its Alloys ", by Dr. J. W. Cuthbertson. (Engineers' Club, Albert Square, Manchester, at 7.30 p.m.)

16 February. Institute of Metal Finishing, London Branch. " Roughness of Nickel Deposits ", by C. J. Swanson. (Northampton Polytechnic, St. John Street, London, E.C.1, at 6.0 p.m.)

18 February. Society of Chemical Industry, Corrosion Group. " Some Recent Advances in Corrosion Research ", by Dr. W. H. J. Vernon. (The Stephenson Building, King's College, Newcastle-upon-Tyne, at 2.30 p.m.)

For Appointments Vacant, see p. *a*, facing Abstracts 448.

NEW TECHNIQUES OF METALLURGICAL RESEARCH *

INTRODUCTION

By Professor A. G. QUARRELL,† D.Sc., Ph.D., A.R.C.S., F.I.M.

The topic chosen is one with a very wide appeal. The industrial metallurgist feels that at a Symposium of this nature he may learn of methods of detecting defects in his product before too much work has been put into it, or may hear of a new technique that can be adapted to control one of his processes. The research metallurgist knows that a new technique may speed up a research many times, or even make possible what could not previously be done at all.

It has been said that most of the great advances in science have been associated with the introduction of new techniques. This may be going too far, but every really original development must widen our horizon and enable us to concern ourselves directly with phenomena formerly hidden from us. Techniques are therefore of the utmost importance in research at every stage: in data collecting, when the working hypothesis is being formulated, and again when the hypothesis is being subjected to experimental test. Indeed, a research may be held up for a long time while a suitable means of conducting the crucial experiments is being worked out. This usually happens when the technique required is novel in a qualitative sense.

Advances in technique may be regarded as qualitative or quantitative. The qualitative advance makes available information of an entirely new kind, whereas quantitative advances enable us to know more exactly what could have been known approximately with existing techniques.

Good examples of the former are provided by metallography and X-ray crystallography. The introduction of metallography enabled men to see for the first time how crystals were arranged in metals and the different phases that were present. The later development of X-ray methods provided a means of determining how atoms are arranged within crystals. Both of these were qualitative advances, and both have greatly improved our understanding of the nature and properties of metals.

Improvements of technique in the quantitative sense may also have beneficial effects, for knowledge that is insufficiently precise is usually confusing. Indeed, many of the laws of science could not be elucidated until techniques of sufficient accuracy and sensitivity had been worked out. If a graph is drawn to represent the growth of metallurgical knowledge with time, it will consist of a series of slopes separated by abrupt steps. The steps represent the "qualitative" developments in techniques, whereas the slopes correspond to the steady improvement that goes on almost continuously. One gains the impression that these slopes are getting steeper and that the advances gained through the improvement of existing techniques are becoming ever more important.

We should all like to know what will mark the next really big step. Speculation is probably idle, but it may lie in the full exploitation of the possibilities opened up by radioactive isotopes. Much more development work needs to be done in this field, but, in principle at least, we are now able to follow the movements of different kinds of atoms within the crystal lattice—something not previously possible.

Such discoveries are, by their nature, few and far between, but there is still plenty of scope for the improvement of existing techniques. Two of the papers to be discussed are concerned with metallography, illustrating that the basic technique which has formed the metallurgist's most powerful method of investigation for almost ninety years is still capable of improvement. Considerable advances have taken place in this field during the last few years; they include phase-contrast microscopy, the fuller application of polarized light, multiple-beam interference methods, the examination of surface profiles by means of tapered sections, and I suppose we could also include microhardness testing. Much work has aimed at improving the methods of preparing metals for microscopic examination, particularly with regard to the time taken. Thus, in principle, the methods used by Sorby to examine polished and etched specimens in 1864 were identical with those in use to-day, but the times taken to polish a specimen were much longer. Now the time required to prepare a satisfactory specimen is measured in minutes, whereas it took a whole week to produce a surface to satisfy Sorby. However, his standards were very exacting.

In their student days metallurgists are required to draw the structures they see through their microscopes. This method is valuable because it ensures that detailed consideration is given to the structures observed, but I wonder how many metallurgists realize that even in the early years of the present century this was the only method by which a microstructure could be recorded. Photography has come to play a tremendously important role in modern metallurgy, and it would probably not be going too far to assign to it the importance of a step on the graph of progress. Spectrography, cinematography, high-speed cinematography, metallography, X-ray crystallography, radiography, are all possible in their modern form only because of the photographic technique, and there can be no doubt that in its various forms it has greatly speeded up the acquisition of new knowledge.

---

* Papers forming the basis of a Symposium organized by the Birmingham Local Section of the Institute of Metals, and held in Birmingham on 29 February 1952.

† Professor of Physical Metallurgy, University of Sheffield.

Mention of photography serves to remind us that many of the techniques we use were introduced for quite different purposes and have been subsequently adapted for metallurgical work. To-day, workers in all branches of science are on the look-out for new techniques that will help them to solve outstanding problems or enable them to obtain information more quickly or more precisely, and they realize that developments in quite a different field may provide them with the method they are seeking. One beneficial effect of this is that in an age of increasing specialization, in which it is becoming increasingly difficult for a worker in one subject to understand and keep abreast of developments in another, they share a general interest in techniques, and scientists of many types meet to discuss common problems. This is well illustrated by the various groups of the Institute of Physics.

Some of the principles that can be used in devising new techniques are so complex that a new career, that of "technique specialist", seems to be growing up. In a sense he is a scientific mechanic, who, when he is told what requires to be done, will work out a suitable method. Having done so, he has no further interest in it, except that the experience he has gained will help him to tackle somewhat similar jobs with greater confidence and efficiency in the future. I think it is fair comment that the "technique specialist" generally produces a solution which is far from simple and which requires elaborate and expensive apparatus.

One of the outstanding changes in our research laboratories in the last twenty years is the great increase in quantity and in complexity of the apparatus used. While recognizing that the days of sealing-wax and string have gone for ever as far as metallurgy is concerned, I hope we shall not go too far in the other direction and come to regard complexity of technique and elaborate apparatus as essential to research.

Let us judge a technique, not in terms of the skill shown by the man who developed it, but by what that technique will enable us to do and by the new horizons it opens up, and let us put a premium on simplicity. The simpler the better, so long as it is capable of doing the job that is to be done; a complicated technique is justified only if it is quite impossible to do the same thing more simply. Our discussion should help in achieving this object, for simplification and improvement come from the free interchange of ideas among men with common interests.

THE USE OF DIAMOND DUST FOR METALLOGRAPHIC POLISHING

By L. G. TOTTLE,* A.R.Ac.S., L.I.M.

Historical Survey

The use of diamond dust as a polishing medium for jewels and hard-metal carbides has been well known for some time, but its application to the preparation of metallographic specimens is comparatively recent. Woodside and Blackett[1] in 1947 described the use of diamond dust for polishing specimens of molybdenum carbide and cast molybdenum, previously prepared by surface grinding on a diamond-impregnated wheel. Tarasov and Lundberg[2] in 1949 demonstrated how it might be used for specimens of high-speed steel which had first been hand polished on a diamond hone in place of the conventional emery cloth.

More recently attention has been drawn by Perryman[3] to the use of diamond dust for polishing metallographic specimens. He described the preparation of specimens of galvanized coatings, nickel + chromium-plated brass and aluminium–silicon alloy, oxide films on copper, and hard inclusions in aluminium-base alloys. The specimens were prepared in the usual manner by grinding down to 4/0 emery paper before being polished on a Selvyt pad impregnated with diamond dust. The diamond dust had a particle size of 0–1 μ, and the Selvyt pad, moistened with white spirit, was used on a special disc. The disc had a raised rim and revolved at a fairly slow speed, so that the diamond dust should not be thrown off the pad and wasted.

Preliminary Treatments

In the research laboratories of The Mond Nickel Co., Ltd., a very large number of ferrous and non-ferrous microspecimens, covering a considerable range of hardness and having a wide variety of structures, have to be prepared each week. The floor space available for metallographic polishing is limited, and therefore it has been necessary to introduce new techniques with a view to reducing the time required for preparing each specimen before micro-examination. In addition to the use of diamond dust as a medium in the final stages of polishing, wet grinding on rotating emery-paper discs and automatic polishing have accordingly been introduced.

For most of the microspecimens, rubbing by hand on emery paper has been discarded. After flat surfaces have been obtained on a grinding machine, they are held by hand on a succession of silicon carbide papers of different degrees of coarseness, attached to 7½-in.-dia. horizontal discs revolving at 520 r.p.m. A water-resistant type of paper is used, which is flooded with water during the polishing operation, this serving to keep the specimen cool and to wash off any loose particles of abrasive powder. The cutting action of these papers is very swift indeed, and a specimen may be prepared from a surface-ground finish to 600 grit (approximately 4/0), suitable for pad polishing, in under 2 min. Two of the polishing wheels are shown in Fig. 1, the left-hand one being ready for use and the other opened for changing the paper disc. A special jig has been developed to enable three ⅝-in.-dia. specimens, or more of smaller diameter, to be prepared at the same time. An assembled jig containing three specimens and the component parts of a jig are shown in Fig. 2.

For final polishing on rotating pads, automatic polishing

---

* Technical Assistant, The Mond Nickel Co., Ltd., Research and Development Laboratory, Birmingham.

machines, modified slightly from the design introduced by The British Non-Ferrous Metals Research Association, have been installed (Figs. 3 and 4). In this type of machine the specimen is traversed mechanically across the face of the rotating pad, while being turned through approximately 90° at each traverse. The pressure on the specimen may be varied at will, but is controlled and is uniform throughout the polishing operation. The time required to polish a specimen mounted in Bakelite, or multiple specimens in the special jig, is about 15 min. At the present time four of these machines are operating almost continually.

Diamond-Dust Experiments

Many of our samples, e.g. those which had been coated with zinc or nickel, complex alloys, &c., could not be polished satisfactorily in a reasonable time by the established abrasives, so that the publication of Perryman's work led to an immediate investigation of the potentialities of diamond dust.

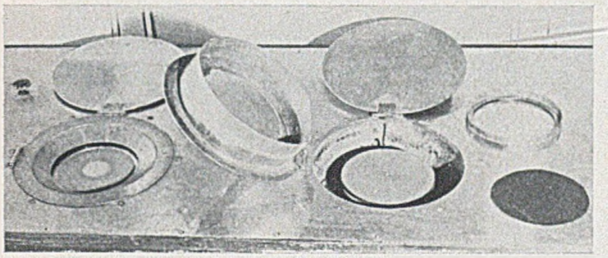


Fig. 1.—Equipment for Grinding on Wet Silicon Carbide Papers.

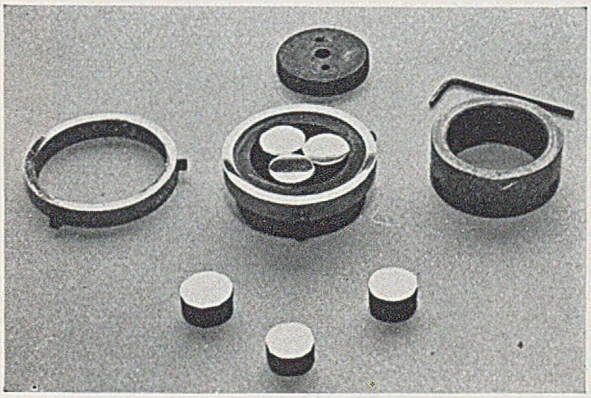


Fig. 2.—Jigs for the Polishing of Bar Samples.

The initial experiments showed that satisfactory results could be obtained on 7½-in.-dia. rimless wheels rotating at 190 r.p.m. Speeds up to 520 r.p.m. are now in use, without noticeable loss of diamond dust. No special apparatus has therefore been required.

About 0·025 g. of diamond dust is first rubbed into the dry Selvyt cloth attached to the wheel, and it is then sparingly lubricated with paraffin or anhydrous alcohol. Needless to say, the pad is not washed out after use, but is carefully preserved in a stoppered bottle if it has to be taken off the wheel. Diamond dust has been used for the polishing of specimens by hand and on the automatic polishing machines with very good results.

The following are a few examples where diamond-dust polishing has proved particularly valuable. Galvanized cast-iron samples are difficult to polish by conventional methods. Fig. 5 (*a*) shows a carefully prepared section polished on γ-alumina. Staining and considerable differences in levels of the zinc, ferrite, and graphite are apparent. Fig. 5 (*b*) shows a flat unstained field obtained by the use of diamond dust. The surface is, however, covered with very fine scratches, and this has been found to be a feature of diamond-dust polishing. These scratches can be removed by a polish of short duration on light magnesia or γ-alumina.

It is frequently necessary to prepare specimens of corroded and scaled materials with a view to examining the interface between the deposit and the base material. These deposits differ considerably in hardness from the body of the sample and, when conventional techniques are used, they are often worn down or broken away. By diamond-dust polishing satisfactory specimens are usually obtained at the first attempt,

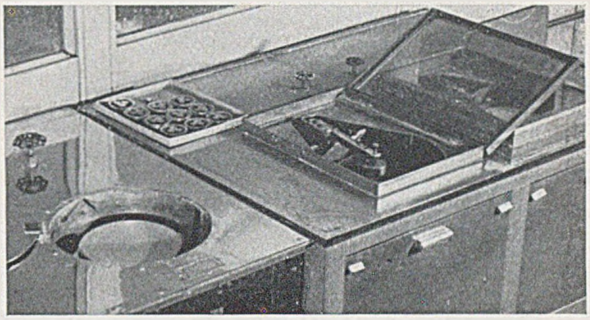


Fig. 3.—Modified Automatic Polishing Machines.

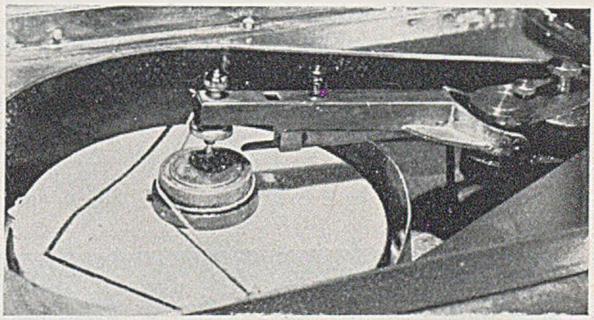


Fig. 4.—Close-up of Modified Automatic Polishing Machine.

thus saving a great deal of time and energy and requiring less skill on the part of the operator. Fig. 6 is a photograph of a section from a corroded nickel–chromium alloy bearing an oxide deposit. This specimen was prepared from the rough, i.e. including grinding operations, in 22 min.

In Fig. 7 (*a*) is shown an oxide film on a Nimonic alloy. This specimen was prepared for polishing by depositing nickel over the oxide film to protect the edge of the section. It was polished with care on alumina, but the oxide film is badly damaged. Fig. 7 (*b*) illustrates a similar specimen prepared on diamond dust. In this sample the section has been mounted in Bakelite without extra protection of the edge and has been polished quite successfully. The time which would be spent in plating the specimen has thus been saved, and a better result achieved.

Complex nickel-base alloys are difficult to prepare by normal means without the introduction of considerable relief or without pulling out some of the constituents. They are also difficult to etch successfully, so that they have to be

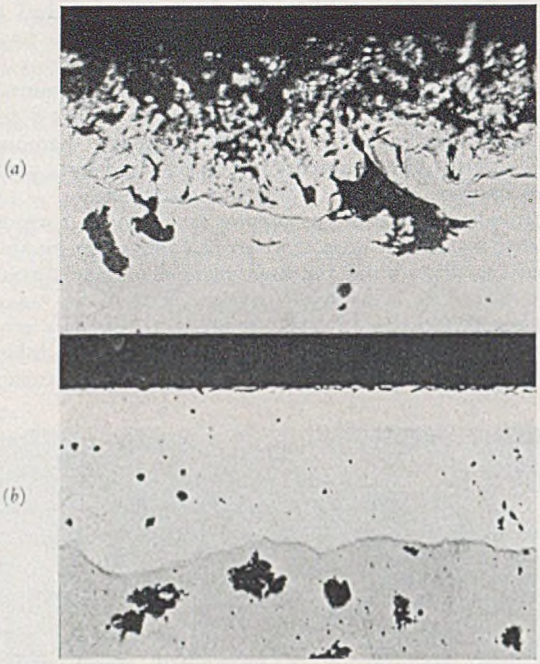


FIG. 5.—Galvanized Nodular-Graphite Cast Iron. Unetched. × 500.
(a) Polished on γ-alumina. (b) Polished on diamond dust.

examined unetched with the aid of phase-contrast illumination. By this method of illumination only very slight differences in level can be tolerated, and it is for this type of specimen that diamond-dust polishing proves especially useful. Fig. 8 is a photograph of an alloy containing three phases differing considerably in polishing and etching characteristics, which has been photographed in the unetched condition, using phase-contrast illumination.

For joining high-temperature-resistant materials, precious-metal brazing alloys have been developed. Fig. 9 shows a deposit of one of these alloys on Nimonic. Once again there is a considerable difference between the polishing characteristics of the base material and the brazing alloy, but the photograph shows that a satisfactory flat field has been obtained right across the interface between them.

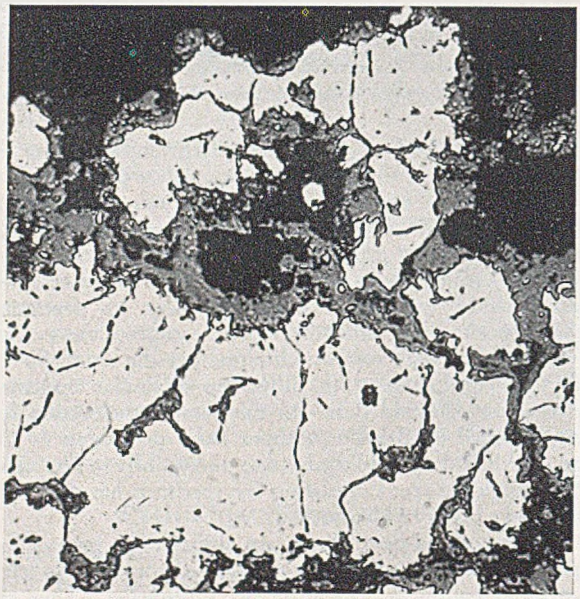


FIG. 6.—Corroded Nickel-Chromium Alloy. × 400.

As a final example, the profiles of metal powders are easy to examine by transmitted light, but it is their cross-sections which are often most informative. When such powders are mounted in finely sieved Bakelite and polished first on fine papers and then on diamond dust, a good specimen may quickly be obtained. Fig. 10 is a photograph of an iron powder produced from iron wire by means of a metallizing pistol. Oxide skins on the particles and small globules of oxide within them may be clearly seen.

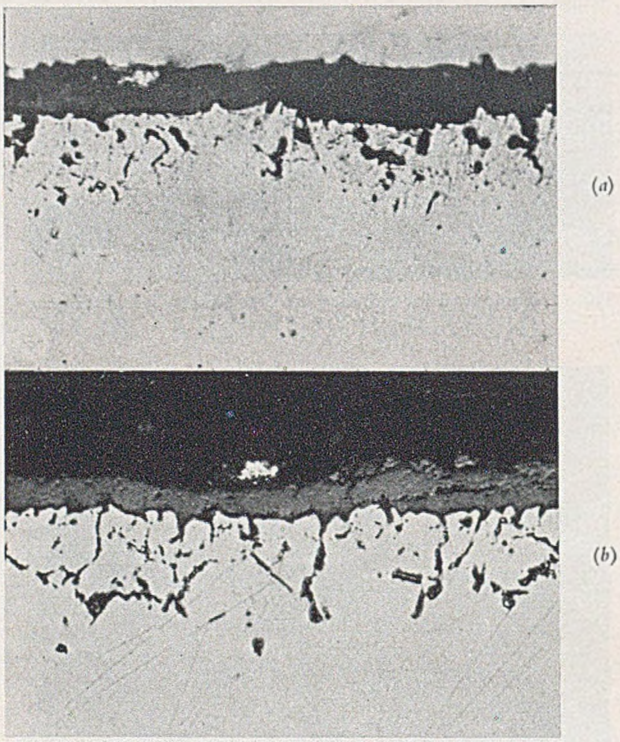


FIG. 7.—Oxide Films on Nimonic Alloy. × 1000.
(a) Polished on γ-alumina. (b) Polished on diamond dust.

Many other applications have been found for the diamond-dust polishing technique, but space does not permit of further illustration.

Advantages of Diamond Dust

It is probable that the versatility of diamond dust as a polishing medium is due to its great hardness, which allows it to cut through particles of widely differing hardness at much the same rate and enables the abrasive particles to retain a good cutting edge for a considerable time. Electron-micrographic examination of the diamond dust shows it to have a sharply angular form such as is illustrated in Fig. 11 for powder of 4–8 μ particle size. The uniform particle size of the diamond dust undoubtedly assists greatly towards the quality of the finish produced.

A disadvantage associated with the use of diamond dust as a polishing medium is that, because of its great cutting power, all samples show a network of fine scratches. These do not hinder low-power examination, but for high-power examination or for photography they must be eliminated by a further short period of polishing on alumina or light magnesia. The duration of this treatment must obviously be a compromise between the desire to obtain a scratch-free field and to preserve flatness and absence of objectionable relief in the sample. From the point of view of freedom from scratches, the final finish is therefore not always equal to that obtained by highly skilled operators on materials for which specialized techniques have been evolved. The main advantages of diamond-dust polishing lie in the good finishes that can be obtained by less highly skilled polishers in short periods of time and the flatness of fields containing phases of widely differing hardness.

8

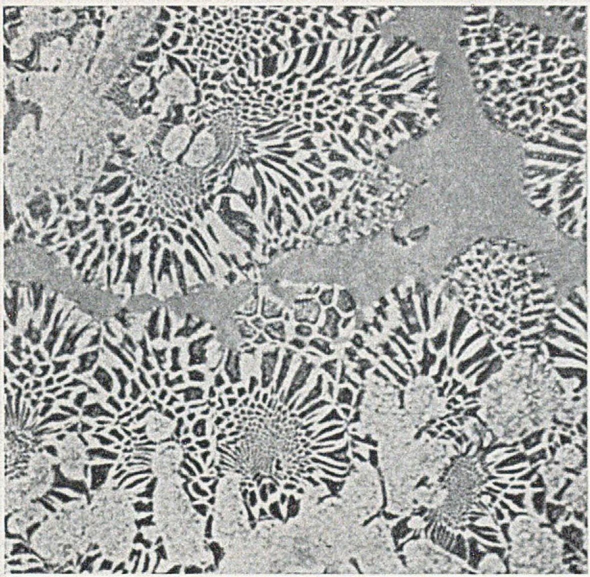


Fig. 8.—Complex Nickel-Base alloy. Photographed, unetched, by phase-contrast illumination. × 250.

9

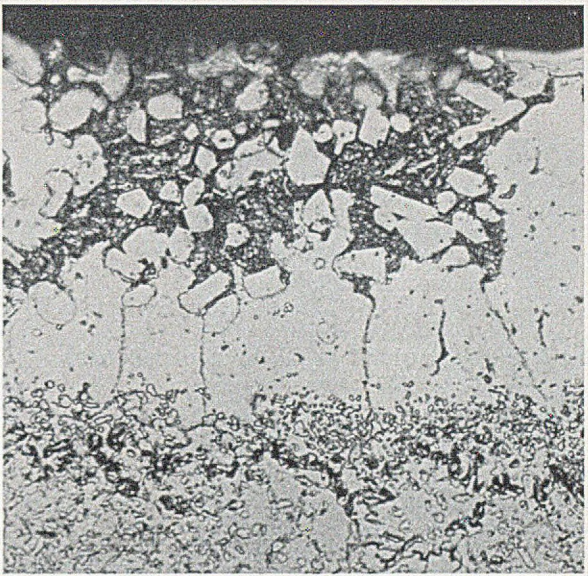


Fig. 9.—Deposit of Precious-Metal Brazing Alloy on Nimonic. × 400.

10

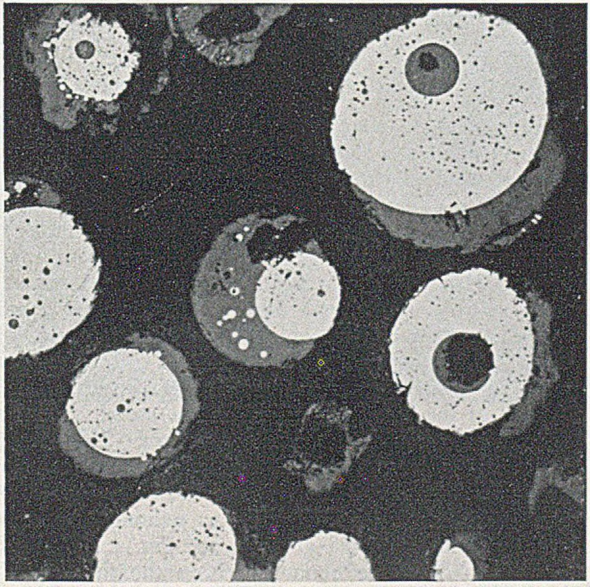


Fig. 10.—Iron Powder. × 400.

11

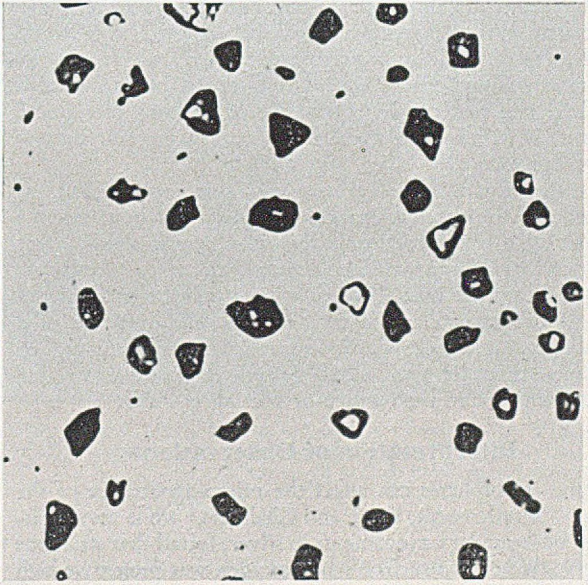


Fig. 11.—Diamond Dust, Particle Size 4–8 μ. × 1000.

The initial cost of diamond dust is fairly high, but so very little is required and it lasts for so many samples, that on these grounds alone it becomes a useful proposition. When time saved in the preparation of otherwise difficult specimens is taken into consideration, however, it becomes most attractive.

This paper has attempted to show how simple it is to introduce the diamond-dust technique using normal equipment. All that is necessary is a supply of diamond dust, preferably of the finest grade and of uniform particle size, a Selvyt pad, and a supply of paraffin or other light-oil lubricant. The speed of the pad is not critical, provided that the dust is not flung off as it revolves, and this may be avoided by rubbing the diamond dust into the pad.

It is understood that supplies of polishing paste, in which the diamond dust (of 0–2 μ size) is held in a petroleum-jelly base, are on the market in this country. In America devices similar in action to the familiar grease-gun are available, so that the amount of paste to be applied may be carefully controlled.

Acknowledgements

The author wishes to express his thanks to the Directors of The Mond Nickel Co., Ltd., for permission to present the paper. His thanks are also due to Mr. H. W. G. Hignett and Mr. H. Evans for encouragement and help throughout the work.

He is grateful to Mr. Hyneman of Industrial Photographs, Ltd., Four Oaks, for preparing a Vectrograph of the polishing machines and to Mr. Phillips of L. M. van Moppes and Sons, Basingstoke, for supplying photographs of diamond powders.

REFERENCES

1. G. C. Woodside and H. H. Blackett, *Metal Progress*, 1947, **51**, 945.
2. L. P. Tarasov and C. O. Lundberg, *ibid.*, 1949, **55**, 183.
3. E. C. W. Perryman, *J. Inst. Metals*, 1950, **77**, 61.

USES OF THE HOT-STAGE MICROSCOPE

By P. J. E. FORSYTH,[*] A.I.M.

Problems Involved in High-Temperature Microscopy

THE microscopical examination of hot metal surfaces involves a number of factors which tend to lower the resolving power of the optical system. The main problem is that of obtaining a large working distance and yet retaining a useful degree of resolution. Naturally, the examination of a hot metal surface by any optical instrument necessitates a limiting safe working distance and a protective cooling system for the front lens. As the specimen will be enclosed in a small furnace, an observation window is necessary, which adds to the optical limitations. In spite of these detrimental factors the definition that can be obtained even with simple optical systems, is sufficient to make high-temperature microscopy a useful technique.

It is not proposed to deal here with these optical systems, but one of the most useful developments has been that of the reflecting microscope objective which possesses the great advantage of a large working distance.

Apart from the optical defects which may be encountered while working at elevated temperatures, the retention of a metallographic surface finish is important, and staining or oxidation must be avoided. This will require a controlled atmosphere within the furnace or heating *in vacuo*.

Interpretation of Observations

Some atmospheres can affect the micro-appearance of the metal. Shuttleworth, King, and Chalmers,[1] while investigating grain-boundary movements in silver, found that striations were visible on the surface when oxygen was present, which were not observed in a nitrogen atmosphere. At this point it must be emphasized that one is examining a free surface of the metal, and the surface appearance should not necessarily be taken as indicative of the underlying structure. This point is well illustrated in Fig. 1, which would seem to show boundary films of a second phase. The structure is that of a leaded brass which has been heated to 500° C. During re-cooling it appeared as if the grain boundaries were being penetrated by the lead. On repolishing, however, it was found that the lead had only migrated along the surface grooves produced by the thermal etching of the boundaries. A similar case is shown in Fig. 2. This specimen is an aluminium–8% silver alloy which was electropolished and then heated to 300° C. It was found on repolishing that the γ phase had existed only as a thin film on the surface, and this is not indicative of the amount of the phase present in the alloy. Thus the surface appearance, e.g. striations, boundary penetration, and the presence of a second phase, does not necessarily represent the true structure of the metal, as the structure is often influenced by the energy associated with the free surface and that of the boundaries meeting the surface.

Applications of Technique

An application which is receiving much attention at the moment is to the study of recrystallization and boundary migration. This phenomenon lends itself well to hot-stage microscopy, as the boundaries in the metal usually show up quite clearly after heating. This may be due to more rapid volatilization of the metal in the boundary regions or to the boundary tension producing a groove on the surface. If the boundary moves to a new position, a new groove is formed, and the old groove may persist or completely heal. There is no doubt that surface migration of the metal atoms plays an important part in this healing process. The persistence of the old boundaries is often useful, as it reveals the amount of migration that has occurred. This persistence may be due to loss of material by volatilization, to staining, or even to precipitation. Fig. 3 shows a surface where boundaries have persisted owing to staining and the temporary retention of the grooves, and Fig. 4 illustrates precipitation where the boundary meets the surface. Fig. 3 also shows small craters where low-melting-point particles have liquated. In fact, the whole grain-boundary network was observed to liquate before the grains themselves. This suggests another potential use for the technique, i.e. the detection of small quantities of constituents or impurities of a lower melting point than the matrix. This is especially useful if the impurities exist as boundary films which may be hard to detect by normal microscopic methods. As soon as a constituent or region of segregate liquates, this can be detected, because the molten liquid sinks or forms a meniscus and appears to the observer as a small crater or groove.

---

[*] Metallurgist, Royal Aircraft Establishment, Farnborough, Hants.

Another line of investigation would be to employ the hot-stage technique together with a mechanism for stressing the specimen while under observation. The combination of the two techniques should be useful in the study of creep phenomena. An apparatus was built at the R.A.E. to study the effect of stress on boundary migration. A small ribbon of zinc was electropolished and then heated while being stressed. It was quite easy to follow the boundary movements at a magnification of ×250 (16-mm. objective) with very little loss of definition. Fig. 5 shows a moving grain boundary as seen at 350° C. Any form of deformation which produces differences in surface level can be examined continuously, i.e. it does not require etching to reveal the extent of the microstructural changes, and therefore is very suitable for the hot-stage technique. This is illustrated in Fig. 6, which is of a zinc specimen subjected to creep stresses at 350° C. The boundary positions at different time intervals are not in this case revealed by thermal etching or by precipitation, as in previous examples, but by the boundary slip that has occurred when the boundary was in each position.

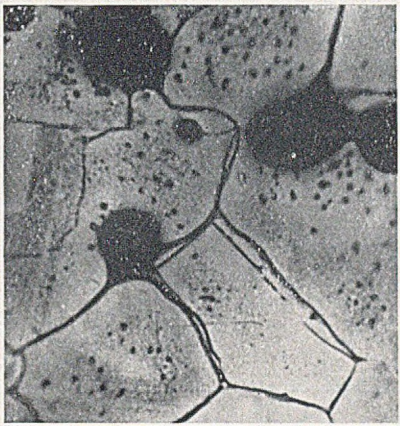


FIG. 1.—A Leaded α-Brass, Showing Boundary Films of Lead Existing at the Surface Only. × 500.

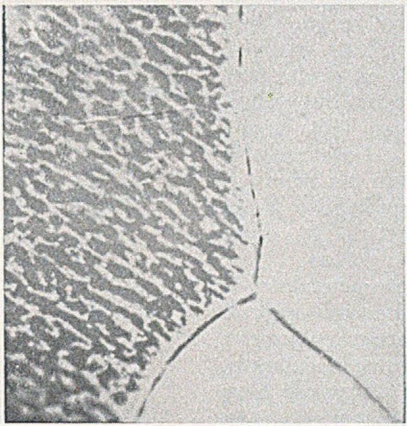


FIG. 2.—Aluminium-8% Silver Alloy, Showing Surface Precipitation. × 1000.

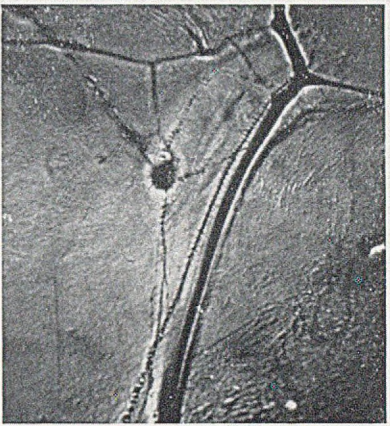


FIG. 3.—Grain-Boundary Residual Markings in Zinc, and Surface Craters Where Impurities Have Melted. × 100.

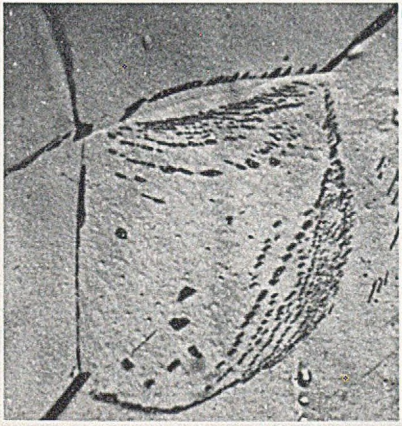


FIG. 4.—Surface Precipitation in Aluminium-8% Silver Alloy at Regions Where Grain Boundaries Meet the Surface. × 1000.

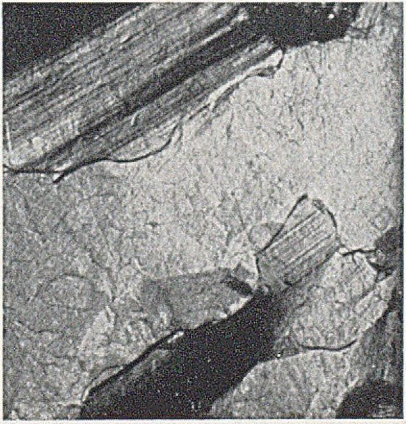


FIG. 5.—Moving Grain Boundaries Observed in Zinc at 350° C. × 100.

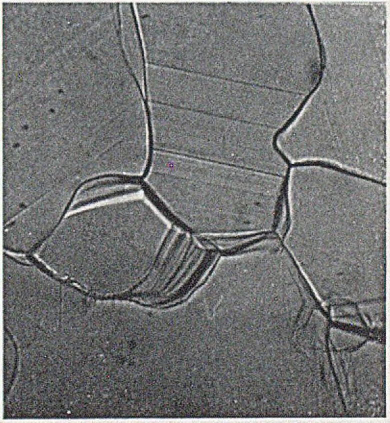


FIG. 6.—Boundary Positions in Zinc Delineated by the Flow That Has Occurred in Each Position. × 1000.

It is interesting to note that Figs. 4 and 6 provide evidence of the intermittent migration of grain boundaries.

Conclusions

The hot-stage technique for examining metallographic specimens may be very useful for investigating recrystallization and boundary migration, and for detecting segregation in alloys, low-melting-point impurities such as boundary films, &c. The technique may also prove valuable for examining solid-phase transformations which produce constituents of a different colour or form, or if there is a bulk change that will involve relief of the constituents. However, the limitations of the technique must not be overlooked, particularly as the complicating factor of the free surface may lead to erroneous observations and conclusions.

REFERENCE

1. R. Shuttleworth, R. King, and B. Chalmers, *Nature*, 1946, **158**, 482.

THE USE OF GEIGER COUNTERS IN X-RAY-DIFFRACTION STUDIES

By G. K. WILLIAMSON,* B.Sc., Ph.D.

Introduction

X-RAY diffraction has been for many years a well-established technique in many branches of metallurgical research. Much of its initial success was due probably to the simplicity of the equipment necessary to give qualitative results, which enabled the techniques to be tested without excessive expenditure. In very few laboratories are X-ray studies now carried out using the simple equipment originally installed. Precision-made multi-purpose cameras have very largely replaced the sealing wax and retort stand, and more reliable automatic X-ray sets the simple but unstable gas tubes. It is not surprising, therefore, that the recording of X-ray intensity by means of films is being replaced by direct detection, using Geiger counters and similar devices.

Much work has been done to explore the possibilities of Geiger counters, although their acceptance and usage is fairly limited, particularly in Europe. Counters are superior to films in most applications of interest to the metallurgist, although it is unlikely that they will ever supersede films in cases where a two-dimensional pattern is required, as in the Laue method. The following review may help potential users to gauge the advantages of this new technique for their own problems.

The Principles of Counters †

Geiger counters (and also proportional counters) are a logical development from the ionization chamber used in many of the classical X-ray studies. X-ray photons, entering any ionization chamber, produce ionization during their absorption. Approximately 10^2 ‡ ion pairs/photon are produced, and these drift towards the electrodes, where they collect, giving up their charge of about 10^{-17} coulombs. Since ionization chambers have a capacity of the order of 10^{-11} farads, there is a voltage change on the electrodes of approximately 10^{-6} V./photon. If the ionization chamber were used in the manner shown diagrammatically in Fig. 1, where a voltage V is maintained between the electrodes sufficient merely to prevent recombination of the ion pairs, any sudden changes in the voltage between the electrodes would be transmitted to the detector D by the condenser C_1 (the resistance R_1 being made sufficiently large to prevent the charge leaking away too quickly). In principle, the detector D could be used to count each ionizing event, but the difficulties of amplifying a voltage " pulse " of 10^{-6} V. renders this impracticable. It is possible, however, to make the ionization chamber do some of the amplification merely by raising the voltage V. At a certain value of the voltage field the electrons, during their acceleration towards the anode, may acquire sufficient energy to produce further ionization, so that the initial N ion pairs result in NA electrons being collected, producing a voltage pulse of $NA \times 10^{-8}$ V.; A is called the gas amplification.

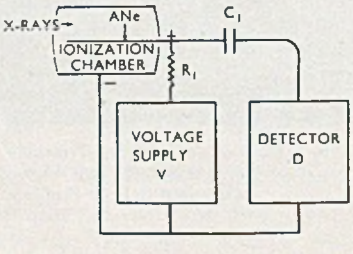


FIG. 1.—The Basic Circuit for Ionization Chambers and Counters. In the case of self-quenching Geiger counters R_1 should be in the range 1–10 megohms.

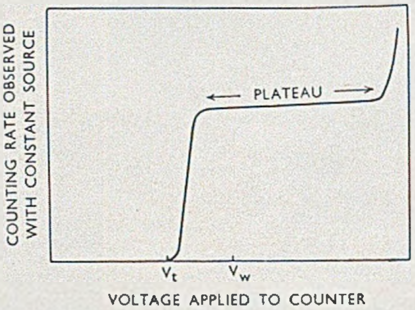


FIG. 2.—A Typical Geiger-Counter Plateau Curve. The plateau usually has a slight slope of less than 0·1% change in response per volt, necessitating some stabilization of V_w, the working voltage.

A chamber working under these conditions is called a proportional counter, since the output pulse is proportional to the energy of the ionizing radiation up to values of A of about 10^4 (Hanna *et al.*[6]). By the use of complicated circuits discrimination against unwanted radiation is possible (Curran *et al.*[7]; Arndt and Riley [4]; Lang [5]).

At still higher fields the gas amplification begins to " saturate ", until finally, in the Geiger region, approximately 10^9 ion pairs are produced, irrespective of the initial number due to the ionizing event. This region is determined by plotting counting rate against applied voltage, and a plateau, two or three hundred volts long, is found where the response is substantially constant, as in Fig. 2. The working voltage V_w is at the lower end of the plateau, usually about 80 V. above the threshold voltage V_t. Depending on the counter design, this voltage may lie anywhere in the range 500–2500 V. The Geiger region is useful because the voltage pulse per ionizing event is approximately 10 V., and this can be detected and counted with little or no amplification by robust and stable circuits. Although it is the Geiger counter with which this paper deals, many of the remarks apply equally well to proportional counters.

---

* Department of Metallurgy, Birmingham University.

† For fuller descriptions of the mechanism of counting see Korff,[1] Wilkinson,[2] and Sharpe and Taylor.[3]

‡ The number of ion pairs is a function of the energy of the ionizing radiation. For a fuller discussion, see Arndt and Riley [4] and Lang.[5]

Counters usually have cylindrical geometry, consisting of a tubular cathode and concentric fine-wire anode. For X-rays the counter is filled to a pressure of about one atmosphere with a noble gas,* usually argon, together with small additions of a polyatomic gas. The X-rays enter through a thin window of glass, mica, or aluminium, usually parallel to the counter axis, though in some designs it is perpendicular.

After an ionizing event an avalanche of electrons falls on the central wire, leaving a sheath of positive ions, which moves relatively slowly towards the cathode. This ion sheath reduces the field to a value below the threshold (and is responsible for the saturation of the gas amplification), thus rendering the counter totally insensitive to further ionizing events for a period known as the "dead time", usually about 10^{-4} sec. The counter slowly recovers its sensitivity until the positive ion sheath reaches the cathode, when it becomes fully sensitive once more. Another electron tends to be released with this neutralization, causing the cycle of events to be repeated, unless the release is suppressed or "quenched". The addition of traces of a polyatomic gas makes the counter self-quenching at lower counting rates, but it is better practice to reduce the anode voltage below V_t by a quenching circuit. If this "imposed dead time" is greater than the recovery time of the counter, the stability, the plateau slope, and the life of the counter are greatly improved.

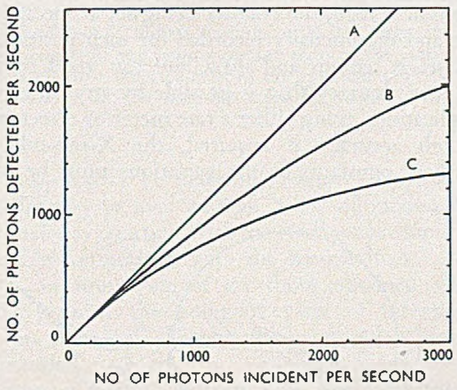


Fig. 3.—The Effect of Counting Losses on the Responses of Counters. Curve *A* is the ideal linear response (obtained at these rates with proportional counters). Curve *B* is the response of a Geiger counter with a 200 μ sec. dead time to a non-fluctuating source of X-rays. Curve *C* is the response of the same counter operated from a pulsed source, such as a self-rectifying X-ray tube running on A.C. The response is linear within 10% for counting rates of up to 2000/sec. for *B* and 1000/sec. for *C*.

Since their arrival is random, some photons may enter the counter when it is insensitive and be lost. For qualitative work these losses become serious at high counting rates, as shown in Fig. 3, and the counter must either be calibrated. or a correction made for these losses † (Schiff[10]; Blackman and Michiels[11]; Hall and Williamson[12]).

A commercially available quenching circuit, which has a predetermined and very stable imposed dead time allowing accurate counting-loss corrections, has been described by Cooke-Yarborough *et al.*[8], and its use, for the reasons given above, results in a great increase in the stability of the counter and in reliable and reproducible calibration. One of the serious effects which may result if external quenching is omitted is a violent increase in plateau slope (Parratt and Hempstead[9]), and there is little point in avoiding the use of this type of circuit for any but the most qualitative applications.

Circuits for Use with Quenched Counters

(a) *The Voltage Supply* (V *in Fig.* 1)

A variable voltage of 500–2000 V. at a few micro-amp. is generally required, and although dry batteries can be used, a mains-driven power pack is less hazardous. Some form of stabilization is essential, since the response of a Geiger counter to a constant source of X-rays is not wholly independent of the voltage. Plateau slopes of 0·1%/V. are common, particularly for old counters and high counting rates, unless a quenching circuit is used. For the counter to be stable to within 1%, the power pack must be stable to within 0·5% or better for the maximum mains fluctuations. Commercially available circuits have more than adequate stability. (Proportional counters require an extremely high stability if they are to be used for wave-length discrimination.)

(b) *The Detector* (D *in Fig.* 1)

One simple form of detector used by Trost[13] is a galvanometer placed in series with the counter so as to measure the mean current flowing. The current flowing is discontinuous, owing partly to pulse produced by each photon and partly to the random arrival of photons; thus the galvanometer must have a long period in order to give a reasonably steady reading and must be sensitive, since even high counting rates produce a total current of only about one micro-amp. Such a detector is not normally used, although it would certainly suffice for many qualitative applications, and it is generally replaced by a counting-rate meter employing some A.C. amplification and with provision for a variable "period" (or integrating time constant) and a variable sensitivity. Many excellent circuits are commercially available, a number of them including a stabilized power pack, and the details need not be described here. An alternative form of detector is one where each voltage pulse is counted. Electromagnetically operated veeder counters, such as a post office register, are suitable for slow counting rates of less than 300/min. (High-speed mechanical counters are capable of speeds of 5000/min.) However, for very high counting rates it is usual to scale down the pulses electronically by a factor of 100 or more before supplying the mechanical register. Such a scaling unit is capable of counting at speeds of up to 10^6 counts/sec.

The choice between the two forms of detector is not always an easy one. The counting-rate meter can easily be made to produce a record of changing intensities by coupling it to a recorder, such as is used for recording temperature fluctuations. This is a great asset, since it enables many types of scanning to be done automatically. The use of direct counting, on the other hand, is inconvenient for use with changing intensities, although devices have been produced to give automatic recording (Lang[14]). The use of a scaling circuit is usually much slower than the use of a counting-rate

---

* Krypton is often used to obtain a high efficiency for detecting Mo$K\alpha$ radiation.

† The effective dead time for counters with no external quenching is a function of the detecting-circuit sensitivity, and this is rarely stable and frequently depends on the rate of count. Proportional counters have negligible dead-time losses up to rates of 10^6 counts/sec., and this is one of their principal advantages.

meter, but a considerably greater accuracy is possible. In order to understand the reason for this greater accuracy, it is necessary to discuss the statistics of counting and the basic functioning of a counting-rate meter.

The Statistics of Counting

Photons do not arrive at regular intervals, but in a random fashion. Thus if N counts per unit time are observed, this will rarely coincide with the average counting rate which would be obtained by counting for an infinite time. In fact, the probable error on such a count will be $0{\cdot}67\sqrt{N}$ counts, a percentage error of $67/\sqrt{N}\%$.* Thus by counting for four times as long the probable percentage error is halved.

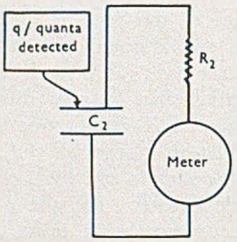


FIG. 4.—The Basic Circuit of a Counting-Rate Meter. Under steady-state conditions the meter reads a current N_q where N is the mean number of counts/sec. If the detection then stops, the meter falls exponentially towards zero with a time constant R_2C_2 as the condenser discharges through R_2, and in this sense the circuit is able to remember previous events.

The statistical fluctuations of the reading of a counting-rate meter are more complex. Fundamentally the rate meter functions as in Fig. 4, a charge of q/photon being fed into the condenser C_2, tending to raise the voltage across the condenser; this charge leaks away exponentially through the resistance R_2, and the mean current is indicated by the milliammeter. The product RC (R in megohms and C in microfarads) is a measure of the "memory" of the circuit in seconds, and Schiff and Evans[15] have shown that statistically the counting-rate meter functions as if it counted for a time $2RC$. Thus, for a counting rate of M counts/sec. the meter effectively registers $2MRC$ counts and will fluctuate about the mean reading with a probable error of $67/\sqrt{2MRC}\%$. For a given rate the statistical fluctuations can be made as small as necessary by increasing the value of RC (most commercial units provide a wide range of possible RC values). The use of a large time constant is thus desirable to obtain accurate and steady readings, but in practice an upper limit exists because the rate meter takes $5RC$ sec. to reach a reading within 1% of the true one, and for this reason a scaling circuit can give a more reliable indication of any change in a shorter time. In most applications time constants much greater than 5 sec. are impracticable. Some of the errors introduced by the use of a large time constant will be discussed later.

Applications of Geiger Counters

(a) *Debye–Scherrer–Hull Spectrum*

This method, which necessitates a polycrystalline specimen, is used extensively in metallurgical research. One common form of geometry used consists of a cylindrical wire specimen, surrounded by a concentric film on which the diffraction pattern shows as a series of "lines". Most spectrometers in essence retain this geometry, a Geiger counter replacing the film. Since a narrow slit is placed in front of the counter, it can detect only radiation diffracted over a very narrow angular region and, if the entire diffraction pattern is to be obtained, provision must be made to rotate the counter into all angular positions. Most spectrometers† use focusing geometry in order to increase the intensity of the pattern, and thereby reduce the statistical scatter, and the usual form of this is shown in Fig. 5. The source of diverging X-rays may be either the focus of the X-ray set, or, for more accurate work, the focus of a bent-crystal monochromator as shown.

For general use the spectrometer is usually equipped with facilities for rotating the counter about the spectrometer axis at a constant angular speed, and the counter is used in conjunction with a rate meter and an automatic recorder; the resulting chart gives a plot of the intensity against Bragg angle. An accuracy similar to that possible with films is obtained, but in most cases the required results can be obtained much more quickly. In cases where high accuracy is essential a scaling unit is used, the counter being set at specific angular positions and the intensity recorded for each position. Such a procedure is tedious and slow, but can yield results of a much higher accuracy than is possible by any other method. In all applications, using either a rate meter or direct counting, where high accuracy is required, the X-ray-tube output must be kept constant, or the variations must be measured and a corrector applied. Hand control of the X-ray tube is very effective, but frequently inconvenient; Harris[22] and Lang[14] have standardized all readings against the tube current, and details of many electronic stabilizers have been published, but the most satisfactory method is direct monitoring of the X-ray output in which the ratio of the reflected intensity to the incident intensity is recorded (Hall, Arndt, and Smith[18]; Hall and Williamson[23]).

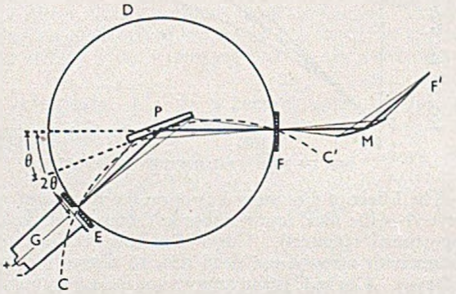


FIG. 5.—The Geometry of a Focusing Spectrometer. A narrow source of X-rays at F, consisting of either the line focus of an X-ray tube or the focus of a curved crystal monochromator M, as shown, diverges on to a specimen P at the centre of the spectrometer. The Geiger counter G can be rotated to any angle 2θ about this centre and detects rays passing through the slit E on the circle DEF. The chord EF subtends an angle $180-2\theta^\circ$ on all points on the circle CPC' and would thus give perfect focusing for a specimen conforming to this circle. The use of a flat tangential specimen such as P, gives negligible aberration if the divergence of the beam is less than 5°.

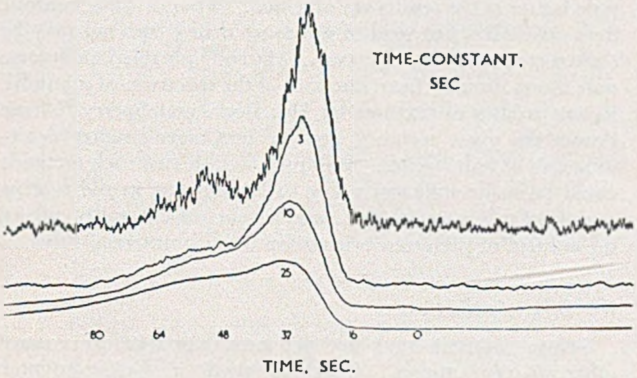
---

* It is important to note that this is not the maximum error. About 1 reading in 20 will have an error 3 times this, 1 in 200 an error 4 times, and 1 in 1200 an error 5 times, and thus a convenient measure of the maximum error is 4 times the probable error. The comparison of calculated and observed errors is a good method of checking the reliability of the apparatus.

† The first spectrometer (Le Galley[16]) used a cylindrical specimen. Later models (Friedman[17]; Hall, Arndt, and Smith[18]) have used focusing of the type discussed by Bragg,[19] Brentano,[20] and Brindley and Spiers.[21]

(b) *Qualitative and Semi-Quantitative Chemical Analysis*

Spectrometers using a counting-rate meter and automatic recording are excellent for applications in chemical analysis, since they may be made to give a direct record of intensity versus θ. Up to 1000 counts/sec. (Fig. 3) the response is approximately linear, and thus the approximate composition of the sample may be determined from the peak intensity. The results may be used with an X-ray-diffraction data index in the usual way, starting with the stronger lines and eliminating possible components with the weaker lines. The results are such an improvement on those obtained with films, particularly from the point of view of intensities,* that future data cards will probably all be compiled from Geiger-counter studies.



[Courtesy Institute of Physics

Fig. 6.—The Effect of Time Constants on X-Ray Lines (*Wainwright*[21]). An increase in the time constant reduces the statistical scatter, but results in a displacement of the apparent peak positions and, if used with too fast scanning rates, may smooth out the weaker lines.

The counter method, since it gives an almost instantaneous indication of intensity at a particular setting, offers many short cuts for high-speed analysis. This is frequently desirable, since a complete scan of the spectrum over the range $\theta = 10°–80°$ takes about the same time as the exposure of a fast film. If the possible constituents are known, or if some of the possibilities have been determined by a short, fast scan from 10° to 20°, analysis is very much quicker, as a system of searching by hand-scanning over a range of 1° about the expected Bragg angles can be used to determine which of the possible constituents is present. Such a "search" technique has been employed by Wainwright[24] to determine whether a series of samples were nickel- or chromium-plated. Wainwright also recommends that the initial scan for the strong lines should be carried out by hand in the interests of speed. Semi-quantitative analysis can be carried out simply, if the constituent phases are known but the relative amounts are not known, by calibrating the spectrometer with a series of known samples and graphing the values of the peak intensities against composition. An accuracy of $\pm 5\%$ in the measurement of intensities is good when a rate meter is used; some early claims to much greater accuracy have been based on the peak counting rate, and have failed to take into account the loss of accuracy due to background level,† dead-time losses (Carl,[25] Lonsdale[33]), circuit and incident X-ray intensity changes (Hall and Williamson[23] found a drift of $\pm 2\%$ even when using monitoring).

Detection of low-intensity lines due to impurities is difficult using a counting-rate meter, since a small time constant is necessary if the scan is to be fast, but this leads to a large statistical scatter which may mask small lines due to impurities. Wainwright[24] has investigated certain of the effects of time constants, and some of his results are shown in Fig. 6. With a small time constant (top curve) two peaks are evident, but the range of the statistical fluctuations is of the same order as the maximum value of the small peak. Increasing the time constant results in smaller fluctuations, but unless the speed of scanning is reduced it may involve loss of the weaker lines owing to the excessive smoothing action, as shown in the bottom curve of Fig. 6. Choice of correct time constant is very important, and in general it is preferable to err on the lower values. Wainwright recommends that fluctuations from the mean greater than 4 times the probable error should be re-investigated. Such re-investigation should preferably be carried out manually with a scaling unit, and this can be done immediately without disturbing the rest of the equipment. The potentialities of the method are very high, as is indicated in Fig. 7, which shows the diffraction pattern of tungsten powder annealed with a small quantity of steel filings.[27] The quantity of iron present is less than 2%, but a large number of lines are readily discernible which are due to the complex compounds formed. The extreme sensitivity of the method may be judged from the fact that the peak intensities of the main lines are about 15,000 counts, but it is unlikely that such sensitivity would be possible without the use of a monochromator.

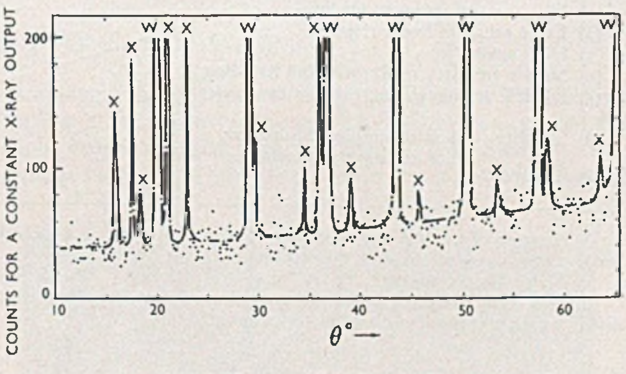


Fig. 7.—A Plot of the Background Level in the Spectrum of Tungsten Containing Less than 2% Impurity. The weak lines *X* are due to the impurity; those marked *W* are due to the tungsten and have peak intensities of about 10,000 counts. These results were obtained using monochromatic radiation, a scaling unit and continuous monitoring.

---

* This improvement may be a source of embarrassment, since lines which are certainly the strongest have sometimes been classed as second- or third-strongest on the cards when films have been used. This danger is particularly great if the specimens for the data index were enclosed in thin tubes.

† The background level has a large effect since the squares of the errors add (Strong[26]). Thus the apparent probable percentage error in R readings of the peak intensity M superimposed on a background intensity m is $67/\sqrt{R(M+m)}$. The probable percentage error in the peak intensity reading however is $67\sqrt{1+\{(1+R/P)(m/M)\}}/\sqrt{RM}$, where P readings of the background intensity are taken. Thus if $M = m$, $R = 10$, and $P = 1$, the true error is 5 times greater than the apparent error.

(c) *Quantitative Analysis*

Analysis corresponding to an accuracy in the measurement of integrated intensities of better than ±2% is possible although extremely tedious (Hall and Williamson[23]). Accuracies in the determination of composition corresponding to ±5% in intensity for strong lines and ±10% for weaker lines, using a counting-rate meter, can be achieved only if stringent precautions are taken at every stage (see, e.g., Lonsdale[33]). Some of the early claims were over-optimistic but most workers now appear to obtain accuracies equivalent to those suggested above. The technique is particularly suited to ceramics and minerals (e.g. Tuttle and Cook[34]; McCrery[35]) and it has been applied to the determination of metal carbides by Redmond.[36] Heal and Mykura[37] have used a high-temperature camera to follow the transformation of austenite to martensite.

(d) *Parameter Studies*

The record of intensity against a calibrated scale of θ would make the Geiger-counter spectrometer seem ideal for accurate parameter studies, and, because of the high speed with which line position can be measured by a "search" technique, it appears to be a very valuable tool for rapid phase-diagram determination. Unfortunately, two sources of error exist, one due to the counting-rate meter, and the other to the specimen. Wainwright[24] has demonstrated very convincingly (Fig. 6) how the use of a counting-rate meter shifts the lines owing to its memory effect. The peak of the line always coincides with a point of inflection, but this is not very satisfactory for accurate determinations. Scanning in both directions (increasing and decreasing θ) and taking the mean value, appears to be a better solution than very slow scanning speeds. The second difficulty is due to the fact that measurements of line position are not readily possible on both sides of the beam, and thus the surface of the specimen must be positioned very accurately, and the spectrometer zero carefully standardized. These errors have been discussed by Wilson,[28] but very careful work will be needed if the results are to be reliable, just as in the case of films.

Table I.—*Relative Merits of Methods of Detecting X-Rays.*

Ideal Features / Method	Films	Ion Chamber	Geiger Counter: Counting-Rate Meter	Geiger Counter: Scaling Unit	Proportional Counter
(1) Directness of intensity reading	Bad	Good	Good	Good	Good
(2) Large range of linear response	Fair	Good	Fair	Good *	V. good *
(3) High sensitivity	Bad	Bad	Fair	V. good	†
(4) Simple auxiliary equipment and handling	Fair–good	Fair	Fair	Bad	V. bad ‡
(5) Possible accuracy of intensity measurements	±5%	±5%	±5%	±1%	±1%
	Bad–fair	Fair–good	Fair		
(6) Stability: (*a*) with simplest equipment	Good	Good	Fair	Fair	Bad
(*b*) with best equipment	Good	Good	Good	V. good	Good
(7) Speed	Fair	Bad	Good	Bad	†

* A range of 0–1000 counts/sec. for Geiger counters, if counting-loss corrections are applied; 0 to 10^5 counts/sec. is possible with the proportional counters without correction.

† As for Geiger counter.

‡ The complexity of the auxiliary equipment necessary before a proportional counter can give even qualitative results is its principal disadvantage compared with a Geiger counter.

(e) *Preferred Orientations*

The accuracy with which a counting-rate meter indicates intensity is adequate for studies of preferred orientation, and it is in this field that Geiger-counter spectrometers have already proved their superiority over film methods. Geometrics which avoid the necessity for an absorption correction over most of or all the pole figure, have been devised by Norton,[29] Schulz,[30] and Chernock and Beck.[31] The ease with which the intensity, representing the density of the poles, may be read at a given setting allows the construction of a pole figure as the results are obtained. Two or three contour lines on a fairly fine grid of not more than 5° spacing may be constructed in an hour or two. Harris[22] obtained an inverse pole figure directly from one scan of the spectrum of uranium. Recent studies of textures by Hu, Beck, and Sperry[32] have shown the great accuracy possible in Geiger-counter determination of pole figures. It is quite feasible that such methods could be made sufficiently fast to be of value in the routine testing of the intermediate products in rolling mills, where the control of preferred orientation is of commercial value.

(f) *Other Applications*

Geiger counters have not yet been employed for many other X-ray studies. Cauchois used a Geiger-counter spectrograph for analysing the fluorescent X-radiation from slags, and was able to detect 10^{-3}% zinc with an accuracy of 10^{-4}%. Many workers have used them for lining-up fine collimators. Lonsdale[38] and Wooster, Ramachandran, and Lang[39] have made use of them for studying the diffuse reflections from single crystals, although in many cases Geiger counters are unsuitable because of the extremely high intensities frequently obtained in single-crystal cameras. Proportional counters in which dead-time losses are negligible and with a linear response up to intensities of 10^7 counts/sec. are more likely to be successful in this field.

Summary

Table I attempts to summarize the good and bad points of the common methods of detecting and measuring X-ray intensities. It should be emphasized that none of the methods, except that employing films, is in general suitable for two-dimensional studies, as in the Laue method, unless some preparatory exploration has been done.

REFERENCES

1. S. A. Korff, " Electron and Nuclear Counters ". **1946**: New York (D. Van Nostrand).
2. D. H. Wilkinson, " Ionization Chambers and Counters ". **1950**: Cambridge (University Press).
3. J. Sharpe and D. Taylor, *Proc. Inst. Elect. Eng.*, 1951, **98**, (II), 209.
4. U. W. Arndt and D. P. Riley, *Proc. Phys. Soc.*, 1952, [A], **65**, 74.
5. A. R. Lang, *Nature*, 1951, **168**, 907; *Proc. Phys. Soc.*, 1952, [A], **65**, 372.
6. G. C. Hanna, D. H. W. Kirkwood, and B. Pontecorvo, *Phys. Rev.*, 1949, [ii], **75**, 985.
7. S. C. Curran, J. Angus, and A. L. Cockcroft, *Nature*, 1948, **162**, 302.
8. E. H. Cooke-Yarborough, C. D. Florida, and C. N. Davey, *J. Sci. Instruments*, 1949, **26**, 124.
9. L. G. Parratt and C. F. Hempstead, *J. Appl. Physics*, 1951, **22**, 1502.
10. L. T. Schiff, *Phys. Rev.*, 1936, [ii], **50**, 88.
11. M. Blackman and J. M. Michiels, *Proc. Phys. Soc.*, 1948, **60**, 549.
12. W. H. Hall and G. K. Williamson, *J. Sci. Instruments*, 1952, **29**, 132.
13. A. Trost, *Z. Physik*, 1941, **117**, 257; *Metallwirtschaft*, 1939, **18**, 687.
14. A. R. Lang, *J. Sci. Instruments*, 1952, **29**, 176.
15. L. I. Schiff and R. D. Evans, *Rev. Sci. Instruments*, 1936, **7**, 456.
16. D. P. Le Galley, *ibid.*, 1935, **6**, 279.
17. H. Friedman, *Electronics*, 1945, **18**, 132; see also J. Bleeksma, G. Kloos, and H. J. Di Giovanni, *Philips Tech. Rev.*, 1948, **10**, 1.
18. W. H. Hall, U. W. Arndt, and R. A. Smith, *Proc. Phys. Soc.* 1949, [A], **62**, 631.
19. W. H. Bragg, *Proc. Phys. Soc.*, 1921, **33**, 222.
20. J. C. M. Brentano, *ibid.*, 1924–25, **37**, 184.
21. G. W. Brindley and F. W. Spiers, *ibid.*, 1934, **46**, 841.
22. G. B. Harris, *Phil. Mag.*, 1952, [vii], **43**, 113.
23. W. H. Hall and G. K. Williamson, *Proc. Phys. Soc.*, 1951, [B], **64**, 946.
24. C. Wainwright, *Brit. J. Appl. Physics*, 1951, **2**, 157.
25. H. F. Carl, *Amer. Mineralogist*, 1947, **32**, 508.
26. J. Strong *et al.*, " Procedures in Experimental Physics ". **1939**: New York (Prentice Hall).
27. G. K. Williamson, unpublished work.
28. A. J. C. Wilson, *J. Sci. Instruments*, 1950, **27**, 321.
29. J. T. Norton, *J. Appl. Physics*, 1948, **19**, 1176.
30. L. G. Schulz, *ibid.*, 1949, **20**, 1030, 1033.
31. W. P. Chernock and P. A. Beck, *ibid.*, 1952, **23**, 341.
32. H. Hu, P. A. Beck, and P. R. Sperry, *Trans. Amer. Inst. Min. Met. Eng.*, 1952, **194**, 76.
    P. A. Beck and H. Hu, *ibid.*, 83.
33. K. Lonsdale, *Amer. Mineralogist*, 1948, **23**, 90.
34. M. A. Tuttle and R. L. Cook, *J. Amer. Ceram. Soc.*, 1949, **32**, 279.
35. G. L. McCreery, *ibid.*, 1949, **32**, 141.
36. J. C. Redmond, *Anal. Chem.*, 1947, **19**, 773.
37. H. T. Heal and H. Mykura, *Metal Treatment*, 1950, **17**, 129.
38. K. Lonsdale, *Acta Cryst.*, 1948, **1**, 12.
39. W. A. Wooster, G. N. Ramachandran, and A. R. Lang, *J. Sci. Instruments*, 1948, **25**, 405.

POWDER METALLURGY IN METALLURGICAL RESEARCH

By W. R. PITKIN * and D. J. JONES,* B.Sc., A.I.M.

Introduction

It is not proposed to discuss the merits or demerits of powder metallurgy as compared with the more orthodox method of melting and casting, but to present a number of examples in which, from necessity or desirability, use has been made of this technique in studying the properties of certain metals and alloys.

Research metallurgists in the lamp and valve industries have accepted, in some cases without choice, the " powder " approach, and have employed this method over the past thirty years as a means of studying, in particular, the structural characteristics of the high-melting-point metals tungsten and molybdenum. The experience gained in this field, with particular reference to the development of the tungsten lamp filament, provides a valuable basis and approach to a consideration of the use of this technique in other fields. It is proposed, therefore, to discuss the subject under the three following headings: (1) tungsten for lamp filaments; (2) alloys for valve filaments; and (3) alloys having specific properties.

Tungsten for Lamp Filaments

To produce a lamp of high efficiency it is necessary that the filament should be substantially free from movement throughout its useful life. For example, Fig. 1 (*a*) and (*b*) show two tungsten filaments which have been burnt in lamps for over 1000 hr. at *c.* 2400° C., that is, at a temperature equal to about 70% of the melting temperature. In Fig. 1 (*a*), the filament has remained substantially unchanged, the light output dropping only slightly over the period of test. The filament in Fig. 1 (*b*), however, shows substantial movement and the light output dropped more markedly. This instability may appear in a general way, as shown in Fig. 1 (*b*), or in a more

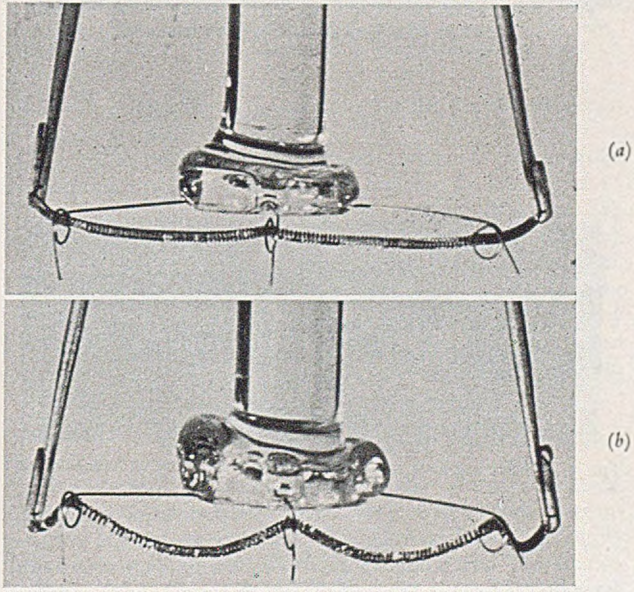


Fig. 1.—Stable (*a*) and Unstable (*b*) Tungsten Lamp Filaments.

---

* Research Laboratories of The General Electric Co., Ltd., Wembley, Middlesex.

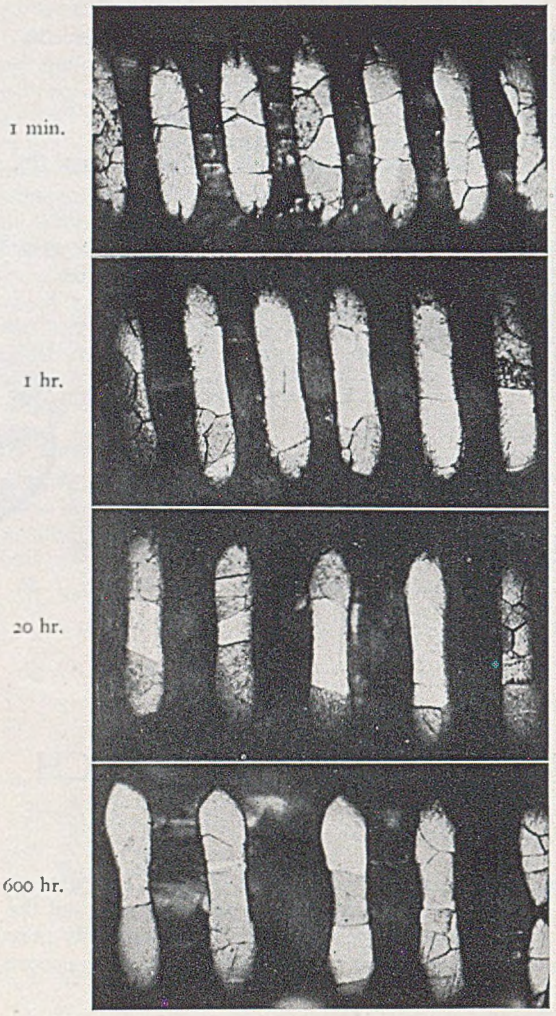


FIG. 2.—Effect of Heat-Treatment at *c.* 2400° C. on the Crystal Structure of a Very Pure Tungsten Filament.

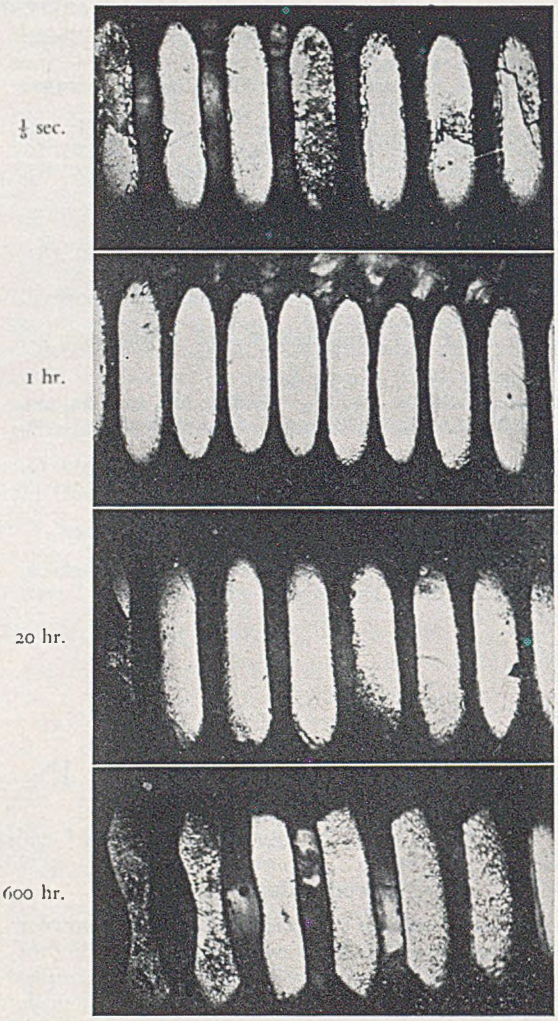


FIG. 4.—Stable Large-Grained Structure in Tungsten Containing Trace Additions of Alkali and Silica, After Heating at *c.* 2400° C. for Various Times.

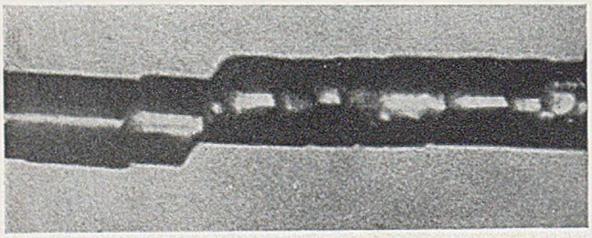


FIG. 3.—Offsetting Due to Grain-Boundary Movement in a Tungsten Filament.

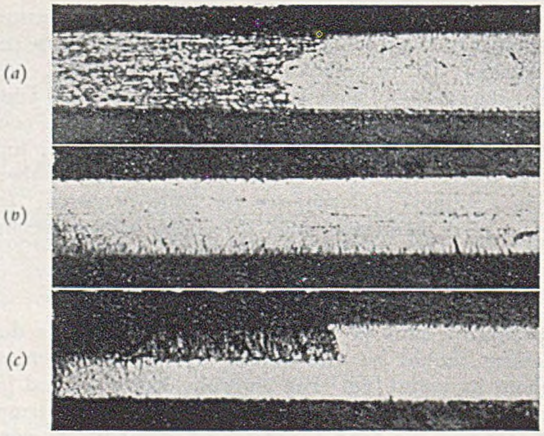


FIG. 5.—Types of Grain Boundary Obtained in Tungsten Wire by Controlled Additions. (*a*) Type at right angles to wire axis; (*b*) complete single crystals; (*c*) spliced or overlapping type.

specific manner such as localized distortion of turns or grain-boundary movement.

As a result of a study of the structure of filaments throughout their burning life, it became evident that instability was, in the main, associated with recrystallization and grain growth. For example, Fig. 2 illustrates the effect of heat-treatment at *c.* 2400° C. on the structure of a very pure tungsten filament. It is of interest to note the crystal size after 1 min. and the slight but progressive growth of these crystals throughout the time of 600 hr. A lamp filament with this type of structure would be unstable owing to the progressive grain growth throughout its life. Furthermore, apart from this effect, trouble may arise owing to grain-boundary movement giving rise to offsetting, as shown in Fig. 3.

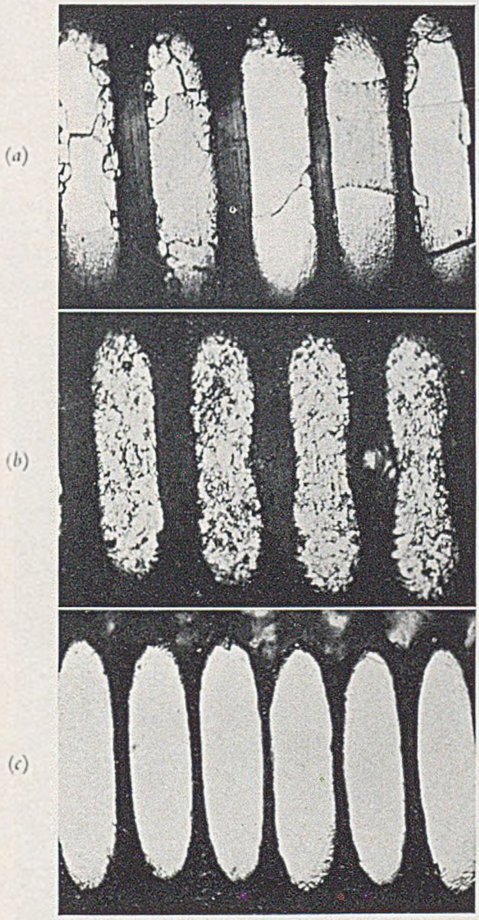


Fig. 6.—(*a*) Unstable Medium Grain-Size; (*b*) Stable Fine Grain-Size; (*c*) Stable Large Grain-Size in Tungsten Filaments, After Burning for 1000 hr. at *c.* 2400° C.

Thus, the problem resolved itself into the control of recrystallization and grain growth. The method of approach was two-fold: (i) production of a fine-grained structure in which further growth was inhibited; and (ii) production of a large-grained structure which remained stable from initial lighting of a filament to the end of its life.

These types of controlled structure are obtained by additions of refractory oxides. To produce a stable fine-grained structure it is usual to add approximately 0·5% thoria; on the other hand exaggerated growth is obtained by the addition of traces of, for example, alkali and silica. Fig. 4 shows the stable exaggerated grain of tungsten containing traces of alkali and silica after heating for various times at *c.* 2400° C. Of particular interest is the almost instantaneous completion of grain growth on first heating to this temperature. This results in a very stable filament.

Reference has been made above to offsetting, which is associated with the existence of crystal boundaries running across a section of a wire, as seen in Fig. 2. To a certain extent, by suitably modifying the additions, some degree of control over the type of boundary may be maintained. For example, in Fig. 5 varying types of boundaries are shown. Of particular interest is the spliced or overlapping type; movement across or along this type of boundary is extremely difficult.

Thus, methods are available, using the powder technique, whereby the structure can be readily controlled. This technique is most convenient, as the addition of refractory oxides, either in the form of a suspension or as a solution of a suitable salt added to the oxide or to the metal powder, is relatively simple.

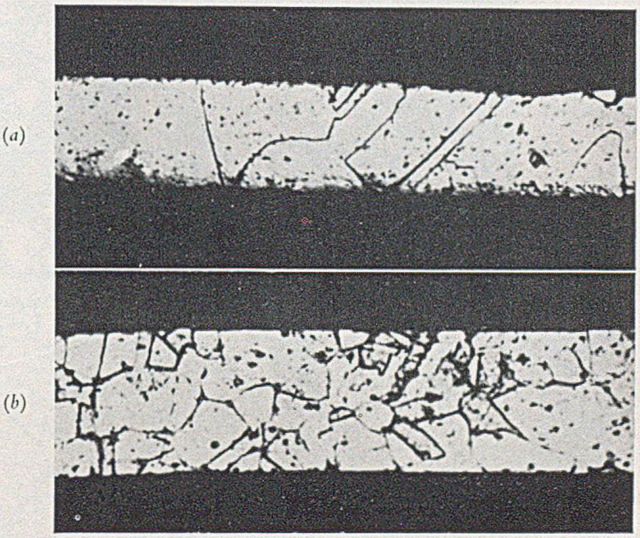


Fig. 7.—Structure of (*a*) Ni–0·25% Al Alloy and (*b*) Ni–0·4% Al–0·2% Al_2O_3 Alloy, Annealed in Dry Hydrogen at 1000° C.

Fig. 6 shows the three typical structures of tungsten filaments referred to, after burning for 1000 hr. at *c.* 2400° C. These are, respectively, the unstable medium grain-size (*a*), the stable fine grain-size (*b*), and the stable large grain-size (*c*).

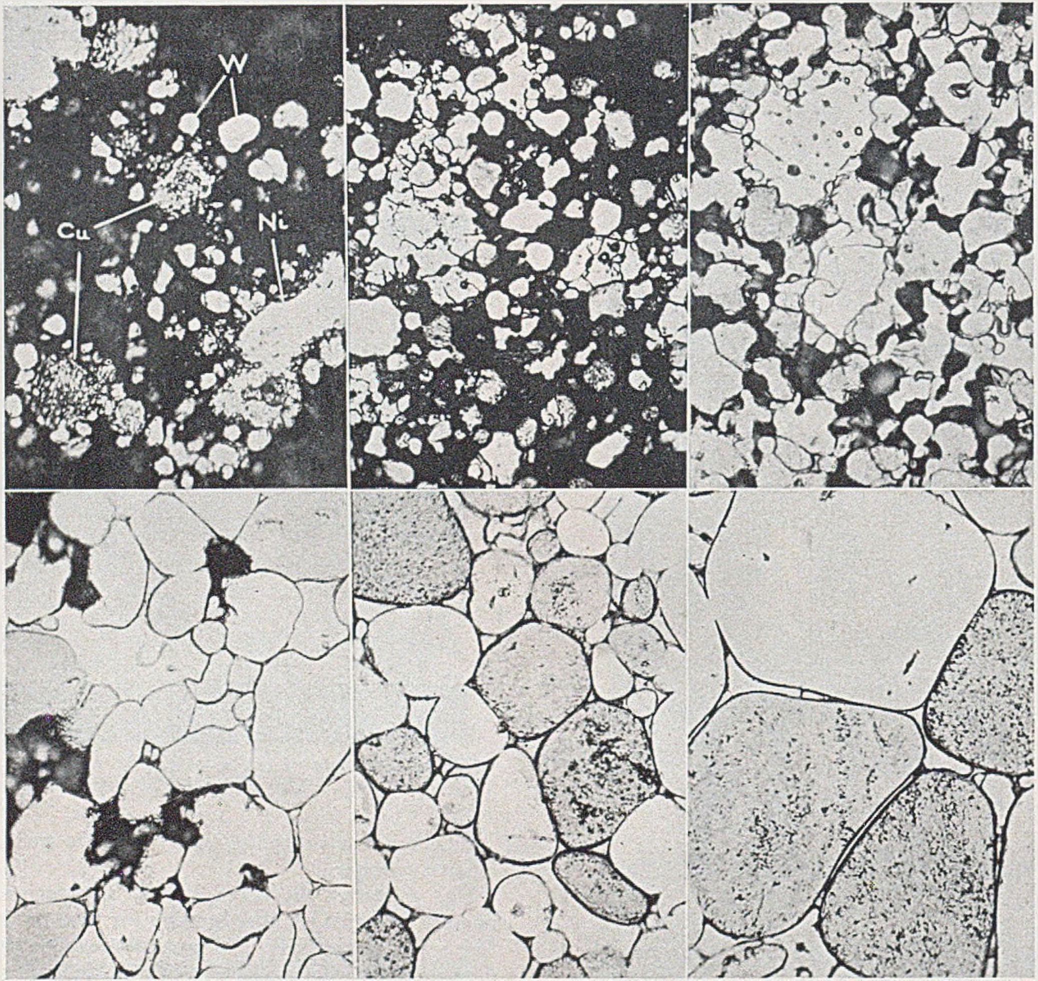
It is worth noting that, apart from the thoria addition, differences in purity of the basic tungsten cannot as yet be detected. This illustrates the important role played by minute impurities, suggesting that the powder technique can usefully be applied to a study of the effect of trace additions to other metals and alloys.

Alloys for Valve Filaments

This is a problem very similar to that already discussed, in which it is necessary to ensure very close control of composition and structural stability. Both these factors are highly important with respect to the thermionic and life characteristics of the valve. As the filaments in a valve are operated at elevated temperatures under a constant load, it is essential that they possess a relatively high resistance to creep.

In this field the powder technique has proved of real value in studying the effect of metallic and non-metallic additions. For example, Fig. 7 shows the effect on the structure of an addition of alumina to a nickel–aluminium alloy. In this particular case, the aluminium in solution in the nickel raises the recrystallization temperature, and the presence of alumina restrains the grain growth, thus preventing offsetting.

In the preparation of alloys of this type, it is undesirable to mix aluminium powder with the nickel powder owing to the readiness with which aluminium oxidizes. The procedure adopted to overcome this difficulty consists of preparing a brittle nickel–aluminium alloy of approximately 50:50 composition by melting the metals in a purified hydrogen atmosphere and allowing the melt to freeze in the crucible. It is crushed and finally ball-milled to a fine powder. This fine powder is then mixed with the nickel powder in the required proportions. The method is particularly useful for all metals that oxidize readily.



1000° C. 1300° C. 1350° C.

1400° C. 1450° C. 1450° C. (prolonged heating)

Fig. 8.—Effect of Temperature on the Progress of Sintering in a W–Ni–Cu Alloy. × 300.

Alloys Having Specific Properties

In this section three typical examples are discussed to illustrate the usefulness of the powder technique.

(a) *Investigation of a combination of metals of limited solubility and of widely differing melting points.*—The example chosen is a tungsten–nickel–copper alloy with a tungsten content of about 90%. Owing to the high melting point of tungsten, the powder method of approach proved invaluable. In this particular instance it was necessary to achieve a high density, coupled with specific mechanical properties. This entailed a study of part of the ternary system and of the effect of diffusion on the mechanical properties. Fig. 8 illustrates the effect of temperature on the progress of sintering in this system, the reactions consisting of the production of a nickel–copper liquid phase, solution of fine tungsten particles in this phase to saturation, then re-precipitation of tungsten on to existing tungsten grains.

(b) *Examination of the expansion characteristics of nickel–iron and similar alloys.*—In the electrical industry a range of nickel–iron alloys are used for sealing to glass, the tolerances on expansion being of the utmost importance. For some specialized applications it has been necessary to determine the expansion characteristics of very pure nickel–iron alloys. The powder technique provides a ready means of carrying out this type of investigation and in particular of obtaining information on the effect of metallic impurities, which can be controlled to very close limits. The results have shown that it is possible by this method to manufacture such alloys with very close tolerances on expansion. This is of particular importance in specialized types of glass-to-metal seals.

(c) *The study of the properties of metals that are completely immiscible.*—For many purposes a combination of metals completely immiscible in each other is of practical importance, e.g. copper-tungsten and silver-tungsten electrical contacts. The preparation of materials of this type is readily accomplished by the powder method, and the technique promises to be extremely useful for future work in this interesting field, particularly with regard to metallic and non-metallic combinations.

SOME METHODS OF MEASURING SURFACE TOPOGRAPHY AS APPLIED TO STRETCHER-STRAIN MARKINGS ON METAL SHEET

By W. H. L. HOOPER,* B.Sc., A.I.M., and J. HOLDEN,† Ph.D.

STRETCHER-strain markings can be very troublesome to users of sheet metal in pressing and other manipulative processes, but while light-reflection effects may be strong, giving the impression of deep surface distortions (Fig. 1) the differences in level are, in fact, quite small. Two techniques for measuring the surface topography of sheet metal showing strain markings are described below, the first using the Talysurf surface-measuring instrument, and the second, light-interference effects.

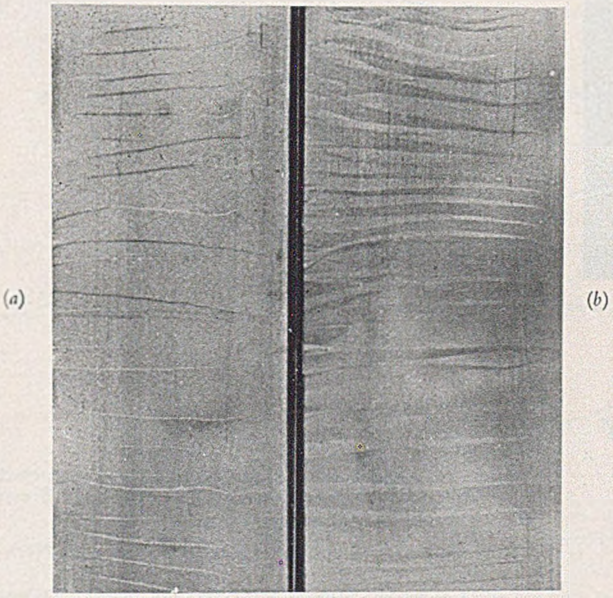


FIG. 1.—Stretcher-Strain Markings in (*a*) Steel and (*b*) Aluminium–Magnesium Alloy Pressings Made in the Same Die. × ¼.

The methods are illustrated by reference to parallel-sided strip specimens of dead-soft mild steel and aluminium–3% magnesium alloy, both of which develop strain markings of very similar superficial appearance when stretched on a tensile-testing machine. The first markings appear in either steel or aluminium alloy as soon as the yield point is reached, and others continue to appear during stretching up to an elongation of about 2%, beyond which strain markings are of relatively slight industrial significance.

Mechanical Measurements

Fig. 2 shows the set-up of the Talysurf exploring head in the examination of a stretched specimen which is securely bolted to a steel table. The head is lowered until the stylus, tipped with a diamond of 0·0001 in. radius, makes contact with the surface under examination, and the table and specimen are then driven at a uniform speed in the right-to-left direction. The rise and fall of the stylus as it travels over the specimen surface are magnified electronically, the amplification chosen in this instance being ×1000, and these magnified displacements are traced by a high-speed autographic recorder, the horizontal magnification in the present instance being ×2.

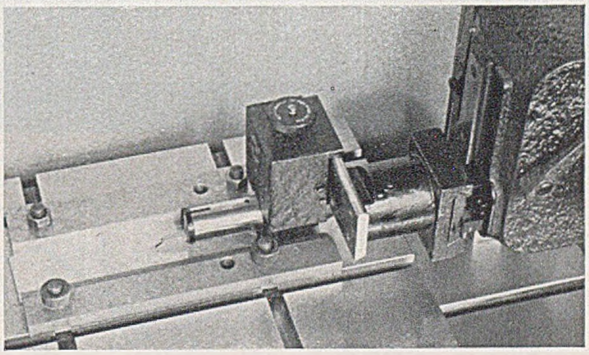


FIG. 2.—Exploring Head of Talysurf Instrument Arranged for the Measurement of Strain Markings.

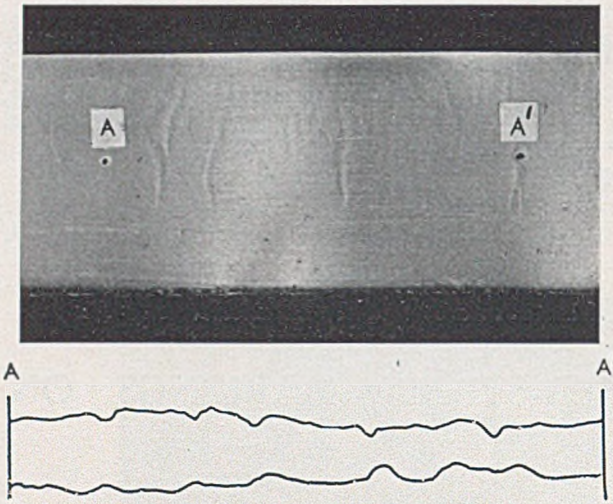
Traces are taken from both surfaces of the specimen between two fixed points, and a typical example of the appearance of a steel specimen stretched ½%, and the record taken at this stage of deformation, are illustrated in Fig. 3. Jevons[1] has described stretcher-strain markings in steel as consisting, in their most typical form, of coincident depressions in opposite

---

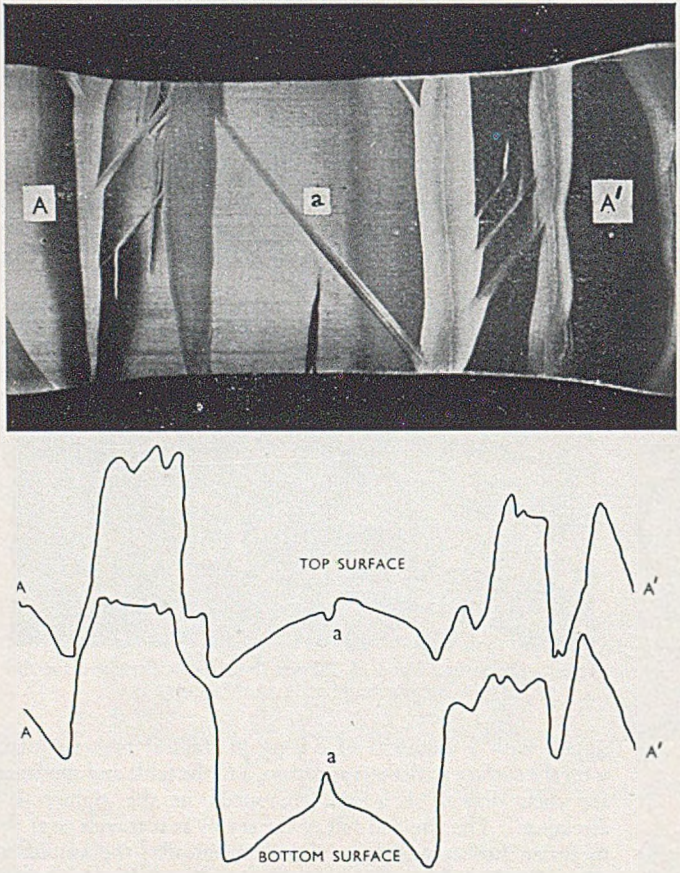
* Research Technical Officer, Imperial Chemical Industries, Ltd., Metals Division, Birmingham.

† I.C.I. Research Fellow, Royal Holloway College, London University.

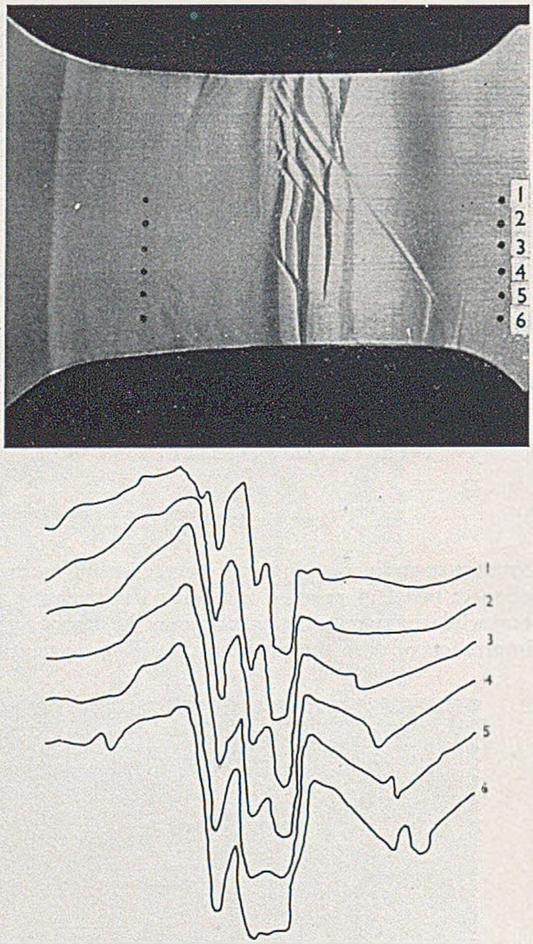
3



4



5



Figs. 3–5.—Stretcher-Strain Markings in Steel Stretched $\frac{1}{2}\%$ (× 1) and Corresponding Talysurf Traces (× 1000 vertical magnification).

Fig. 3.—Wedge-Shaped Markings Associated with Necking of the Sheet.

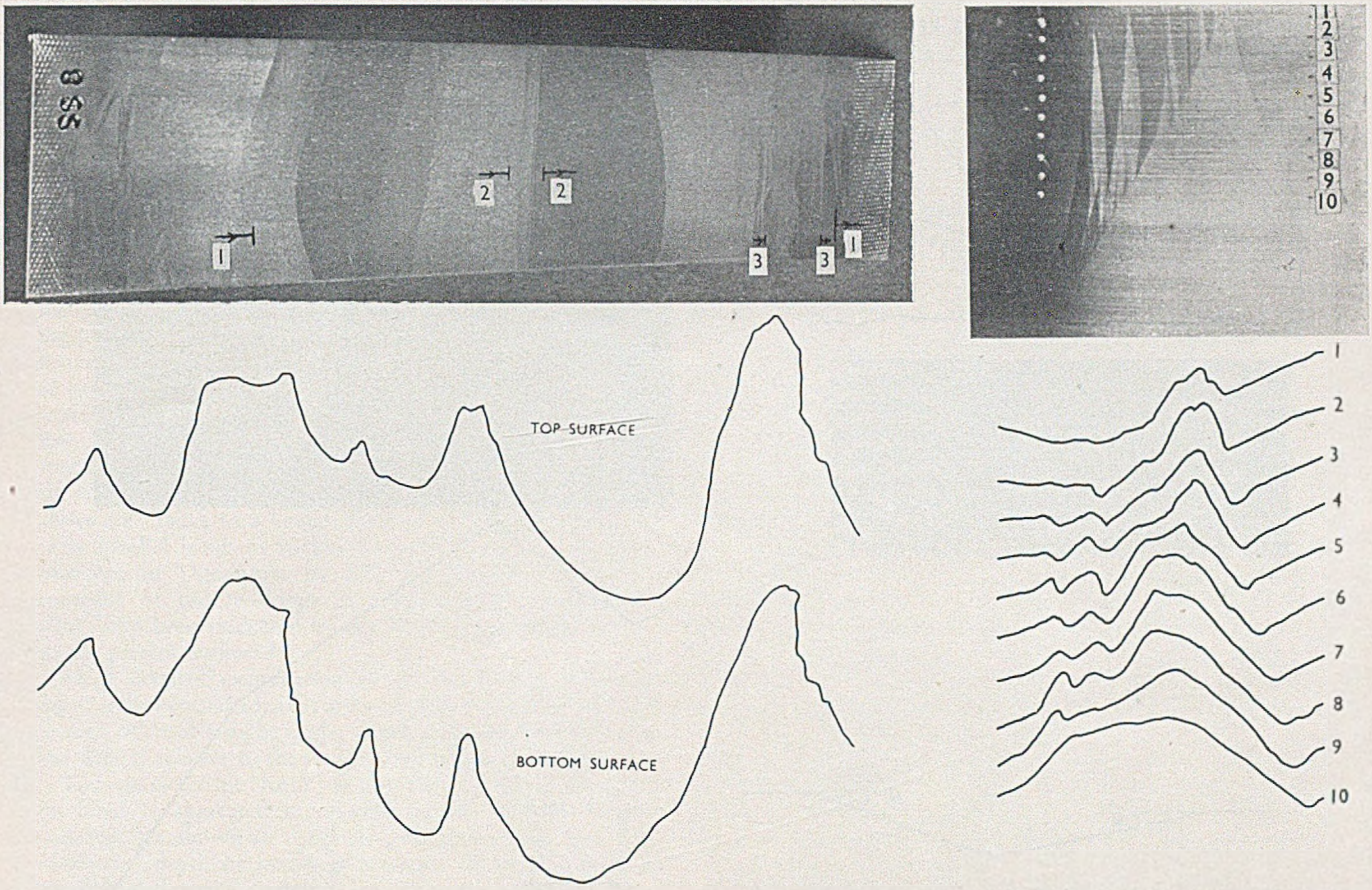
Fig. 4.—Wedge-Shaped Markings and Kinks, Present Together.

Fig. 5.—Talysurf Traces of a Small Area of One Surface.

sides of the sheet, so that a neck or wedge, associated with appreciable local thinning, is formed. The markings illustrated in Fig. 3 are of this type.

Stretched steel specimens may also display another type of marking, namely kinking, which is not associated with significant local thinning. In the specimen illustrated in Fig. 4, both kinks and necks are present, the kinks, in this instance, accounting for the most significant displacement and irregularity of the metal surfaces.

More detailed information on the physical form of stretcher-strains is obtained by making a series of parallel traces over a small area of the stretched surface. The steel specimen shown in Fig. 5 was stretched $\frac{1}{2}\%$, and exhibits markings in various stages of development; the slight roughening of the left-hand half of the gauge-length has been caused by the spread of one group of markings, while a freshly formed group is evident in the middle of the gauge-length. Talysurf traces made at 0·1-in. intervals are shown as curves 1–6 and indicate that, when first formed, the markings consist of kinks and U- or V-shaped notches. The notch walls recede



Figs. 6–7.—Stretcher-Strain Markings in Aluminium–3% Magnesium Alloy Stretched ½% (× 1) and Corresponding Talysurf Traces (× 1000 vertical magnification).

Fig. 6.—Kink-Type Markings. The traces are between 1 and 1 on the photograph.

Fig. 7.—Talysurf Traces of a Small Area of One Surface.

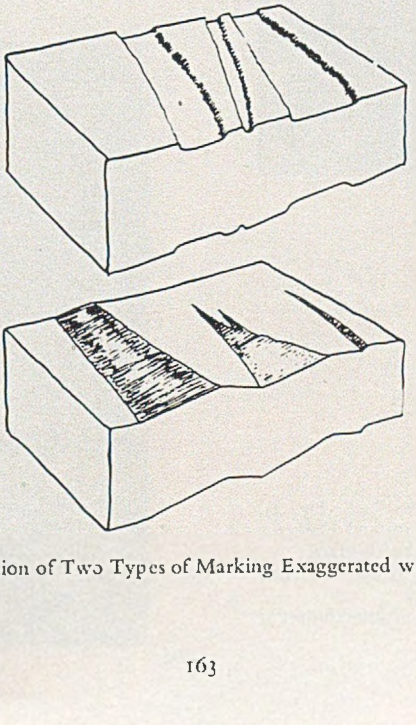
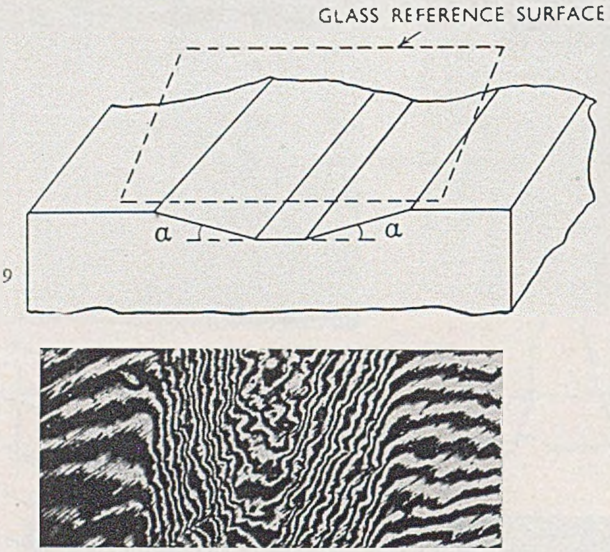
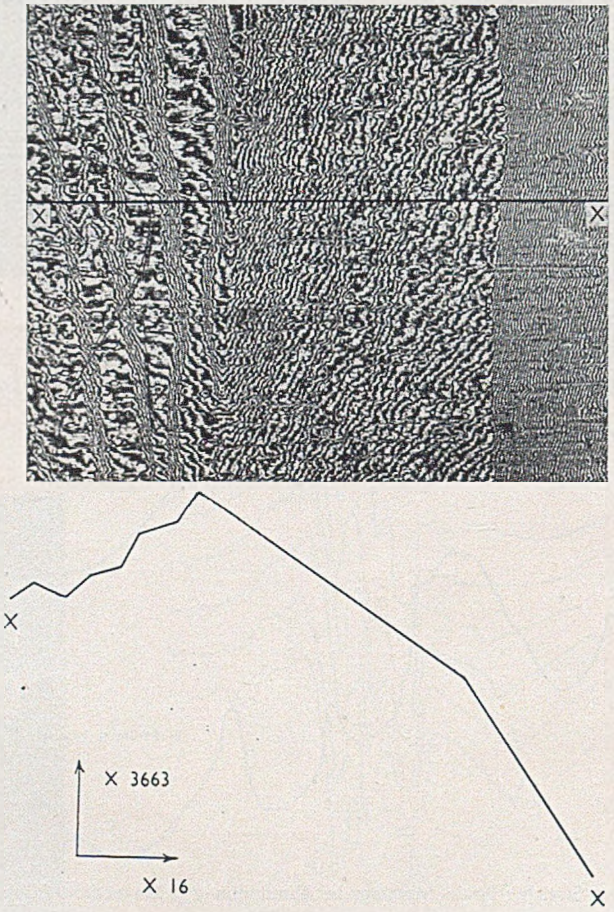


Fig. 8.—Diagrammatic Representation of Two Types of Marking Exaggerated with Respect to Sheet Thicknesses.

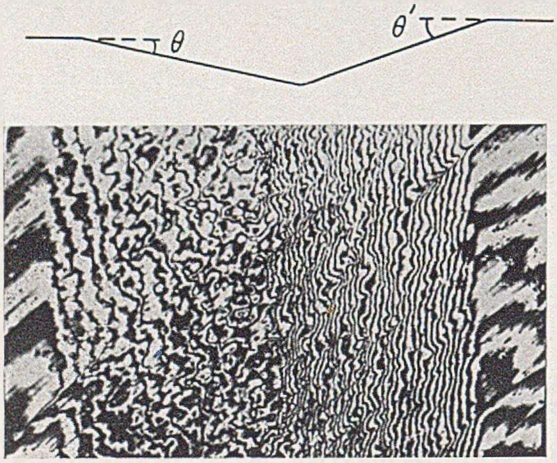
Interferometric Patterns of Strain Markings.



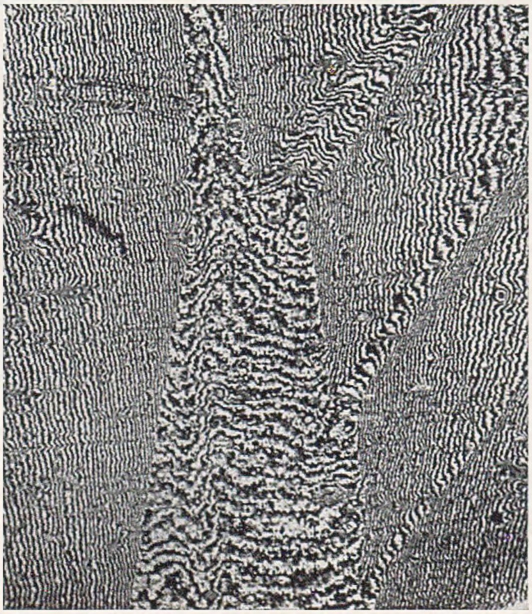
9



11



10



12

Figs. 9–10.—Channel-Marking in Steel (see "*a*", Fig. 4), with Sides at Equal Angles (Fig. 9, × 40) and Different Angles (Fig. 10, × 90) to the Undisturbed Surface.

Fig. 11.—Narrow Tongue-Like Markings (Kinks) in Aluminium–3% Magnesium Alloy. × 16.

Fig. 12.—Complex Forked Marking in Aluminium–3% Magnesium Alloy. × 16.

from each other as stretching progresses, so that flat-bottomed channels are formed. Between $\frac{1}{4}$ and 2% elongation the markings become more numerous but of diminished intensity. With more than about 2% extension they merge and the surface becomes comparatively smooth again.

Similar measurements have also been made of strain markings obtained by stretching annealed strips of aluminium–3% magnesium alloy sheet. The markings reach their greatest intensity when first formed, at the start of plastic deformation. The curves relating to the stretched aluminium specimen shown in Fig. 6 indicate that, at this stage, the markings consist of kinks, a peak on one surface being represented on the other by a valley, so that no thinning occurs. As stretching proceeds fresh groups form, and spread either by a smooth movement or by a more rapid shooting and branching, until at between 1 and 2% extension they merge, and thereafter rapidly diminish in intensity. No examples of local thinning have been observed in the examination of any aluminium specimens at small strains of the order of 0–2%, but it should be mentioned that at higher strains aluminium–magnesium alloys are subject to a further form of marking consisting of close parallel bands or necks, examples of which have been observed in Duralumin by Fell and others,[2–4] while the contours of the markings in aluminium–3% magnesium alloy have been measured by the Talysurf technique by one of the present authors.[5]

More detailed examination of the small area shown in Fig. 7 indicates that adjacent areas of plane surface are inclined to each other at a slight angle, which calculation shows to be less than 1′ relative to the undisturbed reference surface.

The characteristic form of the two types of observed markings, exaggerated in relation to the thickness of the material, are shown in perspective diagrams (Fig. 8). The maximum depth of markings measured in experiments on parallel-sided strips stretched on a tensile-testing machine was approximately the same in both steel and aluminium, i.e. about $\frac{3}{1000}$ in.

Interferometric Measurements

The principles of interferometry are well known and have been comprehensively described in the literature;[6, 7] consequently the discussion below is confined to a description of the technique employed in examining stretcher-strain markings.

The method consists, briefly, of covering the area of specimen under examination with a small piece of thin optical glass and illuminating the covered area by a parallel beam of monochromatic light, provided in this instance by a mercury-vapour lamp in conjunction with a filter which passes the mercury 5461 Å. green line. The specimen is viewed through a metallurgical microscope, modified by placing an extra iris diaphragm in the illumination arm; this does not involve any structural change to the microscope. For simple interferometric techniques the usual 16-mm. objective is satisfactory. Interference fringes are produced, and their changes in separation and direction indicate the topography of the surface in the field of view. Interpretation of the fringe patterns is comparatively simple, and depths and angles of the surface markings can be accurately determined.

Fig. 9 is a photograph of a typical interference pattern caused by the diagonal marking in the steel specimen indicated as (*a*) on the Talysurf traces in Fig. 4. For ease of interpretation, the interference fringes have been arranged to run across the band at right angles, and the sketch in Fig. 9 indicates how this is effected by suitable placing of the glass reference surface. The fringe pattern then clearly indicates that the band consists of a groove with a narrow, flat bed and flat sides inclined at equal small angles to the undisturbed surface. The angle of inclination, obtained from counts of the average number of fringes in a given distance on the sloping channel side and on the undisturbed surface, ranges between 20′ and 30′. Frequently the two sides of groove-like markings are inclined at different angles to the undisturbed surface, and this effect is illustrated by a typical photograph in Fig. 10.

An interference photograph of a group of narrow tongue-like markings in aluminium, similar to those shown in Fig. 7, is reproduced in Fig. 11. The exaggerated vertical section through the line *XX* in this figure, which was drawn from data calculated from the fringe spacing, is clearly similar to the Talysurf records shown in Fig. 7 and confirms the findings of the Talysurf method that, when first formed, strain markings in aluminium are flat, or slightly curved planes, orientated at a small angle to the undistorted surface.

Fig. 12 shows the more complex form of strain markings developed in aluminium at a later stage of stretching, by the shooting and branching of the original markings. It is clear however, that the characteristic flat inclined-plane formation of the markings is unchanged.

While confirming the measurements made upon stretcher-strain markings with the Talysurf instrument, the interference fringes, by illustrating the topography of an area instead of a line section, display the manner of termination and interconnection of individual bands; the complex branching deformations of "random markings" can also be studied in detail (see Fig. 12). In addition to the deformation bands themselves, it is seen that the fringes also illustrate the local surface undulations associated with the electropolishing of the surface, and the parallel scratches produced by the rolling of the specimens.

REFERENCES

1. J. D. Jevons, "The Metallurgy of Deep Drawing and Pressing", 2nd edn., 1942: London (Chapman and Hall).
2. E. W. Fell, *Carnegie Schol. Mem., Iron Steel Inst.*, 1937, **26**, 123.
3. T. Sutoki, *Sci. Rep. Tôhoku Imp. Univ.*, 1941, [i], **29**, 673.
4. J. C. Arrowsmith, K. J. B. Wolfe, and G. Murray, *J. Inst. Metals*, 1942, **68**, 109.
5. R. Chadwick and W. H. L. Hooper, *ibid.*, 1951–52, **80**, 17.
6. S. Tolansky, *Proc. Roy. Soc.*, 1945, [A], **184**, 41.
7. S. Tolansky, *Proc. Phys. Soc.*, 1946, **58**, 654.

AUTOGRAPHIC LOAD/STRAIN RECORDERS

By A. BINNS,* M.I.Mech.E.

Historical

THE use of autographic load/strain recorders with testing machines is quite a long-standing practice; they were applied to some of the earliest machines both of the single-lever and compound-lever types. Generally, the load was recorded by gearing a drum or flat chart-carrier to the poise-propelling gear, and the extension was transmitted from the grip holders by means of cords and sometimes by levers in addition. In some cases clips were fixed to the gauge-length of the specimen and the extension transmitted from this point, in order to eliminate the effect of slip in the grips and extension of the specimen beyond the gauge-length. The shape of the resultant curve depended to a large extent on the skill of the operator in so propelling the poise as to maintain the beam in equilibrium, and this form of recorder could not be considered as entirely satisfactory.

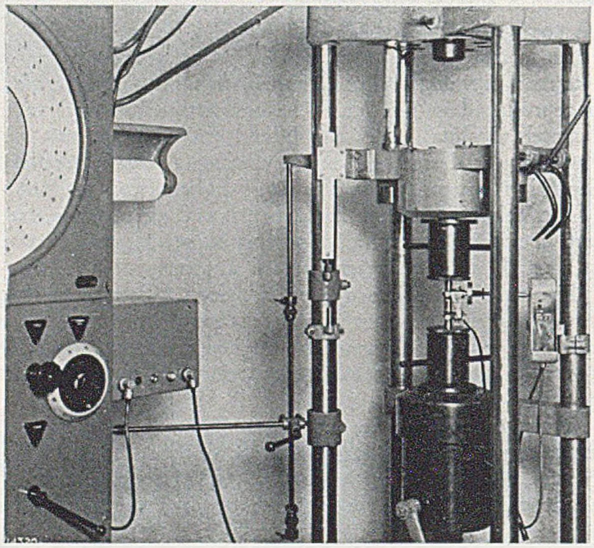


FIG. 1.—Automatic Stress/Strain Recorder Used in Conjunction with 30-Ton Testing Machine.

Wicksteed invented a recorder that was applied to Buckton machines, in which the beam was supported by a spring loaded by the setting of the poise at its maximum position. The effect of increased load on the test-piece was to reduce the pull on the spring and to cause linear movement of the beam with load; in other words, the machine was converted to an automatic spring balance. This device removed the necessity for skill on the part of the operator in propelling the poise, and consequently the records produced were very much better than on the earlier types of instrument.

Simple mechanical recorders were also fitted to dial machines, when these were introduced. One of the difficulties, again, was in connection with the load recording, as on some types of dial machine there was insufficient power in the dial mechanism to drive the pen recorder, so that the friction of this component exercised an adverse effect on the accuracy. In one type of machine this was overcome by fitting a separate load-recording system, but this had the disadvantage that, being a separate system, careful calibration was necessary to ensure that the recorder agreed with the dial indicator.

The usefulness of the recorders mentioned was limited by the fact that the magnification of the extension of the specimen was not greater than about 20 : 1. With mechanical transmission it was not practicable to employ a higher ratio, on account of the degree of friction and inertia in the drum-and-pen mechanism, which necessitated a greater amount of power than could be transmitted from the attachments to the gauge-length of the specimen. Above this ratio of magnification also the system became springy and subject to stretch of the cord, &c. However, mechanical recorders of this type and design, with magnifications up to 20 : 1, are manufactured to modern techniques and still find a useful application, particularly in technical colleges for teaching purposes and also by manufacturers interested in the behaviour of materials during the plastic stage, e.g. deep-drawn materials for press work.

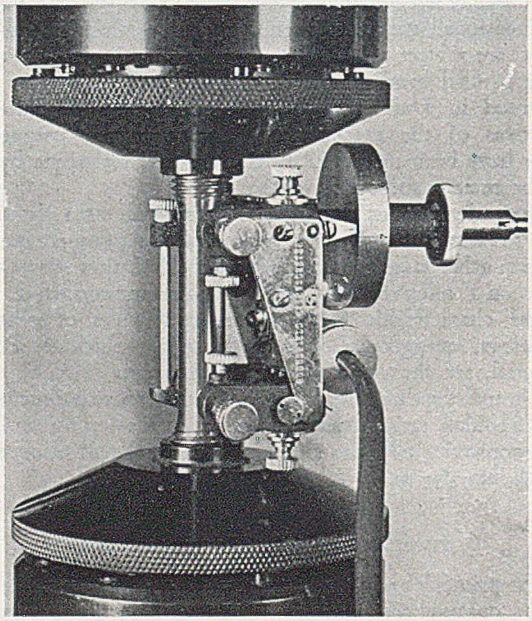


Fig. 2.—Hounsfield Extensometer Attached to Test-Piece.

The High-Magnification Recorder

The high-magnification recorder is a comparatively recent development. To be of use in exploring the elastic behaviour of a test-piece, the recorder should have a magnification of 200, 500, or even 1000 : 1, i.e. 1 in. on the diagram should be capable of representing 1/1000 in./in. of strain. With the development of modern electrical techniques, this problem has been solved.

---

* Manager, Testing Machine Division, W. and T. Avery, Ltd., Birmingham.

In the U.S.A. servo followers and electronic types are fairly well known, the most familiar perhaps being the Templin and Olsen "Atcotran". These mechanisms, however, are quite complicated, and although a large number have been made and seem to give satisfactory results in America, in this country we have not become quite so electronically minded, and the demand generally is for the simpler and more reliable methods.

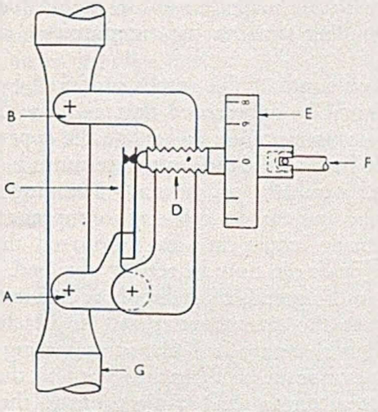


FIG. 3.—Hounsfield-Type Extensometer.

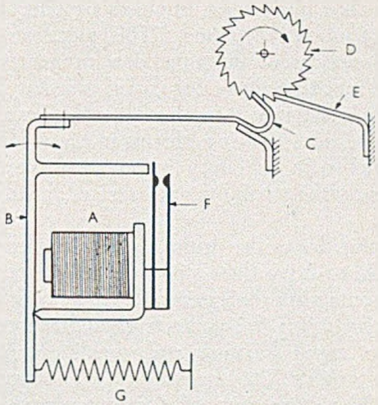


FIG. 4.—Extensometer Driving Unit.

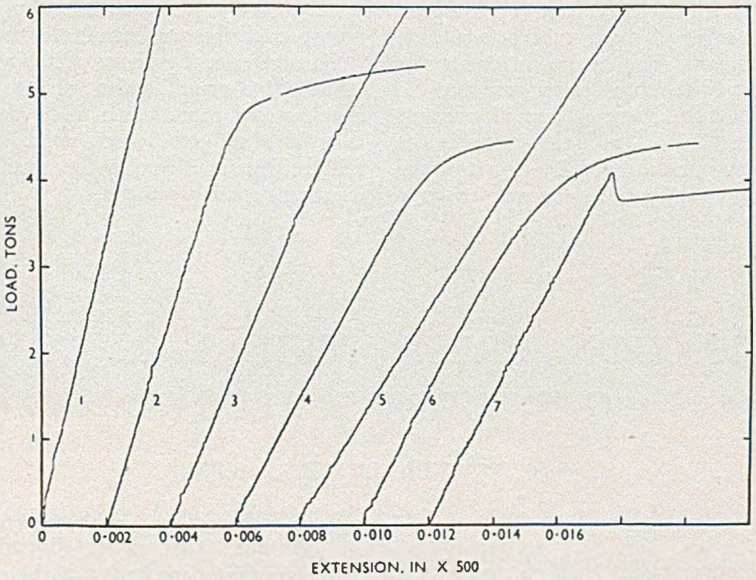


FIG. 5.—Specimen Graphs Obtained with Stress/Strain Recorder (2·00-In. Gauge-Length). Curves are displaced for clarity.

Curve 1.—0·564-in.-dia. Bright-Drawn Mild Steel. Curve 2.—0·500-in. dia. 12 : 12 Nickel–Chromium Stainless Steel. Curve 3.—0·564-in.-dia. Phosphor Bronze. Curve 4.—0·375-in.-dia. Bright-Drawn Mild Steel. Curve 5.—0·564-in.-dia. Duralumin. Curve 6.—0·500-in.-dia. Copper. Curve 7.—0·375-in.-dia. Annealed Mild Steel.

The autographic stress/strain recorder now described is designed on this basis. It consists of five main units as shown in Fig. 1, which illustrates its application to a 30-ton-capacity machine. These are the drum-and-pen mechanism, the drum driving unit, the extensometer, the extensometer driving unit, and the power unit. The load ordinate is plotted automatically by the pen, which is coupled directly to the rack which rotates the pinion connected to the pointer of the dial indicator, and as this is driven by a large pendulum ample power is available for the purpose and the friction and inertia of the recorder mechanism has no measurable effect on the accuracy of the indication.

Fig. 2 shows the extensometer, which is attached to the test-piece and is a standard Hounsfield type. This type of extensometer consists of two arms (*A*) and (*B*) fixed to the gauge-length of the test-piece (*G*) by means of spring-loaded cone pivots, and the one arm carries a pair of electric contacts (*C*) associated with a micrometer screw (*D*) on the other arm. This is shown diagrammatically in Fig. 3. The effect of extension of the test-piece is to open these contacts. Normally, a circuit is made through a battery and indicating lamp, so that immediately the test-piece extends, the light goes out. Contact is restored by rotating the micrometer drum (*E*). This is graduated, and accordingly a series of plots can be made as the test-piece extends. The sensitivity of the contacts is such that a movement of 1/50 of 1/1000 in. can be recorded. Instead of the micrometer drum being rotated by hand, the recorder does this automatically.

A diagrammatic representation of the extensometer driving unit is given in Fig. 4. It is very simple and consists essentially of an automatic telephone stepper switch. When energized, the magnet (*A*) draws the armature (*B*) towards it and moves the driving pawl (*C*) into the next tooth of the ratchet (*D*). The magnet is then de-energized and the spring return (*G*) drives the ratchet forward one tooth. The operation is similar to that which takes place when impulses are sent out from the dial of an automatic telephone. This piece of mechanism has been the subject of intensive development over many years and is mechanically safeguarded so that the ratchet cannot slip back and miss a tooth or travel only a portion of the tooth. The duty that it performs in the recorder is very light compared with the duty in an automatic telephone, and consequently the reliability and service life are very much extended.

A similar unit drives the drum of the recorder. It would not be possible to drive these stepper units directly from the micrometer contact on the extensometer, as both the current and the voltage across the contacts must be kept as low as practicable to avoid burning and welding. Accordingly, a power unit provides the necessary type of current and also the relays. The operation of the extensometer contacts controls a very sensitive Post-Office-type relay, which in turn controls a sequence relay equipped with a maintaining contact, so that once motion is initiated at least one complete step, or an equal whole number of steps, must be made on each unit before motion ceases. The extensometer driving unit always operates ahead of the drum driving unit and the normally open contact (*F*) on it then closes and supplies current to the drum driving unit. When the latter operates, its normally closed contact opens the maintaining circuit of the sequence relay and if, in the meantime, the micrometer screw has advanced sufficiently to break the extensometer relay contact, motion ceases. If not, the complete cycle is repeated.

By this means the drum is rotated and the micrometer screw is advanced in synchronism by a series of equal pulses. These pulses produce a diagram in the form of a series of closely spaced steps either at 100 to the inch or 50 to the inch, in accordance with the magnification in use on the drum unit. The corresponding steps on the extensometer are 50 or 10 to 1/1000 in.

Fig. 5 shows some of the graphs produced by this instrument. It should be appreciated that each small step represents the extensometer plot, and while the appearance of the curves would be better if the steps were missing, the accuracy of the record would be no greater. The graphs shown are very early ones and contain one or two imperfections. Since they were made, experience has improved the technique, and perfect graphs can now be readily obtained. Such faults as there are in the graphs reproduced are not in the recorder, but are due to the extensometer and its attachment to the test-piece. It will be realized that in measuring to such fine limits, great care must be taken in regard to the attachment of the extensometer to the test-piece and to the axial alignment of the test-piece in the testing machine. The recorder will reproduce only what the extensometer tells it, and it is often found that the recorder diagnoses some hitherto unsuspected ailments, both in the test-pieces and the machines.

This recorder, by virtue of its simplicity and the ease with which high-magnification records can be made, should enable much more work to be done on the behaviour of materials during the elastic stage. It is believed also that it will facilitate considerably the obtaining of proof-stress results in routine test laboratories.

SPECIFICATION CREEP TESTING OF NIMONIC GAS-TURBINE ALLOYS

By C. W. WEAVER,* B.Sc., A.I.M.

Introduction

The paper describes a machine developed to meet the growing demand for short-time creep tests, such as those required by D.T.D. Specifications 725, 736, and 747, on materials for gas-turbine rotor blading and other high-temperature applications. Compactness was important, combined with simplicity of design which would allow manufacture from commercial components, by makers not specialists in such equipment.

The machine, of the single-lever type and with a capacity of 1·5 tons, is arranged for continuous furnace operation, the test-piece and shackles being changed "hot". Compared with the usual procedure, this method reduces overshoot, simplifies control of temperature gradient, and, since the temperature is steady again in about 2 hr., saves considerable time.

The creep laboratory is situated in a basement, formerly an air-raid shelter, which after demolition of partition walls gave a T-shaped area, with a "cross-bar" 52 × 18 ft. and an "upright" 36 × 24 ft., all with 7 ft. 6 in. head room. This accommodates 101 machines, with work benches, desks, storage cupboards, and maintenance gear. The design of the machine was governed by the limited space available and by the need to fit the units along the walls and between pillars. To save wall space, the electrical control panels are mounted on the machines, and, for simplicity, a mechanical thermostat is used for furnace-temperature control. A ventilation fan feeds 3000 ft.^{3} of air/min. through ducts near the ceiling.

A typical part of the installation is shown in Fig. 1.

Requirements of Routine Creep Testing

For routine testing a production-line approach was necessary, and the machine had to use the same test-piece and to work interchangeably with Denison T.46 radial machines already in service. It was designed to the following specification:

---

* Development and Research Department, The Mond Nickel Co., Ltd., Birmingham.

(1) specimen size 1/20 in.^{2}, 3-in. gauge-length, $\frac{1}{4}$-in. Whitworth thread; (2) (*a*) load capacity $1\frac{1}{2}$ tons; (*b*) lever ratio 11·2 : 1; (3) temperature control $\pm\frac{1}{2}°$ C.; (4) furnace-temperature gradient 1° C.; (5) height 6 ft.; (6) floor area 20 × 30 in.; (7) furnace size 12 in. dia. × 15 in. long, hinged in a vertical line (this was the largest size of furnace possible); (8) shackle attachments by pin universal joints; (9) direct-straining screw attachment; (10) all-electrical control gear and thermostat apparatus mounted on the frame; (11) electric mains supplies fed by bus-bars from a tapped transformer, and the electrical supplies to the thermostats and time-counters similarly "piped" from a common 24-V. supply; (12) mechanical thermostat; (13) extensometer (i) for stress-rupture testing, a beam-operated dial gauge, (ii) for medium-sensitivity testing, a micrometer extensometer mounted on the specimen.

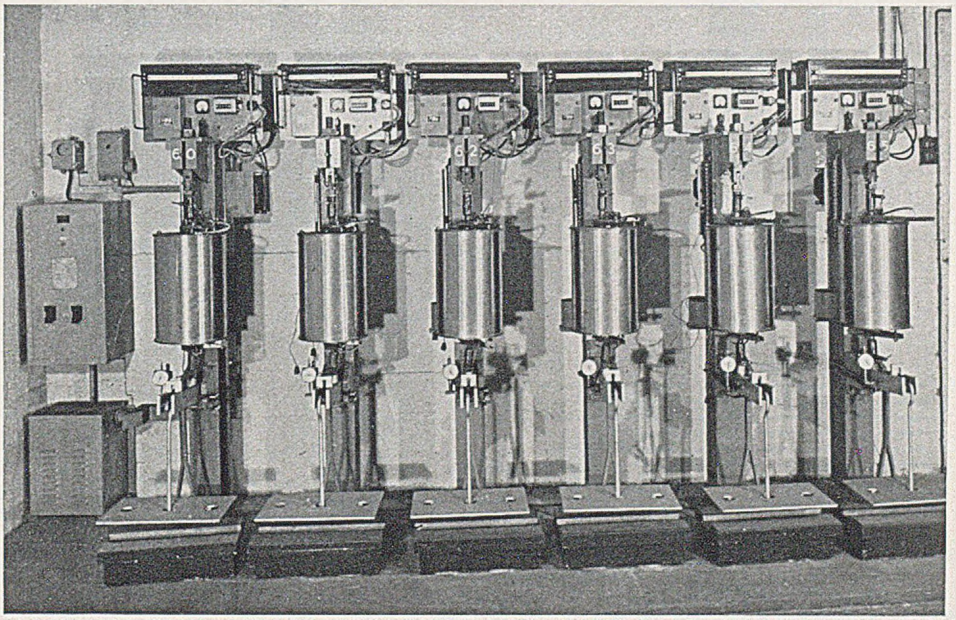


Fig. 1.—Group of Six Machines with Transformer, Rectifier, and Time Switch.

The specifications for Nimonic 80A and Nimonic 90 require tests of 75 hours' duration, at 17 and 19 tons/in.^{2} respectively, in both cases at 750° C. Tests of long duration up to 900° C. are also required on experimental alloys.

Description of the Machine Frame

Fig. 2 shows the complete machine. The 8 × 6 in. **I** section *A* is welded to two 3 × 3 in. angle base members *B*, Rawl-bolted to the floor. At the top is welded a gusseted bearer-plate *C*, to support the shackle assembly, and a bracket *D* for the electrical panel. Lower down is bolted the loading-beam fulcrum *E*, and to the base angles is welded a support post. To relieve the test-piece of load, the beam is raised by a hand lever, with the top of this post as fulcrum, and is then supported by a peg inserted in the post. Furnace hinges and an extensometer bracket are also attached to the frame.

On the plate *C* a seating for the straining screw is drilled and pin-faced, and the fulcrum *E* is assembled and the machine levelled by a jig registering from the hole in this seating, to give axiality of load on the test-piece.

Two steel side plates, broached to take the three knife edges and joined by a welded steel stirrup, form the beam. Each beam is calibrated with its scale pan in a tensile machine, the pan and lever weight, made equivalent to 0·1-ton load, being adjusted to be exactly correct at the middle of the load range. The beams are accepted if the maximum error in $1\frac{1}{2}$ tons is within ±0·3%.

Electrical Control Equipment

Each group of six machines has its own mains transformer for furnace supplies, and a 24-V. transformer and rectifier. The former has tappings at 110, 140, 170, 200, and 230 V. A.C., to bus-bars mounted in trunking and feeding small socket panels behind each machine, from which the voltage most appropriate to the operating temperature is selected. Another trunking contains two 24-V. D.C. lines, one for the thermostat switch and relay circuits and the other for the hour-counters.

Connections are shown in Fig. 3 (*a*). The potentiometer rheostat *P* divides the current between the top and bottom furnace windings, which are in parallel. A choke Q, in series with the furnace, is short-circuited intermittently by a relay *R* operated by the mechanical thermostat switch *T*, so giving control by fluctuating input.

The hour-counter records to 999·9 hr. and is operated every 6 min. by a master time-switch common to six machines, the counters being wired in parallel. A manual switch S_2 and a micro-switch S_3, tripped by the beam when the specimen breaks, control each counter.

Connections to the control panel, except the neutral lead, are made by multi-pin plugs and sockets, the 24-V. supply being separate from the mains connections.

Construction of the Furnace

To provide an expansion member for the thermostat, a metal tube was necessary, and to give robustness and oxidation-resistance at high temperature, Inconel was chosen.

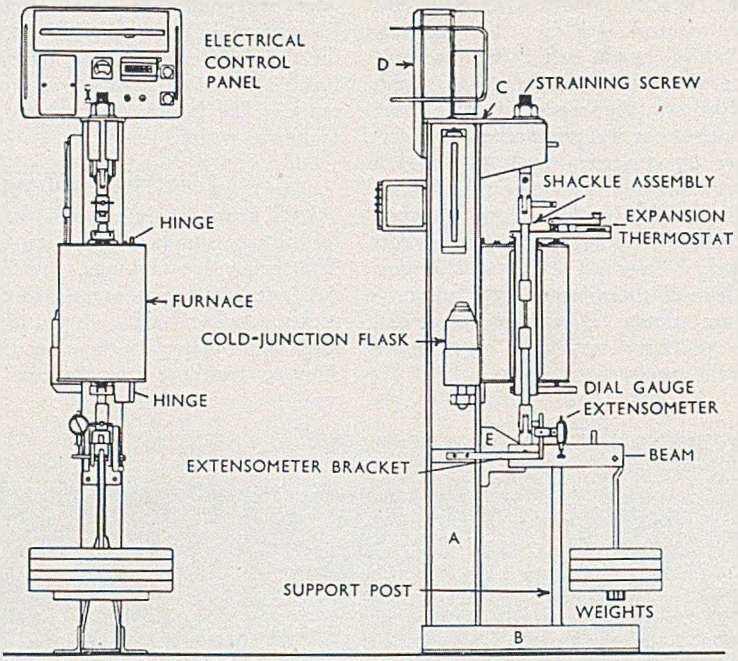
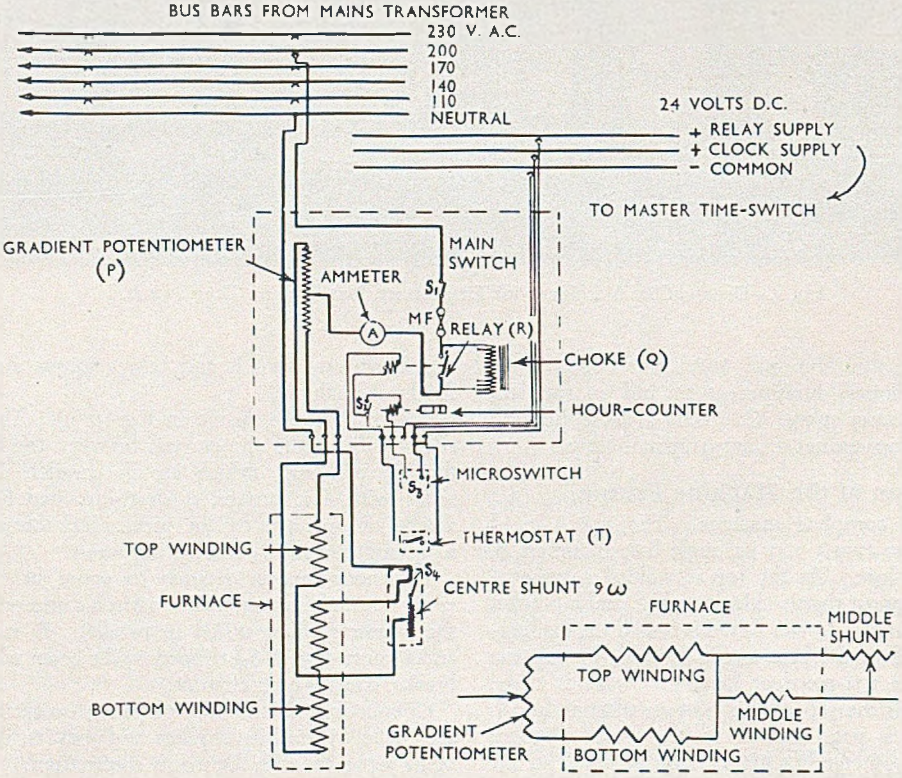


Fig. 2.—General Assembly of Stress-Rupture Creep Machines.



(a) (b)

Fig. 3.—(*a*) Wiring Diagram. (*b*) Furnace-Circuit Diagram.

The circuit, Fig. 3 (*b*), with top and bottom heating windings in parallel, fed by the potentiometer rheostat *P*, permits wide variation of the relative currents without appreciable change of total input. The furnace-element construction is shown in Fig. 4. The tube is insulated, before winding, with $\frac{1}{8}$ in. of special alumina cement. The main windings are first spiralled on a $\frac{1}{4}$-in. former to give the necessary high resistance in the limited length and each extends over $4\frac{1}{2}$ in., with a 4-in. gap between them. To trim the temperature at the middle of the test-piece, a small straight-wire winding is wound over the 3-in. middle section in series with the rest, and having a shunt resistance for adjustment. With the windings in place, insulating cement is brushed on, dried in successive layers and hardened by drying at 300° C., baking for 4–5 hr. at 1000° C., and then cooling to below 500° C., when it forms a strong, hard white brick. The insulation should be 1–2 megohms at 500 V. The furnace leads are insulated with beads and are brought to terminals on the top plate near the hinge. The casing, which is made as large as practicable, is then packed with white diatomaceous earth, previously fired at 800° C. to drive off volatile carbonaceous material.

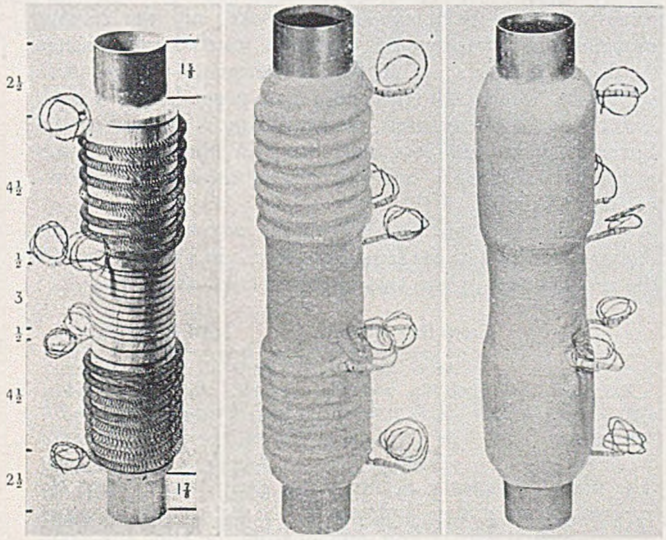


Fig. 4.—Details of Furnace-Tube Construction.

Temperature Control

The expansion controller is shown in Fig. 5. The Inconel furnace tube has two metal arms securely fixed to its ends which project beyond the casing. The lower arm carries a Nilo K * rod which passes up through the furnace body to the lever supported on the top arm, by which the difference in thermal expansion is magnified and transmitted to a small switch. The switch consists of two light levers in a moulded Bakelite box: one of these carries electrical contacts operating the 24-V. D.C. circuit which actuates the mains relay. The external lever is 10 : 1, and those in the box are 24 : 1, giving a magnification of 240 : 1. Coarse setting is by a 40 thread/in. screw bearing on the Nilo K rod, and fine adjustment is by a 40 thread/in. micrometer head at the end of the main lever. The temperature swing varies from $\pm\frac{1}{4}$° to $\pm\frac{1}{2}$° C., and the mean temperature can be held to $\pm\frac{1}{2}$° C. for long periods.

Operation of the Machine

The specimen, after measurement, is assembled in the shackles and adaptors, in a jig which fixes its position relative to the bottom shackle, thus controlling the vertical position in the furnace to facilitate temperature-gradient control. Thermocouples are tied on $\frac{1}{4}$ in. from the ends of the gauge-length, and the assembly is ready for insertion in the furnace. A third thermocouple can be tied at mid-length for gradient checking, but once the middle winding has been adjusted, this is found to remain steady. Fig. 6 shows the insertion into the furnace, the whole procedure being as follows: on conclusion of a test a pair of self-locking grips is fixed to the top shackle above the furnace. The beam and scale pan are lifted and pegged, using the long hand lever and the support post. The top and bottom shackle pins are then removed with a second pair of grips, so that the assembly is now supported by the top self-locking grip. The furnace is swung to the right, the packing is removed, and the assembly is lifted out from the top and placed to cool. If the bar is broken the bottom shackle must first be withdrawn downwards with the second pair of grips. The process is reversed to insert the new assembly. The thermocouples emerge at the top of the furnace. Platinum/platinum–10% rhodium thermocouples, 2 ft. 9 in. long, are used, with a thermos-flask cold junction on the side of the machine. At the cold junction, there are two-pin connectors to copper leads running to a multi-point potentiometer desk. Temperature is steady

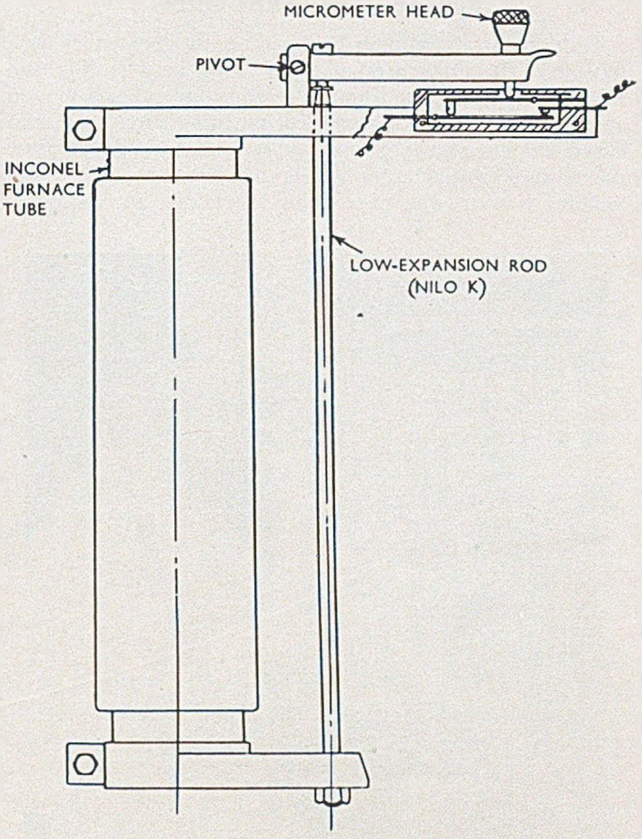


Fig. 5.—General Arrangement of Thermostat.

---

* Nilo K is a Ni–Fe–Co controlled-expansion alloy, produced by Henry Wiggin and Co., Ltd.

within 2 hr., during which time gradient and overall adjustments are made. The test-piece can then be loaded, although in order to conform with the recommended "16–24 hr. at test temperature" of the British Standard Specification, the usual practice is to load next day.

Furnace-Gradient Control

Each furnace is calibrated with three thermocouples on a test-piece, the differences of temperature between top and middle, and bottom and middle thermocouples being plotted against the setting of the gradient potentiometer. The two curves will intersect at the point where there is no temperature difference between the top and bottom couples. The centre winding is then adjusted to bring the middle thermocouple to the same reading. After this adjustment fresh curves can be drawn, and it has been found that from test-piece to test-piece, provided the furnace position is constant, no subsequent adjustment of the centre winding is necessary. Then, in testing, if the top and bottom thermocouples show a gradient, the calibration curve shows how the gradient potentiometer needs to be adjusted.

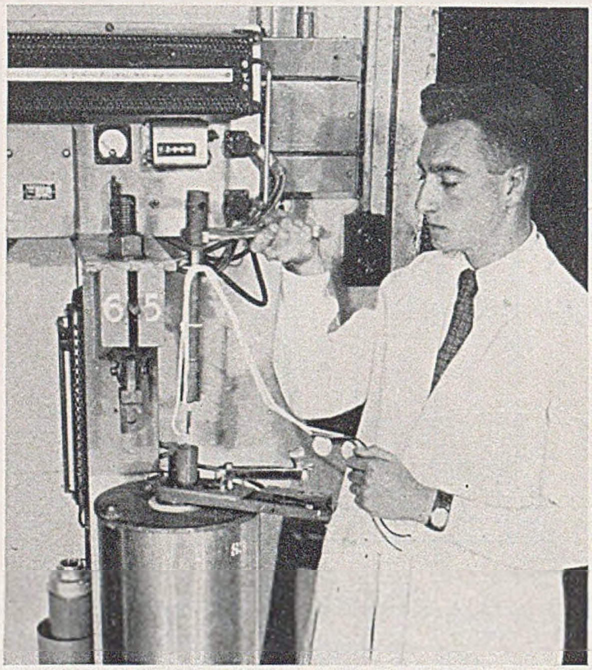


Fig. 6.—Method of Inserting Test-Piece into Hot Furnace.

A check with 12 couples along the 3-in. gauge-length, after careful adjustment, showed only ±0·3° C. variation.

Extensometers

For the short-time specification and exploratory work, a dial gauge mounted on a bracket from the pillar *A* (Fig. 2) records the movement of a horizontal lug on the beam 5 in. from the fulcrum. This gives a magnification of 5 times the extension, so that a 0·0005-in. division on the gauge equals 10^{-4} in. at the test-piece. Since such readings include extension and thermal expansion of shackles and adaptors, the apparent creep rate is slightly greater than the true rate. The nominal rate measurement is, however, adequate for production control and exploratory testing. To prevent damage to the gauge from thermal contraction if a furnace fails, the gauge is mounted on a pivot and held vertical by a spring clip, a short brass stop-sleeve being fitted to the plunger to limit the travel and then to transfer thrust from the mechanism to the body of the gauge. The thrust springs the gauge from the clip to swing on the pivot.

A micrometer extensometer for more precise reading has been adapted from that described by McKeown.[1] Two yokes screw on to the threads of the test-piece, and are locked by thin nuts. Rods from the top yoke pass down through tubes supported from the bottom yoke, and the relative movement is measured by 0·0001-in. ratchet micrometers fixed on the tubes. With this instrument a 6-in. test-piece with ¾-in. Whitworth ends and a 3-in. gauge-length is used. Except for the micrometers, the instrument is made of special materials, yokes and lock nuts being in Nimonic 80, rods and tubes of Inconel, and micrometer mountings of low-expansion Nilo K.

Tests have been made with both the extensometer and the dial gauge in use on the same test-piece. In a test at 800° C. the results given by the two instruments are as follows:

Type of Extensometer	Creep Rate, 10^{-6}/hr.		Total Creep, %		
	At 1000 Hr.	At 2000 Hr.	At 500 Hr.	At 2000 Hr.	At 3500 Hr.
Dial gauge . .	6·1	7·3	0·52	1·50	2·70
Micrometer .	5·3	6·8	0·25	1·10	2·25

The main discrepancy occurs in the early stages, when the apparent primary stage is affected by the shackles and adaptors.

Two tests at 800° C., on samples from the same bar, compared the micrometer extensometer in the machine described, with the Martens mirror-type in a high-sensitivity machine. Creep rates were low, the high-sensitivity test giving 2×10^{-8}/hr. at 2000 hr., compared with 4×10^{-8} for the stress-rupture machine, with total strains, at 2500 hr., of 0·026% and 0·037%, respectively.

Acknowledgements

The author wishes to thank the Directors of The Mond Nickel Co., Ltd., and particularly Dr. L. B. Pfeil, who suggested the development of the machine, for permission to publish this paper. He gratefully acknowledges the help and advice afforded by the Chief Engineer's Department of Henry Wiggin and Co., Ltd.

REFERENCE

1. J. McKeown, *Metallurgia*, 1950, 42, 189.

1442

ATMOSPHERIC CORROSION AND STRESS-CORROSION OF ALUMINIUM–COPPER–MAGNESIUM AND ALUMINIUM–MAGNESIUM–SILICON ALLOYS IN THE FULLY HEAT-TREATED CONDITION[*]

By G. J. METCALFE,[†] M.Sc.Tech., MEMBER

SYNOPSIS

The corrosion behaviour of the aluminium alloys H10-WP and H15-WP in the extruded form has been determined in the stressed and unstressed condition by exposure to sea-water, river water, and various natural atmospheres. The corrosion attack was assessed by visual and microscopical examination and by tensile tests on the corroded material. The most severe attack of both alloys resulted from exposure to the industrial atmosphere of Sheffield, where the average loss of strength after 2 years' exposure was approximately 11%, which is equivalent to a loss of thickness of 0·012 in. There was no indication of stress-corrosion failure of either of the alloys at any of the exposure sites. The high stress-corrosion resistance of the H15-WP alloy is somewhat surprising, since in sheet form the alloy is known to be very susceptible to stress-corrosion failure. The absence of such failure is attributed to preferential attack ("foliation") occurring along grain boundaries and bands parallel to the direction of extrusion which redistributes the concentration of stress at corrosion pits. The rate of loss of strength of both alloys exposed at Sheffield, and of H15-WP alloy exposed to a marine atmosphere, was found to decrease with time, apparently exponentially. Scatter of the results from the remaining sites was appreciable, but it was clear that in general there was a decrease in the rate of corrosion with time.

I.—INTRODUCTION

In order to assess the corrosion behaviour of the aluminium alloys H15-WP and H10-WP in the extruded form when exposed to atmospheric corrosion and to corrosion by salt water and river water, a series of field tests were begun at different stations throughout the country.

Most workers agree that in the early stages of film formation the corrosion/time curve is a logarithmic curve, the rate of corrosion decreasing with time. Early investigation was confined to work on aluminium of varying degrees of purity, using different indices of corrosion. Bryan [1] and Morris [2] determined the loss in weight during corrosion, and Champion [3] measured the quantity of gas evolved. Vernon,[4] on the other hand, determined the increase in weight in his study of films formed in air on aluminium. Guilhaudis,[5] in very recent work, has determined the ultimate tensile strength and elongation of corroded aluminium–5% magnesium alloy and Duralumin exposed to marine conditions.

Champion,[6] in considering the results of his work together with work by other investigators, has suggested that the corrosion/time curve is exponential. He further maintains that, in studying the results of field tests, the corrosion/time curves may be considered to follow an exponential relationship, although there may be an induction period during which no corrosion is observed. This point is illustrated by drawing curves plotted from the average results obtained by other investigators who measured the corrosion effect by determining loss in weight, loss in ultimate tensile strength, loss in elongation, &c. These corrosion/time curves are asymptotic to a limiting value, so that any increase in metal thickness beyond a certain minimum, which corresponds to this limiting corrosion value, may be expected to ensure that the material will never fail by whatever parameter the corrosion is being measured, i.e. loss in strength, pitting, &c.

It is possible to calculate this minimum thickness, and Champion [7] has shown, by taking as an example the corrosion of 17S-T alloy (exposed at La Jolla (California) in field tests of the American Society for Testing Materials [8]) that the residual strength of the material will not fall below that of the original 0·1% proof stress if the thickness is 0·055 in. If, therefore, a safety factor of 2 is adopted arbitrarily, the minimum thickness that should be used for structural purposes at La Jolla would be 0·11 in. Champion has, however, made no firm recommendations as to the safety factor that should be adopted to allow for variations in conditions and for scatter in results on individual specimens.

In the present investigation it was hoped that, by the examination of specimens at suitable intervals, enough information could be obtained in a com-

---

[*] Manuscript received 20 October 1952.

[†] Senior Research Metallurgist, Precious Metals Division, The Mond Nickel Co., Ltd., Acton; formerly at the Fulmer Research Institute, Stoke Poges, Bucks.

paratively short time to allow a confident forecast to be made of the probable behaviour of H15-WP and H10-WP alloys in a variety of service conditions. The effect of stress acting in conjunction with corrosion by the exposure of specimens stressed by bending under constant strain in special jigs was also investigated. Most of the material used was similar, both as regards composition and heat-treatment, to that normally used in production.

In the present paper the average tensile-test results obtained at intervals during 2 years' exposure, are summarized graphically and in tables. Detailed results of the visual and microscopical examination and tensile testing have been deposited in the library of the Institute, and may be consulted there.

II.—EXPERIMENTAL DETAILS

1. MATERIAL

Channel-section extrusions were made from normal batches of H10-WP and H15-WP alloy billets. The sections were 1¼ × 1 × 1 in. and were 0·1 in. thick. The extrusions were made by the normal production method, the H10-WP alloy extrusions being die-quenched and the H15-WP alloy extrusions being quenched after solution heat-treatment. After quenching the extrusions were straightened by stretching, the amount of stretching usually being of the order of 3%, although this figure may have varied between 1 and 8% in practice. Lengths were cut as required and were subjected to the appropriate ageing treatment. Chemical analysis of the extrusions gave the following results:

	Cu, %	Mg, %	Mn, %	Si, %	Fe, %	Ti, %
H10-WP . .	0·04 max.	0·68	0·04	0·98	0·27	...
H15-WP . .	3·81	0·74	0·54	0·95	0·25	0·02

The ageing treatments used for the field test-specimens were 8 hr. at 175° C. for H10-WP and 8 hr. at 170° C. for H15-WP.

2. EXPOSURE SITES AND METHOD OF EXPOSURE

Sites for the tests were chosen in different parts of the country, so that the behaviour of materials in different atmospheric environments could be assessed. Stressed, as well as unstressed specimens, were exposed in frames at each of these sites. The method of stressing was by 4-point loading, so that the middle 7 in. of the test-piece, which was ¾ in. wide and 12 in. long, was stressed in uniform bending at constant strain. Two stresses were used for each material. (i) An arbitrarily chosen design stress: 5·1 tons/in.^{2} for H10-WP and 16 tons/in.^{2} for H15-WP alloy, and (ii) the 0·1% proof stress: 15 tons/in.^{2} for H10-WP and 24 tons/in.^{2} for H15-WP alloy. Fig. 1 (Plate XXXIX) shows one of the frames with specimens in position.

Pieces of channel-section extrusion cut from the same lengths of extrusion as those from which the stressed specimens had been cut, were attached to the stress-corrosion frames and were removed for examination and testing at the same time as the stressed test-pieces.

The nature and location of the sites and the visual appearance of the specimens were as follows:

Site No. I. Severe Industrial Atmosphere.—The British Iron and Steel Research Association's exposure site at the Brown-Firth Research Laboratories, Princess Street, Sheffield.

This site is by the side of a very busy railway siding, and the atmosphere is charged with soot and sulphurous fumes from burning coke, coal, and neighbouring steel-furnace plant. The frames were placed in a horizontal position on stands about 2 ft. 6 in. from the ground; the specimens rapidly became coated with soot and particles of solid matter from the steel-furnace smoke.

Site No. II. Normal Industrial Atmosphere.—The British Non-Ferrous Metals Research Association site on the roof of their building in Euston Street, London.

The frames were placed in a horizontal position on the flat roof of the building, which is only a short distance from Euston main-line railway station. The environment is typical of a normal industrial locality with smoke from factory and household chimneys polluting the atmosphere. The specimens gradually became coated with a fine deposit of soot.

Site No. III. Marine Atmosphere.—An Admiralty raft off Hayling Island in Chichester harbour (by arrangement with the Central Metallurgical Laboratory).

The specimens were exposed to the sea-water spray swept up to them by wind and, in heavy weather, to waves breaking over the raft on which the frames were fixed. The frames were in a vertical position attached to supports on the raft about 7 ft. above water level, and the raft was moored a short distance off-shore. The specimens soon became coated with a deposit of sea-salt and a certain amount of corrosion product.

Site No. IV. Rural Atmosphere.—Fulmer Research Institute, Stoke Poges, Bucks.

The site is in an open field in a rural district about 2½ miles north of Slough. The frames were supported in a horizontal position about 6 in. above the ground. The specimens showed little evidence of corrosion over a very long period, apart from a slight darkening in colour.

Site No. V. Total Immersion in Sea-Water.—Admiralty raft off Hayling Island in Chichester harbour (by arrangement with the Central Metallurgical Laboratory).

The frames were supported on jigs in a vertical position on the raft used at Site No. III. The raft was bottomless, and the frames were immersed in sea-water to a depth of about 3 ft. The specimens became covered rapidly with fouling.

Site No. VI. Tidal Immersion in Sea-Water.—The Central Metallurgical Laboratory's site at Hayling Island.

This site is at half-tide level at Hayling Island. The frames were fixed in a vertical position and were completely immersed at high tide and completely free at low tide. Owing to the severe battering by waves in heavy weather the specimens were easily damaged.

Site No. VII. Total Immersion in Fresh Water.—River site at the Military Engineering Experimental Establishment, Christchurch, Hants.

The River Stour at Christchurch is tidal, and the frames of specimens were totally immersed in the river in a vertical position. Normally the frames were about 4 ft. below the surface, just clear of the river-bed, but owing to changes in current, &c., they were frequently buried in river mud. The specimens soon became coated with an organic film, possibly from sewage that is emptied into the river.

Site No. VIII. Tidal Immersion in Fresh Water.—River site at the Military Engineering Experimental Establishment, Christchurch, Hants.

By placing the frames in a horizontal position under a concrete pier, it was possible to subject the specimens to tidal immersion during normal and dry weather conditions. When flooding occurred, however, the frames were frequently totally immersed for several days at a time. The specimens became coated rapidly with a film of mud.

Site No. IX. Tidal Immersion in Sea-Water.—British Railways Docks, Southampton.

The frames were fixed at half-tide level under a concrete pier in one of the docks. The sea-water was contaminated with oil and bilge water from passing ships, and a greasy film soon formed on the specimens.

III.—RESULTS OF FIELD TESTS AFTER TWO YEARS' EXPOSURE

1. General Details of Tests

Specimens were removed from the sites for examination at intervals, but, owing to heavy storms and possibly to removal by unauthorized persons, specimens exposed at the tidal site on Hayling Island (Site VI) were missing at the end of 2 years' exposure. No results are available, therefore, after 2 years' exposure at this site. The corroded specimens were visually examined and subjected to tensile tests, a 4-in. test-piece cut from mid-way between the centre supports being used for determining the residual strength of the material. The remainder of the stressed portion of the exposed specimen was subjected to microscopical examination.

Tensile tests were also made on test-pieces cut from the web and flanges of pieces of unstressed channel-section which had been exposed to the same corroding conditions. Control tests were made on test-pieces cut from material from the same batch which had been stored in the laboratory and tested at the same time as the corroded test-pieces.

2. Visual Examination

After each period of exposure the specimens were visually examined. Detailed results are summarized for the different sites in Appendix I (p. 276). The conditions of exposure varied appreciably from one site to another, and it seemed that the heavy industrial atmosphere of Sheffield had an even more serious effect on the corrosion-resistance of the material than the marine atmosphere. The corrosion attack on totally immersed or intermittently immersed specimens was usually confined to isolated spots. This may have been due to local effects such as deposition of cathodic particles on the surface, or to local breakdown of a protective film. The high resistance of the material to immersion in sea-water seems to have been due to protection by marine growth, and, in the case of specimens exposed at Southampton Docks, also to the deposition of a thin film of oil.

3. Microscopical Examination

From each corroded test-piece a specimen approximately $\frac{1}{2}$ in. long was removed and sectioned longitudinally for microscopical examination at a magnification of 100. The method of assessment and description of the corrosion damage are given in Appendix II (p. 277).

Microscopical examination has shown that the corrosion attack of the H10-WP alloy is usually of the intercrystalline type and varies in intensity to an appreciable extent from site to site. The most severe attack was at Sheffield (Site I), and after 2 years it was fairly general. Fig. 2 (Plate XXXIX) shows typical attack after this period. At Euston (Site II), which is semi-industrial, the corrosion was much less severe, as shown in Fig. 3 (Plate XXXIX), and at the marine-atmosphere site off Hayling Island (Site III), the attack was somewhere between that experienced at Sheffield and that at London. Fig. 4 (Plate XXXIX) shows typical attack on exposure to a marine atmosphere. The remaining atmospheric-exposure site, i.e. the rural atmosphere at Stoke Poges (Site IV), caused very slight intercrystalline corrosion at isolated spots, as shown in Fig. 5 (Plate XXXIX). In the total-immersion tests on the H10-WP alloy at Christchurch (Site VII) the attack (see Fig. 6, Plate XXXIX) was fairly severe at isolated spots and in some instances penetrated the thickness of the specimens. In spite of this, however, the results of tensile tests, which are described in the next section (p. 273), show that the residual strength after 2 years' exposure ranged from 13·4 to 18·9 tons/in.² and the average loss in strength was only 1·3 tons/in.². Total immersion in sea-water off Hayling Island (Site V) caused much less severe attack. The results of tidal immersion at Christchurch (Site VIII) showed negligible attack after 2 years' exposure. At Southampton Docks (Site IX), where the specimens were exposed under a concrete pier at half-tide level, isolated spots of intercrystalline attack had occurred.

As judged by visual and microscopical examination, the corrosion of H15-WP alloy was more severe, in general, than that of H10-WP. The attack was frequently of the foliation type accompanied by intercrystalline corrosion which, however, was preferentially orientated longitudinally parallel to the direction of extrusion. Intercrystalline corrosion of the longitudinal type appeared to start first and was followed by foliation of the surface layers. In some cases where attack of the sub-grain boundaries started, the foliation was much more severe. Corrosion attack was most severe at Sheffield (Site I) as illustrated in Fig. 7 (Plate XL). Attack of the sub-grain network is shown in Fig. 8 (Plate XL). At Site II (Euston) the attack was of the same type, though much less severe (Fig. 9, Plate XL), and that at Hayling Island (Site III) (Fig. 10, Plate XL), was intermediate between that at Sheffield and at London. Fig. 11 (Plate XL) shows typical attack of the sub-grain network. The foliation type of attack on the H15-WP alloy was obvious by eye only after the specimens had been exposed for more than 1 year, and in some specimens it was marked at some points which would not necessarily form part of the tensile test-piece machined from the specimen. In one case an exceptionally low tensile-test result was obtained (20·3 tons/in.^{2}), which was associated with foliation. No obvious attack was observed in the total-immersion tests at Hayling Island (Site V) and at Christchurch (Site VII) only isolated spots of longitudinal intercrystalline corrosion were present after 2 years' total immersion. The attack was even less severe in the intermittent-immersion tests at Christchurch (Site VIII), though at the isolated spots where attack occurred, cavities such as those shown in Fig. 12 (Plate XL) were observed.

TABLE I.—*Average Tensile Strength and Loss in Strength of H10-WP Alloy Specimens Exposed at Various Sites.*

Original U.T.S. = 18·8 tons/in.^{2} (average of 9 tests).

Site	Property	Exposure Time, months				
		1	3	6	12	24
I	U.T.S., ton/in.^{2}	18·4	17·9	17·7	16·7	16·8
	Loss in strength, ton/in.^{2}	0·4	0·9	1·1	2·1	2·0
II	U.T.S., ton/in.^{2}	18·2	18·7	18·5	17·8	17·6
	Loss in strength, ton/in.^{2}	0·6	0·1	0·3	1·0	1·2
III	U.T.S., ton/in.^{2}	18·9	18·7	18·6	18·2	17·6
	Loss in strength, ton/in.^{2}	−0·1	0·1	0·2	0·6	1·2
IV	U.T.S., ton/in.^{2}	18·6	18·7	19·0	18·4	18·5
	Loss in strength, ton/in.^{2}	0·2	0·1	−0·2	0·4	0·3
V	U.T.S., ton/in.^{2}	18·6	18·9	18·2	18·3	17·4
	Loss in strength, ton/in.^{2}	0·2	−0·1	0·6	0·5	1·4
VI	U.T.S., ton/in.^{2}	18·6	19·1	18·6	18·4	No results
	Loss in strength, ton/in.^{2}	0·2	−0·3	0·2	0·4	
VII	U.T.S., ton/in.^{2}	18·7	18·4	18·2	16·7	17·5
	Loss in strength, ton/in.^{2}	0·1	0·4	0·6	2·1	1·3
VIII	U.T.S., ton/in.^{2}	18·4	18·0	18·3 (8)*	18·2	18·1
	Loss in strength, ton/in.^{2}	0·4	0·8	0·5	0·6	0·7
IX	U.T.S., ton/in.^{2}	18·6	18·8	19·2	17·7	18·6 (8)*
	Loss in strength, ton/in.^{2}	0·2	0	−0·4	1·1	0·2

TABLE II.—*Average Tensile Strength and Loss in Strength of H15-WP Alloy Specimens Exposed at Various Sites.*

Original U.T.S. = 31·2 tons/in.^{2} (average of 9 tests).

Site	Properties	Exposure Time, months				
		1	3	6	12	24
I	U.T.S., ton/in.^{2}	30·7	29·7	28·9	27·8	27·5
	Loss in strength, ton/in.^{2}	0·5	1·5	2·3	3·4	3·7
II	U.T.S., ton/in.^{2}	30·0	30·4	29·9	29·4	29·8
	Loss in strength, ton/in.^{2}	1·2	0·7	1·3	1·8	1·4
III	U.T.S., ton/in.^{2}	30·5	30·1	29·7	29·0	28·3
	Loss in strength, ton/in.^{2}	0·7	1·1	1·5	2·2	2·9
IV	U.T.S., ton/in.^{2}	31·2	31·3	31·5	30·2	30·0
	Loss in strength, ton/in.^{2}	0	−0·1	−0·3	1·0	1·2
V	U.T.S., ton/in.^{2}	30·4	30·7	30·4	30·7	30·3 (6) *
	Loss in strength, ton/in.^{2}	0·8	0·5	0·8	0·5	0·9
VI	U.T.S., ton/in.^{2}	29·9	30·0	29·3	29·5	Specimens lost
	Loss in strength, ton/in.^{2}	1·3	1·2	1·9	1·7	
VII	U.T.S., ton/in.^{2}	30·6	30·6	30·3	29·3	30·0
	Loss in strength, ton/in.^{2}	0·6	0·6	0·9	1·9	1·2
VIII	U.T.S., ton/in.^{2}	30·4	29·5	29·5	29·2	30·5
	Loss in strength, ton/in.^{2}	0·8	1·7	1·7	2·0	0·7

* All U.T.S. values, except where indicated, are the average of 9 tensile tests, 3 of which were stressed at design stress, 3 at 0·1% proof stress, and 3 unstressed during exposure to various environments.

The result of a much more severe form of foliation attack than that observed in any of the specimens examined is shown in Fig. 13 (Plate XL), which is a photomicrograph of a section of H15-WP alloy extrusion from part of an aluminium alloy frame that had been used for carrying a brine tank.

The high corrosion-resistance of the specimens exposed to tidal immersion at Christchurch (Site VIII) was surprising in view of the comparatively severe local attack at isolated spots on the specimens totally immersed in the river. This was the case with both alloys.

The results of total-immersion tests off Hayling Island (Site V) were also very satisfactory from the point of view of resistance to corrosion attack. This may again have been due to protection by marine growth which rapidly covered both specimens and frames. It is interesting to note that marine growths appear to protect aluminium rather than to accelerate attack.

At Southampton (Site IX), where only specimens of H10-WP alloy were exposed, there was frequently a deposit of grease on the specimens which came from the oil effluent of passing ships. Such deposits appeared to protect the specimens, and although it is re-assuring to find that marine growths and oil deposits protect rather than accelerate attack, it would be wise to design on the assumption that such protective agents did not operate, since they may not invariably occur in practice.

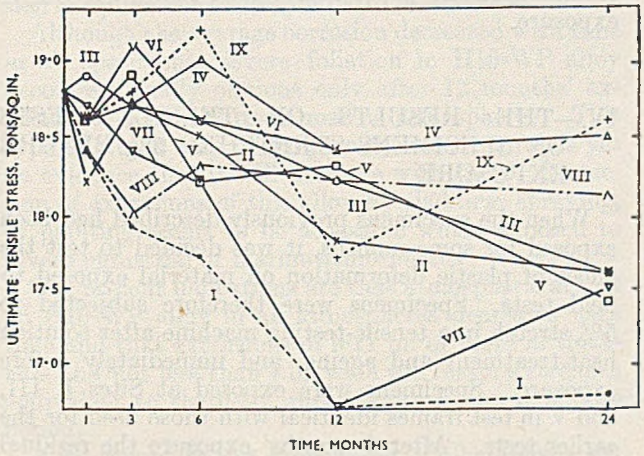


Fig. 14.—Average Tensile Strength of H10-WP Alloy Specimens Exposed at Various Sites.

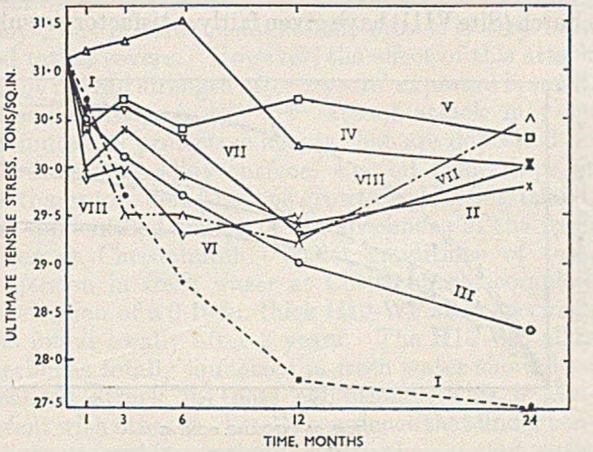


Fig. 15.—Average Tensile Strength of H15-WP Alloy Specimens Exposed at Various Sites.

Key.

● Site I. Atmosphere at Sheffield.
× Site II. Atmosphere at London.
○ Site III. Atmosphere at Hayling Island.
△ Site IV. Atmosphere at Stoke Poges.
□ Site V. Total Sea-Water Immersion at Hayling Island.
∨ Site VI. Tidal Sea-Water Immersion at Hayling Island.
∇ Site VII. Total Fresh-Water Immersion at Christchurch.
∧ Site VIII. Intermittent Fresh-Water Immersion at Christchurch.
+ Site IX. Tidal Sea-Water Immersion at Southampton.

4. Tensile Tests

Perhaps the most striking feature of the results of these tests was the absence of any effect of stress on the rate of corrosion damage on either of the materials. This is particularly surprising in the H15-WP material, which in sheet form is known to be susceptible to stress-corrosion. In general, the residual strength of the unstressed corroded material was similar to that of specimens stressed at the design stress or the 0·1% proof stress. The rate of loss of elongation was also unaffected by stress during corrosion.

The lack of susceptibility of the extruded specimens of H15-WP alloy to stress-corrosion is almost certainly due to structural inhomogeneities in this material, which give rise to planes of preferential corrosive attack parallel to the surfaces of the extrusion. Any corrosion pits and fissures which start from the surface fail to exert any stress-raising effect, since they are turned parallel to the surface and to the direction of stressing along these planes of inhomogeneity.

In practice, it is difficult to stress an extrusion in any direction which is not parallel to some of the planes of these inhomogeneities, so that in most cases these observations have an important practical significance in considering stress-corrosion susceptibility. However, when extrusions are machined it may be possible to stress normal to these planes, in which case rapid failure as a result of stress and corrosion may occur. This aspect is being further investigated in the laboratory, together with the effects of heat-treatment.

Since it was clear from the results of these tests that stress had a negligible effect on corrosion damage, as judged by the residual strength of the corroded specimens, it was decided to determine the average residual strength of stressed and unstressed specimens and to use this average value (i.e. the average of 9 tests) when considering the corrosion behaviour of the material after different periods of exposure. These average values are summarized in Tables I and II, and are expressed graphically in Figs. 14 and 15. From these graphs it is clear that:

(i) The Sheffield atmosphere (Site I) caused the greatest deterioration in the properties of both alloys.

(ii) The percentage loss of strength is approximately the same for both alloys, i.e. about 12%, although, as shown graphically, the loss in tensile strength of H15-WP is usually greater than that of H10-WP.

(iii) In both alloys the atmospheric-exposure sites seem to have the greatest adverse effect apart from the rural site at the Fulmer Research Institute (Site IV) where the attack during 2 years had negligible effect on the mechanical properties.

(iv) The results of total-immersion tests at Hayling Island (Site V) show very little loss of strength, but at Christchurch (Site VII) the scatter is appreciable, owing apparently to more severe local attack.

(v) It seems that in the tidal-immersion tests at Southampton Docks (Site IX) a certain amount of protection was afforded by the oil effluent from ships.

(vi) The intermittent immersion tests at Christchurch (Site VIII) have given fairly satisfactory results, which are appreciably superior to those of specimens totally immersed. There was a deposit of earthy material on these specimens which had been placed at approximately half-tide level under a concrete pier. Normally such deposits tend to promote attack on metals, but it seems that in this particular case, the deposit was protective, and it is possible that some other organic material, probably from sewage in the river, was also present in the deposit.

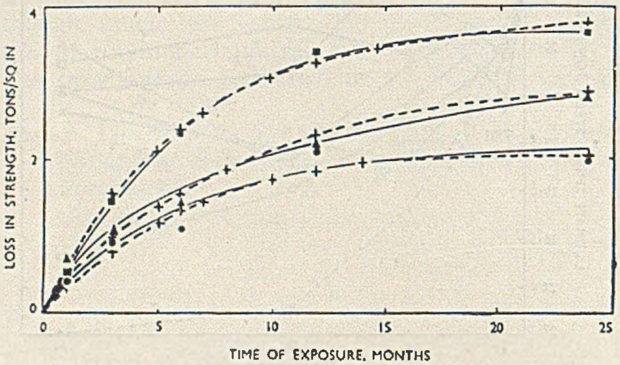


Fig. 16.—Variation of Loss in Strength with Increasing Exposure Period.

Key.

--- + --- Calculated Curve.
—— ■ —— H15-WP Alloy at Sheffield (Site I).
—— ▲ —— H15-WP Alloy at Hayling Island (Site III).
—— ● —— H10-WP Alloy at Sheffield (Site I).

The loss in strength was plotted against the period of exposure for all sites and examined in the light of Champion's views [6] regarding the decrease in rate of corrosion of aluminium alloys with time. It is possible to draw a curve of the type suggested by Champion to fit the experimental curves drawn through the average of 9 results for H15-WP and H10-WP alloy specimens exposed at Sheffield (Site I) and H15-WP at Hayling Island (Site III) (Fig. 16). There was, however, considerable scatter of the results of individual specimens (one specimen showing six times as much damage as another). The maximum damage of the H15-WP specimens was associated with severe local foliation which did not start until after 12 months' exposure and appeared to indicate a new type of attack, the incidence of which was much delayed and appeared to be progressing rapidly. In view of this it is not considered that the exponential type of curve can be applied directly to the design of structures exposed to atmospheric corrosion. From the test results on both materials from the remaining sites, smooth curves could not be drawn owing to scatter in the results, even taking average figures, but the corrosion damage was small and showed a tendency to occur at a decreasing rate with time. The scatter in the original ultimate tensile strength of uncorroded H15-WP alloy, in particular, was about ±2·5 tons/in.^{2}, which would tend to mask the corrosion effect, particularly after short periods of exposure.

IV.—THE RESULTS OF TENSILE TESTS ON SPECIMENS STRETCHED 5% BEFORE EXPOSURE

When the specimens previously described had been exposed for some months, it was decided to test the effect of plastic deformation on material exposed to field tests. Specimens were therefore subjected to 5% stretch in a tensile-testing machine after solution heat-treatment and ageing, and immediately before exposure. Specimens were exposed at Sites I, III, and V in test frames identical with those used for the earlier tests. After 6 months' exposure the residual strength of both materials was quite high and superior to that of specimens that had not been stretched. The corrosion attack did not appear to be as severe. After 12 months' exposure, however, corrosion attack appeared to have increased in severity, and after 18 months the attack definitely appeared to be worse than that on the unstretched specimens.

The results of microscopical examination clearly indicated the increased severity of attack, particularly on the H10-WP alloy specimens which at Sheffield showed general intercrystalline corrosion attack occurring to a maximum depth of 0·022 in. after only 18 months' exposure. There is some indication also that specimens stressed at 0·1% proof stress are more heavily attacked than those stressed at the design stress. In the case of H15-WP specimens, however, the effect of cold work on the corrosion-resistance is not nearly so pronounced.

A similar effect is revealed by the results of tensile tests, from which there is a clear indication of a loss of approximately 2 tons/in.^{2} in the last 12 months' exposure in the H10-WP specimens at Sheffield, whereas the loss in strength of H15-WP alloy exposed at Sheffield during the same period seems to be rather less than this. There was no evidence of rapid stress-corrosion failure of any of these specimens that had been given 5% stretch before exposure.

V.—COMMENTS AND CONCLUSIONS

The results of the field tests have shown that the most severe attack on H10-WP and H15-WP alloys occurs at Sheffield (Site I), where there is an average loss of strength of 2·0 and 3·7 tons/in.^{2}, respectively, after 2 years' exposure, representing a loss on original strength of 10·7 and 11·8%, respectively, or approximately 0·012 in. thickness. These values are greater than those reported in the A.S.T.M.[8] tests at Altoona and New York City on 51S alloy after 3 years' exposure.

The maximum corrosion is much greater in the H15-WP specimens, being equivalent to the loss of approximately 33% or 0·033 in. thickness. This figure is based on one test specimen exposed for 2 years at Sheffield and is associated with foliation-type attack. In H10-WP specimens the maximum corrosion is equivalent to 15% or 0·015 in. thickness.

Although the average corrosion decreased with time in all specimens, severe foliation in H15-WP alloy becomes visually obvious only after 12 months' exposure to Sheffield or to marine atmospheres. This type of attack may be dangerous, since there is as yet no evidence that it will decrease with time. Protection of extrusions of this alloy by cladding, spraying, or painting seems to be advisable where exposed to severe industrial or marine atmospheres.

The present indications are that atmospheric attack on H10-WP alloy in all atmospheres decreases with time and that extrusions of 0·11 in. or more in thickness and not less than ¼ in. wide will retain a strength equivalent to the 0·1% proof stress for an indefinite period.

With regard to the application of Champion's exponential equation to the corrosion of these alloys, it seems that this might be fitted fairly well to the average corrosion/time curves in the severe industrial atmosphere, but that there is less agreement in marine and semi-industrial atmospheres, where the corrosion attack is comparatively small and may possibly be masked by the scatter in the original properties of the materials. Considering, however, the worst case, i.e. the Sheffield atmosphere, calculation on the basis of Champion's equations shows that the minimum thickness required for both H10-WP and H15-WP alloys to ensure that the residual strength of the material will never fall below the original 0·1% proof stress, is of the order of 0·058 and 0·055 in., respectively. These values are of the same order as those quoted by Champion [7] in his calculations of the A.S.T.M. results on 17S-T.

At the same time it would be unwise to base any design directly on such a calculation for average loss. In many cases, failure of a structure may be caused by severe local attack. In other cases local penetration of a member with a large area may have no significant influence on strength. The actual test-pieces cut from the specimens after exposure to corrosion were only ¼ in. wide in the 2 in. gauge-length. One such specimen of H15-WP alloy that had a residual strength of 20·3 tons/in^2. had lost strength to the extent of 11·2 tons/in.^{2}, which is equivalent to three times the average (3·7 tons/in.^{2}) for the 9 specimens exposed for 2 years at Sheffield. Hence it is possible that an H15-WP alloy extrusion, 0·10 in. thick, ¼ in. wide, and 18 in. long, may lose strength to such an extent that failure would occur when stressed to only five-sixths of the 0·1% proof stress after 2 years. In such case a factor of safety of 2 on the average corrosion rate would obviously be inadequate in calculating the thickness required for infinite life. The position with H10-WP alloy specimens exposed at Sheffield is not so serious, since the maximum loss in strength is only 15% higher than the average, and a factor of safety of 2 might be adequate for this alloy under these conditions.

The remaining sites at which there is total and intermittent immersion of the specimens in sea-water and fresh water, cause comparatively little attack on both alloys, but in some instances the attack is very local and rather severe. However, the effect of this attack on the overall strength after 2 years' exposure is small. In some instances the low rate of attack may be attributed to protective agents that are deposited on the aluminium-alloy surface, e.g. oil from ships at Southampton, dense marine growth at Hayling Island, and unidentified soiling of the specimens in the fresh water at Christchurch. Under conditions of total immersion in fresh water at Christchurch, complete penetration of a 0·10 in.-thick H10-WP alloy specimen was found locally after 2 years. The H15-WP alloy specimens totally immersed in fresh water showed no localized attack in these conditions. This is consistent with other workers' [9] experience that under conditions of total immersion in relatively stagnant water the local pitting is much deeper and less widespread in alloys of higher purity. Apart from the localized pitting on total immersion in fresh water, visual examination of the specimens shows a marked superiority of H10-WP alloy over H15-WP alloy, particularly during the second year of exposure.

The corrosion attack of H10-WP alloy is usually of the intercrystalline type, and the effect of cold working the material by stretching is to increase the rate of attack. Although this causes no marked deterioration in properties compared with those of unstretched material after 18 months' exposure, it is expected that the higher rate of attack will result in the cold-worked material having a lower residual strength than material that has not been cold worked after a certain minimum period of exposure which may be of the order of 2 years at the Sheffield site. The effect of stretching the material before exposure seems to have very little effect on accelerating the corrosion attack of H15-WP alloy.

In neither of the alloys has there been any evidence of stress-corrosion occurring during the field tests. This is not unexpected in the case of the H10-WP alloy, in view of its comparatively high general corrosion-resistance. Judging by the results of tests on H15-WP alloy sheet, reported by many other investigators, it was thought that some stress-corrosion failures of the extruded material might occur. This has not been the case, however, and the resistance to this type of failure is attributed to the longitudinal character of the corrosion attack, since a corrosion pit starting at the surface will tend to spread longitudinally rather than penetrate deeper into the material, and so will reduce the risk of high stress concentration.

ACKNOWLEDGEMENTS

This paper is published by kind permission of the Ministry of Supply and of Almin, Ltd., who were the joint sponsors of the work. The author wishes to thank his colleagues at the Fulmer Research Institute for their assistance, and in particular Mr. E. A. G. Liddiard for his personal interest and helpful comments.

The author is also indebted to the following associations, laboratories, and firms, for placing various exposure sites at his disposal: (1) The British Iron and Steel Research Association, and the Brown-Firth Research Laboratories, Sheffield; (2) The British Non-Ferrous Metals Research Association, Euston Street, London; (3) The Admiralty, Central Metallurgical Laboratory, Emsworth, Hants.; (4) The Military Engineering Experimental Establishment, Christchurch, Hants.; and (5) British Railways, Docks Department, Southampton.

APPENDIX I

Results of Visual Examination

Site I—Severe Industrial Atmosphere

Alloy H10-WP.

The specimens soon became coated with a thick deposit of soot and corrosion product. After 6 months' exposure the surface became roughened as a result of pitting, and after 12 months specimens were coated with a firmly adherent mixture of soot and corrosion product. There was little change in appearance after 2 years' exposure, except that the corrosion attack appeared to be rather more severe on the underside of the specimens.

Alloy H15-WP.

The specimens rapidly became coated with soot and a little corrosion product. The surface became roughened after 6 months' exposure, and after 12 months' exposure the attack had appreciably increased. After 2 years fairly severe general attack had occurred, particularly on the underside of the specimens. Unstressed channel-section specimens showed severe general foliation.

Site II—Normal Industrial Atmosphere

Alloy H10-WP.

A light sooty deposit formed on the specimens, but corrosion attack was only slight after 6 months. Twelve months' exposure showed a heavier coating of soot and corrosion product, particularly on the tension side. There appeared to have been slight increase in corrosion attack after 2 years' exposure.

Alloy H15-WP.

The appearance of the specimens was very similar to that of H10-WP, corrosion attack being visible after 12 months and slight general pitting occurring after 2 years' exposure.

Site III—Marine Atmosphere

Alloy H10-WP.

The start of corrosion attack was comparatively slow, very slight general attack being visible after 3 months and developing into very slight general pitting after 6 months. An appreciable amount of sea-salt was mixed with the corrosion product. Little change in the appearance had occurred after 12 months' exposure, and after 2 years the specimens were covered with a thin uniform layer of corrosion product and a deposit of sea-salt. Underneath, however, only slight general pitting of the aluminium had occurred.

Alloy H15-WP.

The rate of attack was slightly greater than on the H10-WP alloy specimens, slight general pitting occurring after 6 months and some severe isolated patches of pitting occurring after 12 months' exposure. There was a fairly heavy deposit of corrosion product and sea-salt on the specimens after 2 years and some fairly deep isolated pits with indications of slight foliation, particularly on the underside of the unstressed specimens.

Site IV—Rural Atmosphere

Alloys H10-WP and H15-WP.

The attack on both materials at this site was very slight, even after 2 years, only slight darkening of the surfaces occurring, accompanied by very slight pitting.

Site V—Total Immersion in Sea-Water

Alloy H10-WP.

There was slight fouling of the specimens, and a few barnacles had attached themselves to the surface after only one month's exposure. After 3 months there was considerable fouling and marine growth on all the specimens, which were stained brown. The fouling increased, and it appeared that where barnacles became attached there was appreciable protection of the material. The corrosion attack of the aluminium, even after 2 years' exposure, was very slight, and it seemed that the marine growth served to protect the aluminium, since it was only in exposed places that very slight pitting had occurred.

Alloy H15-WP.

The behaviour was similar to that of the H10-WP alloy. Fouling occurred rapidly and seemed to afford appreciable protection. Corrosion attack was negligible.

Site VI—Tidal Immersion in Sea-Water

Alloy H10-WP.

The specimens rapidly became coated with a deposit of sea-salt and a little corrosion product; very slight pitting, however, was detected only after 6 months' exposure, and after 12 months little change had occurred.

Alloy H15-WP.

The attack was a little greater than in the H10-WP specimens, slight general attack being visible after 3 months' exposure and slight blistering on the tension surface had occurred after 6 months' exposure. This blistering became more general after 12 months and was possibly the beginning of foliation attack. These tests, however, were discontinued after 12 months, owing to severe damage by heavy seas and loss of specimens.

Site VII—Total Immersion in Fresh Water

Alloy H10-WP.

The specimens quickly became covered with a rust-like colouring which was probably an organic deposit from sewage that is emptied into the river. A deposit of mud and green algæ had formed after 3 months' exposure, but corrosion attack was very slight. Isolated spots of pitting were visible after 12 months' exposure, and this had increased in severity in some places after 2 years.

Alloy H15-WP.

The specimens behaved in a very similar manner to those of H10-WP alloy. They became covered fairly quickly with slight rust and a deposit of green algæ formed on the upper surface. The corrosion attack was very slight, but after 2 years' exposure isolated spots of foliation were observed on two of the specimens. The specimens were frequently covered with a thin film of mud.

Site VIII—Tidal Immersion in Fresh Water

Alloy H10-WP.

The specimens became covered rapidly with a deposit of mud, but corrosion attack was negligible. The mud became more adherent with time, but even after 2 years' exposure negligible corrosion had occurred.

Alloy H15-WP.

The deposition of mud was similar to that on the H10-WP alloy specimens, but fairly general corrosion attack of the tension surface had occurred after 6 months' exposure. This became rather more severe as time proceeded, and slight foliation was detected after 2 years' exposure.

Site IX—Tidal Immersion in Sea-Water

Alloy H10-WP.

There was considerable fouling by barnacles on the unstressed aluminium-alloy frame and specimens, but the deposition of barnacles was much slower on the stressed specimens. The frames and specimens became coated also with a thin film of oil, apparently from passing ships. It is not surprising, therefore, that negligible corrosion of the material occurred.

The specimens that had been stretched 5% before exposure at sites I, III, and V corroded rather more rapidly, but the general appearance was similar to that of specimens that had not been stretched.

APPENDIX II

Method of Assessment of Corrosion Attack

All specimens were cut from the corroded middle portion of the stressed test-pieces and from the unattached ends of the unstressed channel test-pieces. The section examined was parallel to the longitudinal axis and normal to the neutral plane of the test-piece. Examination was carried out at a magnification of 100, and the maximum depth of attack noted on each set of 3 specimens, the total length of the 3 microsections being approximately 1½ in. The diameter of the field of view of the microscope was 0·060 in. approximately, and the grading of the degree of attack was as follows:

(a) *Negligible attack.*—No attack over at least 5 to 6 consecutive fields, i.e. 0·3 in.

(b) *Very slight attack.*—Attack only observed at intervals of 2 to 3 fields, i.e. 0·15 in. approx.

(c) *Slight attack.*—Spots of attack observed in consecutive fields, i.e. 0·050 in.

(d) *Slight general attack.*—Attack of the same depth occurring over 1 or 2 adjacent fields with gaps of 2 fields or more between these regions where no attack occurred.

(e) *General attack.*—Continuous attack of approximately the same depth for several fields.

(f) *Isolated spots.*—This term is self-explanatory and refers to isolated areas where attack to an abnormal extent has occurred.

(g) *Intercrystalline.*—This describes the type of corrosion where the grain boundary is attacked preferentially.

(h) *Longitudinal intercrystalline corrosion.*—Preferential attack along grain boundaries just below and parallel to the surface.

(j) *Pitting.*—Corrosion of the material where no preferential attack occurs.

(k) *Foliation.*—Where corrosion attack penetrates below the surface and then runs approximately parallel to the surface in the direction of the longitudinal plane of the specimen, the attack is termed foliation. Ultimately, formation of corrosion product causes the outer layer of material to lift and peel off ("foliate").

REFERENCES

1. J. M. Bryan, *Rep. Food Invest. Board, Dept. Sci. Indust. Research*, 1938, 217.
2. J. M. Bryan and T. N. Morris, *J. Soc. Chem. Ind.*, 1940, 59, 159.
3. F. A. Champion, *Trans. Faraday Soc.*, 1945, 41, 593.
4. W. H. J. Vernon, *ibid.*, 1927, 23, 150.
5. R. Guilhaudis, *Rev. Mét.*, 1952, 49, (11), 791.
6. F. A. Champion, *Metal Ind.*, 1948, 72, 440, 463.
7. F. A. Champion, *ibid.*, 1949, 74, 7.
8. E. H. Dix, Jr., and R. B. Mears, *A.S.T.M. Symposium on Atmospheric-Exposure Tests on Non-Ferrous Metals*, 1946, 57.
9. Unpublished work by the British Non-Ferrous Metal Research Association and Aluminium Laboratories Ltd., Canada.

1443

THE MEASUREMENT OF THE RELATIVE HARDNESSES OF FINE POWDER PARTICLES *

By J. B. MATTHEWS,† B.Sc., Ph.D., F.R.I.C.

SYNOPSIS

The experimental technique originally devised by Chalmers (*J. Inst. Metals*, 1941, 67, 295) for the determination of the surface hardness of metallic strips, by their bombardment with a hard powder such as sand or carborundum, has been applied to the measurement of the relative hardnesses of different powder particles. The method depends on the decrease in specular reflectivity produced by the indentation of the reflecting surface by the bombarding powder particles, when the hardness of the surface is not far removed from that of the particles. Formal relationships have been derived connecting the decrease in reflectivity with the relative " hardness " of the particles.

I.—INTRODUCTION

The standard methods of measuring the hardness of solids require test-pieces possessing a plane surface with an area of at least 5 mm.^{2}, and therefore they cannot be used to determine the hardness of fine powder particles. Similarly, the available micro-hardness techniques cannot be used for small powder particles. Mindt [1] has described a method for comparing the hardnesses of powder particles, in which the hardness is measured by the amount of abrasion produced by the particles under standardized conditions. This method, however, will not give results with a high degree of precision, and cannot discriminate between powders, the particle hardnesses of which are nearly equal. The method of measurement described in the present paper, on the other hand, can be applied to compare powders consisting of particles with relatively small differences in hardness. It was devised in order to determine the relative hardnesses of commercial grades of aluminium powder, but there is no fundamental reason why its application should not be extended to any powders, the particle hardnesses of which are covered by the range of substances capable of being strain-hardened by indentation.

II.—GENERAL PRINCIPLES OF THE METHOD

The principles on which the new method is based can be illustrated in the following way.

When a hard spherical body is allowed to fall on a softer metallic surface, the volume of the indentation formed in the latter is to a first approximation directly proportional to the kinetic energy lost during impact, and inversely proportional to the resistance to penetration, or hardness, of the surface. The kinetic energy lost in impact is dissipated in various ways, one of which is the production of a state of strain in the metal surrounding the indentation, with a resultant increase in hardness, known as strain-hardening. If an identical body is now dropped from the same height, so that it hits the surface close to the first indentation, the volume of the indentation produced by the second falling body will be less than that produced by the first, because of the strain-hardened condition of the surface due to the first indentation.

If the receiving surface is chosen so that its original hardness is not very much less than that of the falling body, then it is possible that the second impact will not produce an indentation, because the strain-hardening produced by the initial impact may be sufficiently large to render the surface completely resistant to the second falling body. In fact, the second indentation will be equal in volume to the first only if it is produced outside the surface area affected by the strain-hardening resulting from the latter.

It is possible to extend this argument to the case of a stream of powder particles, falling in a uniform manner on a given surface, by treating this case as equivalent to a stream of identical spheres, which is permissible if the powder particles are neither needle-shaped nor flaked and are confined within narrow limits of size. The size of the average indentation produced by a given powder on a given surface can then be calculated from a knowledge of the disappearance of plane surface area as a function of the quantity of powder fallen, and a knowledge of the ratio of the plane-projected area of the average indentation to the corresponding area of plane surface rendered unresponsive to further indentation. Furthermore, from a comparison of the results obtained for different powders, using the same standard surface for each, values can be obtained

---

* Manuscript received 26 August 1952.

† Shell Petroleum Co., Ltd., Thornton Research Centre, Chester.

which give a measure of the relative hardnesses of the powder particles.

The disappearance of plane surface area as a function of the quantity of powder fallen, and the ratio of the plane-projected area of the average indentation to the corresponding area of plane surface rendered unresponsive to further indentation, can both be determined experimentally by measuring the changes in specular reflectivity of the surface caused by the impact of the stream of powder particles.

TABLE I.—*Densities and Analytical Data for the Powders.*

Powder	Density, g./c.c.	Cu, wt.-%	Mg, wt.-%	Si, wt.-%	Mn, wt.-%	Fe, wt.-%	Ni, wt.-%	Zn, wt.-%	Pb, wt.-%	Ti, wt.-%
Aluminium *A* . .	2·799	4·40	0·38	0·72	0·60	0·53	0·03	0·05	0·01	0·01
Aluminium *B* . .	2·795	4·53	0·98	0·70	0·60	0·54	0·03	0·06	0·01	0·01
Aluminium *C* . .	2·801	4·35	0·01	0·67	0·68	0·56	0·02	0·02	Tr.	0·02
Aluminium *D* . .	2·811	4·31	0·27	0·72	0·70	0·57	0·03	0·02	Tr.	0·01
Mg–Al alloy . .	2·147	...	...	...	...	...	...	...	...	...
Sand . .	2·662	...	...	...	...	...	...	...	...	...

III.—EXPERIMENTAL

1. APPARATUS AND TECHNIQUE

The experimental technique used for following the changes in specular reflectivity, caused by the impact of a stream of powder particles, is the same as that described by Chalmers [2] for measuring the hardness of thin surface layers of metals and other materials. The apparatus consists essentially of two parts: a long tube for dropping known quantities of powder vertically on to a standard surface inclined at an angle of 45°, and a means of measuring the specular reflectivity of the surface during the course of the experiment. In the investigation described in the present paper the tube was 100 cm. in length and 0·74 cm. in dia., and the funnel for introducing the powder into the tube had an outlet 2 mm. in dia.

2. MATERIALS USED

The powders used in this investigation were four grades of blown aluminium powder, designated *A*, *B*, *C*, and *D*, a granulated alloy containing equal parts of aluminium and magnesium, and sand. All these powders were carefully sieved, and in each case only the fraction passing through the 120 B.S.S. mesh and retained by the 170 B.S.S. mesh was used. Thus, the limiting diameters of the particles could be taken as being $1{\cdot}27 \times 10^{-2}$ and $0{\cdot}9 \times 10^{-2}$ cm. The densities of the powder particles were determined pycnometrically, and the values obtained, together with analytical data for the aluminium powders, are given in Table I.

Since the softest powders were presumably the aluminium powders, the metal chosen to give the standard surface was commercially pure aluminium (*c.* 98% purity). This was used in sheet form, of thickness $\frac{1}{32}$ in., and strips were given a high polish, using a polishing mop in conjunction with Canning's Peerless Polishing Powder, followed by annealing by heating to 280° C. for a few minutes. Surfaces prepared in this manner were found to be suitable for the purposes of the investigation, but, although there was no noticeable change in reflectivity on annealing and ageing, variations in reflectivity and hardness were apparent from one surface to another. It is desirable to eliminate such sources of error as far as practicable, and it is possible that this could be achieved by preparing the standard surfaces by electropolishing.

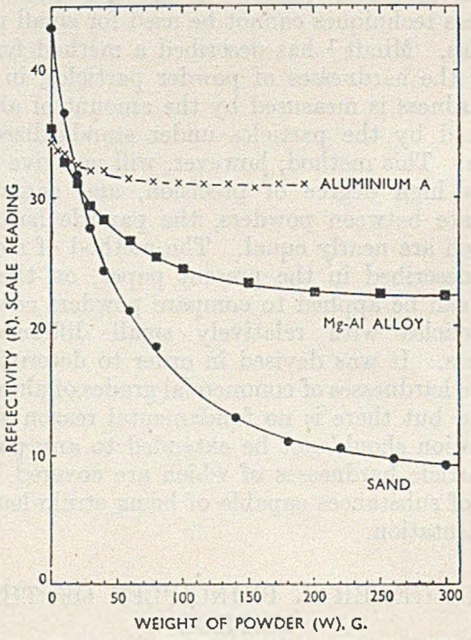


FIG. 1.—Variation of Reflectivity with Weight of Powder Dropped.

3. RESULTS

A typical set of results obtained for the variation of specular reflectivity with the quantity of powder dropped is shown graphically in Fig. 1. Significant data obtained for the whole series of experiments are given in Table II, in which R_0 is the initial reflectivity of the surface and R_∞ the final reflectivity, expressed in units of the galvanometer scale reading. Since the reflectivity does not in practice reach a limiting value—the final portions of the graphs shown in Fig. 1 consisting of straight lines of finite slope—the value of R_∞ has been arbitrarily fixed as the value corresponding to the point at which the final linear portion of the curve is tangential to the initial curved portion.

TABLE II.—*Significant Reflectivity Values.*

Powder	R_0	R_∞	$(R_0 - R_\infty)$	$\frac{R_0 - R_\infty}{R_0} \times 100$
Aluminium *A*	39·2, 40·3, 34·4, 32·5	35·0, 35·3, 31·3, 29·25	4·2, 5·0, 3·1, 3·25	10·7, 12·4, 9·0, 10·0
Aluminium *B*	31·15, 34·8	27·6, 31·15	3·55, 3·65	11·4, 10·5
Aluminium *C*	35·1, 34·65	29·5, 30·75	5·6, 3·9	15·7, 11·2
Aluminium *D*	38·4, 39·5	33·8, 34·6	4·6, 4·9	12·0, 12·4
Mg–Al alloy	35·5, 36·5	23·0, 25·5	12·5, 11·0	35·2, 30·2
Sand	43·3, 43·5, 35·0	9·0, 8·5, 7·5	34·3, 35·0, 27·5	79·1, 80·4, 78·6

IV.—INTERPRETATION OF RESULTS

In order to arrive at a formal relationship between the reflectivity data and the relative hardnesses of the powder particles, it is necessary to obtain information on, firstly, the relative average kinetic energies possessed by the particles just before hitting the standard surface (referred to subsequently as impact energies), and, secondly, the relationship between the energy of impact and the plane-projected area of the average indentation.

1. The Relative Impact Energies of the Powder Particles

In view of the very large number of powder particles used to effect measurable changes in reflectivity, it is permissible to consider the powders as consisting of equal-sized spheres of 0·0109 cm. dia. (the average of the diameters of particles retained between B.S.S. Mesh Nos. 120 and 170). Now, the terminal velocity attained by a sphere falling freely through air is the velocity at which the frictional drag force is equal to mg, where m is the mass of the sphere and g is the acceleration due to gravity. The frictional drag force is some function of the velocity of the sphere, its radius, and the density and viscosity of air. The precise relationship is not known except over limited ranges of the variables, but it is possible to calculate the terminal velocities for the various powder particles, by the method of successive approximations, from the known empirical relationship between the Reynolds' number and the coefficient of drag for spheres.[3] For the lightest and heaviest particles used (of density $\rho_s = 2{\cdot}147$ and $2{\cdot}811$ g./c.c.) the terminal velocities falling through air at 20° C. ($\rho_a = 1{\cdot}205 \times 10^{-3}$ g./c.c., and $\nu = 0{\cdot}150$ cm.^{2}/sec.) are found to be approximately 70 and 85 cm./sec., respectively. (ρ_a is the density of air and ν the kinematic viscosity of air.)

From the known experimental relationship[3] between the Reynolds' number and the coefficient of drag, it can be assumed that, over the small range limited by the two terminal velocities given above, a linear relationship holds between the logarithms of these two parameters. With this assumption, and taking the frictional drag force corresponding to the terminal velocity to be equal to the weight of the particle, it may be shown that the limiting velocity of the particle (U_∞) is related to its density (ρ_s) by the expression :

$$U_\infty \propto \rho_s^{0{\cdot}79} \quad . \quad . \quad . \quad . \quad (1)$$

These limiting velocities are lower than those which would be attained if Stokes' law were valid. Assuming the validity of Stokes' law, it is possible to calculate the terminal velocity, and also to obtain a close estimate of the distance which the particle must fall in order to attain it. In the case of the lightest and heaviest particles used, the Stokes' law terminal velocities become 76·9 and 100·7 cm./sec., respectively, and the corresponding distances fallen in order to reach 99% of these velocities are 22 and 37 cm.

Therefore, it may be safely assumed that in falling through the dropping tube, of length 100 cm., the velocities attained by the particles, which are their velocities just before impact, are their terminal velocities calculated from the relationship between their Reynolds' numbers and coefficients of drag.

2. The Relationship Between Impact Energy and Size of the Indentation

Detailed investigations have been carried out by Martel[4, 5] into the conditions governing the size of the indentation produced in plane metallic surfaces by the impact of a rebounding hammer. Martel, using cone-shaped indenters, found that the ratio of the energy of impact to the volume of the indentation produced is constant for any given metal. Edwards and Willis,[6] on the other hand, using spherical indenters, deduced the following relationship between the radius of the plane-projected circular area of the indentation (r) and the energy of impact (E) :

$$\frac{r}{4\sqrt{E}} = \text{constant} \quad . \quad . \quad . \quad (2)$$

In a subsequent paper, Edwards[7] discusses both Martel's and his own work, and maintains the view that, generally speaking, his own experiments with different hemispheres indicate that the diameters of the indentations agree very closely with equation (2). But he admits that this expression is not absolutely accurate, and needs an additional correction factor before it truly represents the resistance to penetration under impact. On the most simple picture of the physical process of indentation it can be assumed that the volume of the indentation is proportional to the energy absorbed during impact, but it does not immediately follow that the diameter of the indentation also obeys the same proportionality. If the final shape of the indentation is not spherical, and the divergence from the sphericity varies with the depth of penetration, then the ratios of volume and diameter of the indentation respectively to the energy of impact will not correspond, since a third factor is involved. Accepting the view that there is a fundamental difference between the Martel ratio and the Edwards and Willis ratio, it follows that in practice the choice of which to apply depends on the relationship being sought. In the present investigation, as will be seen later, the requirement is the relationship between the plane-projected area of the indentation and the energy of impact. Hence, the Edwards and Willis ratio will be adopted.

3. The Relationship Between Reflectivity Data and Relative Hardnesses of the Powder Particles

A parallel beam of radiation incident on a plane reflecting surface will be reflected as a parallel beam, the angles of incidence and reflection being equal. If, however, the plane reflecting surface is indented by a sphere, all radiation incident within the area of the indentation, apart from the central ray, will be scattered from the main direction of the reflected beam. Therefore, if the intensity of the reflected beam is measured at a point which can be considered to be at infinity, the decrease in reflectivity produced by spherical indentations on the plane surface will be proportional to their total plane-projected area. Furthermore, the following conditions apply:

(*a*) The powder particles fall uniformly over the surface area subjected to impact.

(*b*) Subsequent impacts to the first do not affect the area of the indentation produced by the first impact; this has been shown to be the case experimentally by Edwards and Willis.[6]

(*c*) An indentation in the surface produces strain-hardening, not only below the indentation, but also for a certain distance along the plane surface outside its perimeter. Thus, Moore [8] finds that for good reproducibility in the ball indentation test the indentations must not be formed at distances closer than five times their diameters.

From (*b*) and (*c*) it may be inferred that appreciable surface strain-hardening occurs in the immediate vicinity of an indentation, and it is probably true to say that the degree of this hardening falls off with distance from the perimeter of the indentation. If no hardening occurred, either in or around the indentation, the result of continual bombardment by the particles would be a continuous depression of the surface, the average appearance of which would roughly correspond to a set of close-packed equal circular areas of indentation. On this basis the final reflectivity value of the surface would be the same in all cases and would be a fraction, equal to $\pi/2\sqrt{3}$, of the initial reflectivity.

Experimentally, it has been found (see Fig. 1) that the final reflectivities obtained cover a wide range of values, being greater for the softer powders. Hence, there is apparently an area surrounding each indentation which is unresponsive to further impact by the powder, and the softer the powder particles the greater is this area. Photomicrographical examination of the bombarded surfaces confirmed this conclusion.

Suppose the radius of the average spherical indentation is r, and the distance of closest approach of the perimeters of the circular areas of indentation in the limiting surface structure is $2D$. Then, from geometrical considerations, the ratio of the total plane-projected area of indentation to the original total surface area, in the limiting case, is given by:

$$\left\{\frac{\text{Indented area}}{\text{Total area}}\right\}_{\infty} = \frac{\pi r^2}{2\sqrt{3}\,(r+D)^2} = 0{\cdot}9067\,\frac{r^2}{(r+D)^2} \quad . \quad (3)$$

Therefore, since the specular reflectivity is proportional to the plane surface area:

$$\frac{R_0 - R_\infty}{R_0} = 0{\cdot}9067\,\frac{r^2}{(r+D)^2} \quad . \quad . \quad (4)$$

in which R_0 and R_∞ are the initial and final reflectivities, respectively.

From this is obtained the expression:

$$\frac{D}{r} = \sqrt{\frac{0{\cdot}9067\,R_0}{R_0 - R_\infty}} - 1 \quad . \quad . \quad (5)$$

Again, it follows on the premises (*a*), (*b*), and (*c*) that:

$$\frac{d(x+y)}{dW} = k\{A - (x+y)\} \quad . \quad . \quad (6)$$

in which k is a constant, x is the total plane-projected area of indentation, y is the total strain-hardened surface area unavailable for further indentation, when the total weight of dropped powder is W, and A is the initial area subjected to impact.

From equation (6) the following expression is derived:

$$\log_{10}\left(1 - \frac{x+y}{A}\right) = -\frac{kW}{2{\cdot}303} \quad . \quad (7)$$

Now, x can be put equal to $A(R_0 - R_W)/R_0$ and also, from considerations of the limiting case, it follows that:

$$\frac{x}{x+y} = \left\{\frac{\text{Indented area}}{\text{Total area}}\right\}_{\infty} = \frac{(R_0 - R_\infty)A}{R_0 A} = \frac{R_0 - R_\infty}{R_0} \quad . \quad (8)$$

Hence,

$$(x+y) = \frac{R_0\left\{\dfrac{(R_0 - R_W)A}{R_0}\right\}}{R_0 - R_\infty} = A \times \frac{R_0 - R_W}{R_0 - R_\infty}. \quad (9)$$

Substituting in equation (7) :

$$\log_{10}\left(\frac{R_W - R_\infty}{R_0 - R_\infty}\right) = -\frac{kW}{2\cdot303} \quad (10)$$

According to equation (10) a straight line should be obtained when $\log_{10}\left(\frac{R_W - R_\infty}{R_0 - R_\infty}\right)$ is plotted against W, the gradient of which is equal to $-k/2\cdot303$. This is confirmed by the experimental data plotted in Fig. 2.

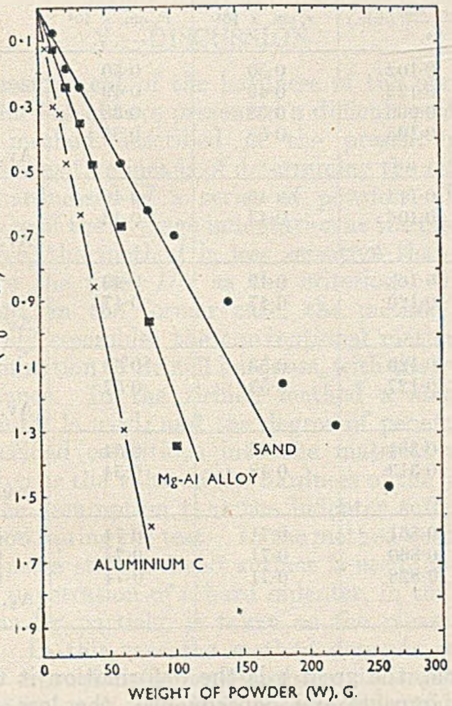


Fig. 2.—Logarithmic Relationship Between Decrease of Reflectivity and Weight of Powder Dropped.

Imagine that a quantity, W, of powder has fallen, rendering an area $(x + y)$ of the surface unresponsive to further impact, and at this stage suppose an additional small quantity, dW, of powder is allowed to fall. Then, the actual amount of powder in this small quantity dW which is effective in producing further indentation is given by :

$$\frac{dw}{dW} = \frac{A - (x+y)}{A} \quad . \quad . \quad . \quad (11)$$

in which dw is the weight of powder effective in producing indentation. It follows that the actual weight of powder which has been effective in producing indentation, when the total weight of powder dropped is W, is given by :

$$w = \int_W^0 dW\left(1 - \frac{x+y}{A}\right) \quad . \quad . \quad (12)$$

Substituting for dW from equation (6) :

$$w = -\frac{1}{kA}\int_A^0 d(x+y) \quad . \quad . \quad (13)$$

Therefore, if $w_{0\cdot5}$ is the weight of powder which has been effective in producing indentations when half the total change in the surface has been produced, then :

$$w_{0\cdot5} = -\frac{1}{2k} \quad . \quad . \quad . \quad . \quad (14)$$

Also, from (10), the total weight of powder ($W_{0\cdot5}$) which has been dropped in producing the half change is given by :

$$W_{0\cdot5} = -\frac{2\cdot303\log_{10}0\cdot5}{k} \quad . \quad . \quad (15)$$

In the surface which has undergone half of the total change the number of indentations, $N_{0\cdot5}$, is given by :

$$N_{0\cdot5} = \frac{w_{0\cdot5}}{\frac{1}{6}\pi(d)^3\rho_s} \quad . \quad . \quad . \quad (16)$$

in which (d) is the average particle diameter and ρ_s the density of the average particle.

Therefore, the ratio of indented area to original total area at the stage of half-change is given by :

$$\left\{\frac{\text{Indented area}}{\text{Total area}}\right\}_{0\cdot5} = \frac{x}{2(x+y)} = \frac{R_0 - R_\infty}{2R_0}$$
$$= \frac{6w_{0\cdot5}\pi r^2}{\pi(d)^3\rho_s A} \quad . \quad (17)$$

from which is obtained the following expression for r :

$$r = \sqrt{\frac{\rho_s(d)^3(R_0 - R_\infty)A}{12w_{0\cdot5}R_0}} \quad . \quad . \quad (18)$$

Equation (18) can be put in the form :

$$r = \sqrt{\frac{k\rho_s(d)^3(R_0 - R_\infty)A}{6R_0}} \quad . \quad . \quad (19)$$

In columns 2 and 3 of Table III the observed values for k and $W_{0\cdot5}$ are given, and in column 4 the values of $W_{0\cdot5}$ calculated from equation (15). It will be seen that there is good agreement between the observed and calculated values for $W_{0\cdot5}$ in the case of the aluminium powders, but that in the case of the magnesium–aluminium alloy powder the observed value is slightly less than the theoretical value, and in the case of sand the discrepancy is more pronounced. In other words, in the case of the harder powders, there is a departure from linearity in the relationship between $\log_{10}\{(R_W - R_\infty)/(R_0 - R_\infty)\}$ and W in the region of high values of W, which may be due to the number of simplifying assumptions which have been made. In this connection it should be stressed that the value of R_∞ used in evaluating equations (5) and (19) is that calculated from the slope of the linear portion of the curve relating $\log_{10}\{(R_W - R_\infty)/(R_0 - R_\infty)\}$ and W, and is not the arbitrarily chosen experimental value.

It is now necessary to assign some value to A. This value could be determined from direct measurement of the elliptical area subjected to impact. But, since the perimeter of this area is not clearly defined, it is more convenient to put A equal to the area of the ellipse formed by the intersection of a plane cutting at 45° to its axis a cylinder of diameter equal to the internal diameter of the dropping tube. Accordingly, A was found to be 0·608 cm.².

Table III.—*Data Derived from Experimental Observations.*

Powder	k, g.$^{-1}$ × 10^2	$W_{0\cdot5}$, g. (obs.)	$W_{0\cdot5}$, g. (calc.)	$\frac{R_0 - R_\infty}{R_0}$ (calc.)	r, cm. × 10^4	r', cm. × 10^4	$\frac{D}{r}$
Aluminium *A*	6·63	11·0	10·5	0·102	0·50	0·50	1·98
	5·34	13·5	13·0	0·122	0·49	0·49	1·73
	4·42	16·2	15·7	0·090	0·38	0·38	2·17
	7·64	8·5	9·1	0·105	0·55	0·55	1·94
							(Av. = 1·955)
Aluminium *B*	5·11	13·8	13·6	0·114	0·46	0·46	1·82
	4·68	14·1	14·8	0·106	0·43	0·43	1·93
							(Av. = 1·875)
Aluminium *C*	3·91	17·4	17·7	0·163	0·49	0·49	1·36
	4·97	12·7	13·9	0·120	0·47	0·47	1·75
							(Av. = 1·555)
Aluminium *D*	6·27	11·0	11·1	0·120	0·53	0·53	1·75
	6·91	10·1	10·0	0·127	0·57	0·57	1·67
							(Av. = 1·710)
Mg–Al alloy	2·03	33·0	34·2	0·361	0·46	0·54	0·58
	2·58	25·5	26·9	0·313	0·48	0·54	0·70
							(Av. = 0·64)
Sand	1·68	35·0	41·2	0·861	0·71	0·74	0·026
	1·68	35·5	41·2	0·860	0·71	0·74	0·027
	1·73	36·5	40·1	0·828	0·71	0·74	0·047
							(Av. = 0·033)

All the data are now known from which to calculate r, using equation (19). The values obtained are given in column 6 of Table III. They are of the same order of magnitude as the values obtained by Chalmers [2] for sand upon tin from direct microscopic and optical measurements, allowing for the difference in size of particles. Now, we know that the area of the indentation is a function of the kinetic energy of the particle just before impact (equation (2)) and also that this kinetic energy of the particle is a function of its density (equation (1)). Therefore, in order to make a true comparison between the powders, it is necessary to introduce a correction factor based on equations (1) and (2). If r' is the radius of the indentation corresponding to an energy of impact equal in all cases to the actual experimental energy of impact of, say, the average aluminium powder particle, then:

$$r' = r\left\{\frac{2\cdot799}{\rho_s}\right\}^{0\cdot645} \quad . \quad . \quad . \quad (20)$$

where 2·799 (g./c.c.) is the density of aluminium A powder and ρ_s is the density of the powder in question.

Values obtained for r' are given in column 7 of Table III. From an examination of them it is obvious that the same degree of penetration is not achieved in all cases, the hardest powder, i.e. sand, penetrating further than the softer powders. The variation of the individual results, unfortunately, does not allow us to draw any definite conclusions with respect to the softer powders, but it appears that the magnesium–aluminium alloy produces a larger indentation than the aluminium powders, excluding aluminium powder D which apparently gives an equal indentation. Such results can be attributed to the mutual deformation of the particle and plane surface on impact. The softer the powder particle, the greater is the deformation it undergoes upon impact, and consequently the less energy is there available for deformation of the plane surface. When the hardness of the surface is such that no permanent deformation is produced in it on impact, then obviously the energy of impact is dissipated entirely in deforming the particle alone. This phenomenon, however, was not apparent on microscopic examination of the powder particles, because the amount of deformation is only a small fraction of the volume of the particle, the radius of the indentation formed by sand particles being only about one hundredth of the radius of the particle. It follows that the observed variation in r' provides a measure of the hardness of the powders, but the sensitivity of this comparison is low, and only large differences in hardness are detectable.

A much more sensitive measure of the relative hardnesses of the powder particles is given by the ratio D/r, in which D, it will be remembered, is half the width of the annulus surrounding the indentation which is strain-hardened to a sufficient degree to resist deformation by the powder particle. It is probably true to say that this ratio is constant in the case of each powder over the whole of the observed range of values of r, and consequently it is not necessary to introduce any correction factor for the experimental differences in impact energy. Accepting the ratio D/r as a measure of the hardness of the powder particles, we see immediately that there is a marked difference in particle hardness between sand, magnesium–aluminium alloy powder, and the aluminium powders. The experimental errors forbid any definite conclusions to be drawn concerning the relative hardnesses of the four aluminium powders employed, but it would appear that *A* and *B* are softer than *C* and *D*.

V.—DISCUSSION

The measurement of the hardness of the particles in a powder has always presented a difficult problem, but the method described in the present paper provides a sensitive means of determining the relative particle hardnesses of a series of powders. If we take the size of the average indentation as the measure of hardness, the method is less sensitive than it is if we take the ratio D/r as the criterion. On the other hand, in the former case, the method then more closely resembles the conventional method for the determination of Brinell hardness, with the following difference. In the Brinell method a standard hard indenter is used, and the degree of penetration under specified conditions into the material under test is taken as the value of the hardness of the latter, making the assumption that the indenter suffers no deformation during the test. In the method described in this paper a standard soft surface is used, and the degree of penetration of a hard indenter, in the form of the powder particle, is taken as the measure of hardness. In this case the method depends on the fact that mutual deformation of the indenter and the surface occurs, and it is necessary in order to obtain any measurable results that their respective hardnesses should not be very different. Using the alternative criterion, the ratio D/r provides a more sensitive comparison between powders, and the method now utilizes the work-hardening which occurs when the standard soft surface is indented, in addition to the mutual deformation of the indenter and surface. The hardness of the particle is related to the ratio D/r in such a way that the harder the particle is, the smaller this ratio becomes.

The range of particle hardnesses which can be compared by this method is determined by the hardness of the standard surface. By choosing an appropriate series of standard surfaces, however, it should be possible to extend the range of particle hardnesses covered. Moreover, by suitable standardization it may be possible to establish a scale of hardness. This, unfortunately, would be an additional hardness scale with which to contend, but perhaps it would be possible to correlate it with existing hardness scales.

Because of the small diameters of the indentations, which are of the order of magnitude of the wavelength of light, diffraction phenomena will become more important as the indentation of the surface proceeds. A rigorous treatment of the problems which then arise would be difficult. However, if the indentations are considered to be completely light-absorbing areas, viewed along the direction which a specularly reflected beam from the non-indented surface would take, and the non-indented areas as constituting a diffraction mirror, the simple deduction follows that practically the whole intensity of the diffracted light is concentrated in the direction of the specularly reflected beam. This leads to approximately the same result as the assumption made in this paper.

ACKNOWLEDGEMENT

This work was carried out during the war years at the Armaments Research Department, Fort Halstead, Kent, and the author wishes to thank the Chief Scientist, Ministry of Supply, for permission to publish the results.

REFERENCES

1. W. Mindt, *Physikal. Ber.*, 1923, **4**, 1092.
2. B. Chalmers, *J. Inst. Metals*, 1941, **67**, 295.
3. S. Goldstein, “ Modern Methods in Fluid Dynamics ”. Oxford : **1938** (Clarendon Press).
4. R. Martel, “ Impact Testing of Hardness ”, *Commission des Méthodes d'Essai des Matériaux de Construction, Paris*, **1895**, Vol. 3, Sec. A. (Métaux), p. 261.
5. D. Tabor, “ The Hardness of Metals ”, Chapter 8 . Oxford : **1951** (Oxford University Press).
6. C. A. Edwards and F. W. Willis, *Proc. Inst. Mech. Eng.*, 1918, **82**, 335.
7. C. A. Edwards, *J. Inst. Metals*, 1918, **20**, 61.
8. H. Moore, *Internat. Congr. Test. Mat., Copenhagen*, **1909**, (II), 55, 57, 123, 252; cf. H. O'Neill, “ The Hardness of Metals and Its Measurement ”, p. 252. London : **1934** (Chapman and Hall, Ltd.).

CRYSTAL FRAGMENTATION IN ALUMINIUM DURING CREEP *

By D. McLEAN, B.Sc., MEMBER

(Communication from the National Physical Laboratory.)

SYNOPSIS

Seven specimens of super-pure aluminium having grain-sizes of 1–9¼ grains/mm., were made to creep at 200° C. under loads varying from about ⅓ to ¾ ton/in.^{2}; the extensions produced ranged up to 50%. Observations and measurements were made relating to the sub-crystals formed. These are consistent with a polygonization model for secondary creep and permit a quantitative check of this model to be made. Certain of the observations appear to be inconsistent with the theory of sub-crystal formation advanced by Wilms and Wood (*J. Inst. Metals*, 1948–49, 75, 693; 1951, 79, 159).

I.—INTRODUCTION

SEVERAL metals when made to creep by a few per cent. or more change their structure in a characteristic way, the original crystals fragmenting into sub-crystals. The first evidence for this was given by Jenkins and Mellor [1] in the case of iron. Detailed and explicit evidence for aluminium has been provided by Wood and his associates [2] and support has come from other workers.[3-5] Subsequently, evidence for lead was advanced by Gifkins,[6] for a tin–antimony alloy by Betteridge and Franklin,[7] and for zinc by Cottrell and Aytekin.[8] It thus seems that the phenomenon is a fairly general one.

One explanation is that this fragmentation is the outcome of polygonization. The latter process has been found to occur in certain metals when appropriately strained and subsequently heated,[9] and it is natural to suggest that it also operates when strain and heat are applied simultaneously, as during creep testing. In a recent paper,[10] dealing with aluminium, evidence was presented that under the conditions used, the fragmentation occurring was due to polygonization, which took place during second-stage creep and accounted for the deformation, with the exception of the small amount attributable to grain-boundary movement. A similar suggestion has been made by Greenough and Smith [4] and by Mott.[11] The purpose of the present work was to extend the investigation over a range of grain-sizes and of stresses. Observations regarding slip bands and grain boundaries were also made on the specimens and are recorded elsewhere.[12, 13] All the evidence obtained is consistent with the view expressed above.

An alternative theory of fragmentation, due to Wilms and Wood,[2] appears to envisage that the parent grains split along inner boundaries during deformation and that the sub-crystals so formed then rotate relatively to each other. A direct experimental check of these two theories favours that of polygonization.

II.—EXPERIMENTAL

The seven specimens used were made from super-pure aluminium strip containing 0·002–0·003% iron, 0·003% silicon, and 0·003% copper. Details of preparation and technique have been given previously.[10, 12] Briefly, the specimens were annealed to give grain-sizes of 1, 4½, and 9¼ grains/mm. They were polished, so that micro-examination of the surfaces could be carried out during deformation, and were made to creep at 200° C. under loads ranging from ⅓ to about ¾ ton/in.^{2}. A summary of the experimental work is given in Table I. The tapered specimens bore a stress that necessarily varied along the length of the specimen, and after test they were sectioned at positions corresponding to the stresses indicated in the table.

TABLE I.—*Summary of Experimental Work.*

(a) *Two Parallel-Sided Specimens*

Electropolished in orthophosphoric acid–carbitol solution.

Identification mark :	W	L
Grain-size :	Fine (9¼ grains/mm.)	Coarse (1 grain/mm.)
Stress :	½ ton/in.^{2}	
During extension :	Removed from creep machine periodically for examination of surface and measurement of extension (slip-band and grain-boundary displacements also measured and reported separately).	
Extension continued to fracture at :	50·3%, 901 hr.	23·6%, 1145 hr.
After fracture :	Sections prepared, anodized, and subjected to micro-examination between crossed polaroids. No X-ray examination made.	

---

* Manuscript received 21 June 1952.

† Metallurgy Division, National Physical Laboratory, Teddington.

Table I.—*continued.*

(b) *Five Tapered Specimens*

Electropolished in perchloric acid–acetic acid solution.

Specimen no.:	I	II	III	IV	V
Grain-size:	Medium (4½ grains/mm.)				
Creep time, hr.:	1	5	25	118½	696
Stress range due to taper:	About ⅓–¾ ton/in.²				About ¼–½ ton/in.²
After creep times given above:	Removed from creep machine. Examined and measured. Sectioned transversely at positions corresponding to stresses given below, and electropolished.				
Stress, ton/in.²	Extension, %, at Position of Each Section				
¼	...	...	...	...	1·2
⅓	0·33	0·64	0·92	1·3	3·6
½	0·86	2·1	3·4	5·4 *	17·0 *
¾	3·1	8·4 *	15·0 *	29·5 *	...
The sections were:	X-rayed. Anodized and subjected to micro-examination between crossed polaroids. Etched and subjected to micro-examination for sub-boundaries.				

* The disorientation between individual sub-crystals was large enough to be measured.

III.—OBSERVATIONS AND MEASUREMENTS

1. Detection of Sub-Crystals

The detection of sub-crystals is possible by X-ray examination, by micro-examination in polarized light of an anodized section, or by micro-examination of the deformed surface. On the deformed surface sub-crystals are revealed either by the appearance of surface tilts or by the "white line" made to appear at the junction between regions of different tilt by suitable adjustment of the microscope. These are the white lines first reported by Wilms and Wood.[2]

There is evidence that the junctions between regions of different surface tilt coincide with the sub-crystal junctions as seen by polarized light. For example, Fig. 1 (Plate XLI) shows an area of the deformed surface photographed with oblique lighting to reveal the bands, running approximately from top to bottom of the figure, of different surface tilt. (The horizontal black lines are slip bands.) The same field, after polishing flat and anodizing, is shown in Fig. 2 (Plate XLI), as it appeared between crossed polaroids. It resembles the X-ray micrographs, Fig. 6 (*a*) and (*b*) of Honeycombe's paper,[14] taken after creep deformation. Corresponding sub-boundaries are identified by the same letter in Figs. 1 and 2; they have very similar shapes in both, a fairly certain indication that the same interfaces are delineated. Since Fig. 1 registers surface tilts and Fig. 2 lattice tilts, tilts in the surface and the lattice coincide. No direct evidence was obtained that the boundaries of the sub-crystals detected by X-rays coincided with those observed microscopically. However, as both methods showed sub-crystals in the same specimens (allowance being made for the greater sensitivity of the X-rays), it is reasonable to assume coincidence.

2. Measurement of Sub-Crystal Disorientation and Diameter

On the fine- and coarse-grained (parallel-sided) specimens the sub-crystal diameter and the mean component α of disorientation about the normal to the microscope stage, were measured in the manner previously described[10] on three different sections, as indicated in Table II.

As before, the angle of disorientation proved to be similar for sections just under the surface and for the mid-section parallel to the surface, suggesting that the deformation did not vary much throughout the volume and that no special surface effect intruded seriously. On the transverse section, however, the angle was higher. The former values may be expected to be reduced by the general tendency towards preferred orientation during extension, while the latter should be unaffected; consequently, the angles measured on the transverse section are the true ones. This was not taken into account in the earlier paper, and hence the calculated extension is under-estimated there.

Table II.—*Values for Component α of Disorientation Between Sub-Crystals and for Sub-Crystal Diameter.*

Angle measurements: mean of 100 readings.
Diameter measurements: mean for 400–500 sub-crystals.

Position of Measurement	Fine-Grained Specimen (final extension 50·3%)		Coarse-Grained Specimen (final extension 23·6%)	
	α, degrees	Sub-Crystal Dia., cm. × 10³ *	α, degrees	Sub-Crystal Dia., cm. × 10³ *
Just under surface	5·3	2·73	4·2	2·53
Mid-section parallel to surface	6·5	2·42	3·6	2·81
Transverse section	11·8	2·00	6·2	2·40

* Measured in polarized light after anodizing.

Table III.—*Values of Sub-Crystal Disorientation θ and Diameter, and Dislocation Density ρ.*

Grain-Size of Specimen	Disorientation, θ		Sub-Crystal Dia., cm. × 10³	ρ, No. of Dislocations/cm.² × 10⁻⁹
	Degrees	Radian		
Fine (9½ grains/mm.) (final extension 50·3%)	16·6	0·29	2·28	8·9
Coarse (1 grain/mm.) (final extension 23·6%)	8·7	0·15	2·51	4·2

The mean component α is related to the mean total disorientation θ by the equation:[10]

$$\cos\theta = \cos^2\alpha \quad . \quad . \quad . \quad . \quad (1)$$

The values of θ obtained by means of equation (1) from the α values measured on a transverse section, are given in Table III.

The measured sub-crystal diameters should also be affected in a similar way by extension, but in this case none of the measurements can be considered as unaffected. The best procedure seems to be to obtain from the first two rows of Table II a mean sub-crystal size as measured parallel to the surface, and then to form the mean between the resultant and the figure for the transverse section. The values so obtained are included in Table III.

For the measurements on the medium-grain-sized (tapered) specimens, the sub-crystal diameter as measured on a transverse section must be multiplied by the following empirical factor to obtain a more correct average diameter :

50% extension (fine-grained specimen)	$\text{factor} = \frac{2 \cdot 28}{2 \cdot 00} = 1 \cdot 14$
24% extension (coarse-grained specimen)	$\text{factor} = \frac{2 \cdot 51}{2 \cdot 40} = 1 \cdot 04$

To a sufficiently close approximation this factor is $\left(1 + \frac{\epsilon}{4}\right)$, where ϵ is the fractional elongation.

Measurements in the case of the medium-grained specimens were made only on transverse sections. The values obtained are given in Table IV. As indicated in Table I, in a proportion of the sections the disorientations were too small to be measured by the polarized-light technique, which has a sensitivity of $\frac{1}{2}^\circ$–1°; with one exception, only those sections on which the disorientation could be measured are referred to in Table IV. The actual measured sub-crystal diameter and the diameter corrected by the factor mentioned above, are both given. The creep extensions are included in order to put the other measurements in their correct context.

Table IV.—*Measurements of Sub-Crystal Disorientation, Diameter* d, *and Dislocation Density* ρ *on Medium-Grain-Sized Specimens.*

Angle measurements : mean of 100 readings. Diameter measurements : mean for 350–1000 sub-crystals.

Stress, ton/in.^{2}		$\frac{1}{3}$	$\frac{1}{2}$		$\frac{3}{4}$		
Creep time, hr.		696	118$\frac{1}{2}$	696	5	25	118$\frac{1}{2}$
Total elongation, %		3·65	5·4	17·0	8·4	15·0	29·5
Disorientation,* degrees	a	...	2·6	9·9	2·76	3·49	11·83
	θ	...	3·6	14·0	4·0	4·7	16·6
Diameter † d, cm. × 10^3	uncorrected	7·89 (*b*)	9·73 (*a*) 6·85 (*b*)	3·3 (*a*)	6·00 (*a*) 3·65 (*b*)	3·74 (*a*) 3·58 (*b*)	2·16 (*a*)
	corrected	7·98 (*b*)	9·85 (*a*) 6·95 (*b*)	3·44 (*a*)	6·10 (*a*) 3·72 (*b*)	3·89 (*a*) 3·71 (*b*)	2·21 (*a*)
Density ρ, no. of dislocations/cm.^{2} × 10^{-9}		...	0·44	5·0	0·8	1·8	8·8

* a is the component of disorientation actually measured ; θ is the total disorientation calculated from a, using equation (1).

† (*a*) Measured in polarized light after anodizing. (*b*) Measured using phase-contrast illumination after etching in Wyon and Crussard's reagent.[5]

It will be seen that the sub-crystal diameter apparently decreased with increasing extension under a given load. However, the whole of this apparent change may really have been a question of sensitivity of measurement. After a slight extension the lattice tilt at some sub-boundaries would be too small to detect, but with further extension the tilts would increase and eventually become sufficiently large to be recorded. This view is supported by the fact that in the early stages of creep sub-boundaries existed which were not detected by the polarized-light method (cf. (*a*) and (*b*), Table IV).

Comment on these results is made in the next Section.

3. Summary

During the primary stage of creep, sub-crystals could not be detected. X-ray examination showed disorientation of the crystals, increasing in amount as the extension became greater. The diffuse X-ray spots were eventually replaced by sharp discrete spots, by which time secondary creep had set in. The clusters of sharp spots indicate that the crystals had broken up into a limited number of relatively perfect areas of different orientation. Sub-crystals could be observed with the microscope some time after they had been detected by X-rays. They were first seen near grain boundaries, especially when the grain-size was large (Fig. 4, Plate XLI). They were later found throughout the grains. The difference in orientation between adjacent sub-crystals increased progressively as creep proceeded.

At first no displacement along the sub-crystal boundaries was noted. After one or two hundred hours' creep, sub-boundary slip became perceptible (Fig. 5, Plate XLI, *XX* and *YY*). The lines along which the slip occurred coincided with the white-line network seen when the microscope was stopped down and the field thrown slightly out of focus (Fig. 6, Plate XLI). Eventually migration of the sub-boundaries was observed, and is the cause of the slight multiplicity of the sub-boundaries at *XX* (Fig. 5).

The size of the sub-crystals did not vary appreciably with grain-size (Table II). In coarse-grained specimens the sub-crystals were arranged in bands (Fig. 2, Plate XLI). These bands were less prominent in medium-grained specimens and much less prominent in fine-grained specimens (Fig. 3, Plate XLI). They were most clearly marked in crystals in which one system of slip predominated and the direction of the bands then lay at right angles to the direction of the slip bands. When the bands were in two directions, one band lay parallel to the slip bands. This may be partly attributable to screw dislocations, as suggested previously, but seems also partly due simply to the concentration of dislocations in the prominent slip bands. The evidence for this is that bands of sub-crystals have been observed to follow bent slip bands (Figs. 36 and 37, Plate LXXIX, of ref. 10), whereas if due solely to screw dislocations, they would be parallel to crystallographic plates and therefore straight.

The sub-crystals grew smaller as the stress, and therefore the rate of strain, increased, as observed by Wood and Rachinger.[15] The rate at which they formed increased as the stress increased. At a stress of $\frac{3}{4}$ ton/in.^{2} sub-crystals were detected by X-rays after 5 hr., but were not visible after $118\frac{1}{2}$ hr. at $\frac{1}{3}$ ton/in.^{2}.

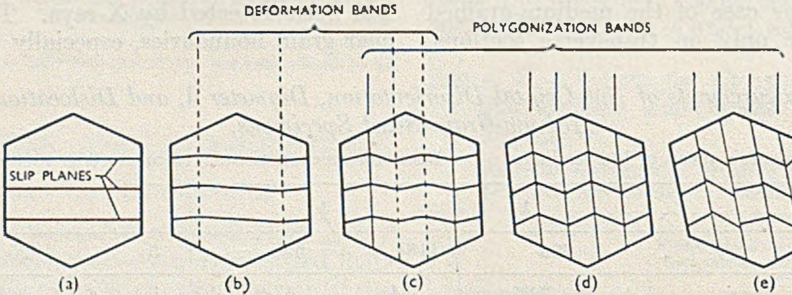


Fig. 8.—Model for Lattice Deformation During Creep. (*a*) Initial condition of crystal; (*b*) 1st-stage creep; (*c*) 1st–2nd-stage creep; (*d*) 2nd-stage creep; (*e*) late 2nd-stage creep. Changes shown for one operative slip direction only.

Some difficulty may be experienced in realizing how dislocation walls can produce the equiaxed sub-crystals observed, since ideally these are straight,[9] but if it is considered that in general there are several slip directions, and that associated with each direction of slip there exist both edge and screw dislocations which polygonize in different planes, it will be seen that complex shapes and complex rotations of the sub-crystals can readily arise.

Polygonization, of course, produces a relative rotation of the sub-crystals by kinking of the lattice, as shown diagrammatically in Fig. 8, discussed in Section IV. Wood and Wilms have put forward the alternative view that this rotation occurs primarily by relative displacement at the sub-boundaries. Fig. 7 (Plate XLII) is a typical interference micrograph of a sub-boundary (*PP*) in a specimen that had crept 11·9% in 180 hr. at $\frac{1}{2}$ ton/in.^{2}. The difference in orientation between the crystals on the two sides of the boundary was about 3° and the sub-crystal diameter was about 0·005 cm. If the difference in orientation had been produced by the crystals sliding past each other, the displacement at the sub-boundary would have been about 12,500 Å. No displacement at the sub-boundary can, however, be seen, although the microscope sensitivity is such that a displacement of 250 Å. can be detected. In the later stages of creep, when movements could be detected at the sub-boundaries, the amount of movement was less than one fifteenth of that which would be expected from the orientation differences between adjacent sub-crystals. For example, in the fine-grained specimen after 901 hours' creep, the average sub-boundary displacement was less than $0{\cdot}2 \times 10^{-4}$ cm.; since the average sub-crystal diameter was $2{\cdot}3 \times 10^{-3}$ cm., this would lead to an average disorientation of the sub-crystals of less than 1°. The measured disorientation was 16·6°. In a second case the measured average disorientation was 8·7°, whereas that calculated from the average sub-boundary displacement was less than $\frac{1}{2}$°.

Nevertheless, it is confirmed that sub-boundary slip does occur. Although under the present conditions these displacements were small, it is probable that at higher temperatures they would be relatively larger. They may be highly significant and are referred to again in the discussion (Section VI).

IV.—MODEL FOR CRYSTAL DEFORMATION DURING CREEP

1. Basis

A model representing crystal deformation during creep has been built up on three bases: (1) The well-established fact that sub-crystals form during creep. (2) The dislocation theory of sub-crystal formation. (3) The deduction [12] that during transient creep many dislocations travel to the grain boundaries. No effect of grain boundaries is considered.

The model is illustrated diagrammatically in Fig. 8, where, for clarity, the changes for one operative slip direction only are shown. Other operative slip directions would produce an additional similar effect, as already mentioned. The stages shown separately would, of course, overlap in practice. Initially, stage (*a*), the lattice planes are flat. During primary creep some dislocations approach the boundary to produce deformation bands (*b*), and others form deformation bands in the crystal interior (*c*). During secondary creep polygonization develops, first near the crystal boundaries (*c*), and then in the whole crystal (*d*).

The structure depicted in (*d*) and the process leading up to it appear to characterize secondary creep, as far as deformation of each crystal is concerned. Subsequently displacement occurs at the sub-boundaries (*e*).

2. Test of Model

The measurements of θ were used to check the model in Fig. 8, employing the relation : [10]

$$E\% \cong 100\theta \quad . \quad . \quad . \quad . \quad (2)$$

which is based essentially on this model. To avoid confusion it may be desirable before proceeding further to consider just what part of the total extension is involved in equation (2). In the earlier paper interest centred on estimating the "missing creep". This was compared with the extension calculated from equation (2), which should include both the creep due to "fine" slip and that due to slip on easily visible slip bands. In the present paper, the important point to decide is what part of primary and secondary creep is to be included. It is clear that extension due to grain-boundary movement is to be excluded. To go further, it may help to consider two extreme cases which are modifications of Fig. 8. In case (i) all the primary creep is due to dislocations which travel to the grain boundary, where they eventually polygonize to cause tilts of θ_1 near the boundary, and the extension is 100 θ_1. During secondary creep, when sub-boundaries have formed throughout the crystal interior, we have the case treated previously; the tilt is θ_2 and the extension 100 θ_2. Neglecting the intermediate condition, when only a few sub-boundaries have formed, or when they are too weak to be certain of capturing dislocations, the total extension is 100 $(\theta_1 + \theta_2)$. In case (ii) even during primary creep, dislocations do not travel a distance greater than the sub-crystal diameter, owing to internal obstructions such as solute atoms, and polygonize to cause sub-boundaries throughout the crystal interior at which the mean tilt is θ_3, giving a total extension of primary plus secondary creep of 100 θ_3. Case (i) is probably approached by large-grained specimens of pure metal, and case (ii) by fine-grained metals and particularly by alloys. In the present investigation case (i) is probably nearer the truth for the coarse-grained specimen, and case (ii) for the others; the difference in practice is not very large, and for simplicity case (ii) has been assumed throughout.

Table V.—*Comparison Between Calculated and Measured Secondary Creep.*

Grain-size of specimen . .	Fine (9½ grains/mm.)	Coarse (1 grain/mm.)	Medium (4½ grains/mm.)				
Load, ton/in.^{2} . . .	½	½	½		¾		
Creep time, hr. . . .	901	1145	118½	696	5	25	118½
Total creep (*A*), % . .	50·3	23·6	5·4	17·0	8·4	15·0	29·5
G.b. creep * (*B*), % . .	7·5	0·4	0·5	1·5	0·4	0·7	1·4
(*A*)–(*B*), % . . .	**42·8**	**23·2**	**4·9**	**15·5**	**8·0**	**14·3**	**28·1**
Calculated creep,† % . .	**29·0**	**15·0**	**6·2**	**24·4**	**7·0**	**8·0**	**27·9**

* Extension due to grain-boundary displacement. † Calculated from equation (2).

The data are given in Table V for all the conditions available. The two quantities to be compared are printed in heavy type.

The agreement between calculated and measured creep (*A–B*) is fairly close, not being more in error than 2 : 1 in any instance. The scatter in the calculated creep, as compared with the measured values, is probably due mainly to the wide variations between individual disorientations, so that to get a good mean in every case, a very large number of readings would have to be taken. However, the degree of agreement in Table V provides further quantitative support for stage (*d*) of the model in Fig. 8.

It should be emphasized that the creep calculated here relies not on measurements made on the specimen surface, but on sections cut through the interior, and is therefore representative of the whole specimen volume.

V.—DISLOCATION DENSITIES *

Since few estimates of the dislocation density based on fairly direct theoretical interpretation have been made, it seems of interest to report values here, as these can be calculated from the simply derived equation [10]:

$$\rho = \frac{2\theta}{d\lambda} \quad . \quad . \quad . \quad . \quad . \quad (3)$$

where ρ is the dislocation density (i.e. the number of dislocations intersecting a surface 1 cm.^{2} in area), λ is the unit slip distance, and d and θ, as before, are the sub-crystal diameter and disorientation, respectively. The qualification should be made that equation (3) counts only the dislocations in sub-boundaries. As, in the later stages of creep, the sub-boundaries appeared fairly undistorted when subjected to both micro- and X-ray examination, it is probable that the values relating to the later stages do include practically the whole dislocation density. It is equally probable that

---

* Counting only dislocations in the sub-boundaries.

the values relating to the early stages under-estimate the total density. The densities are included in Tables III and IV (pp. 288 and 289). The value taken for the sub-crystal diameter *d* was that found by the polarized-light method, i.e. the same method as that used for determining the disorientation. The densities are within the range of previous estimates, namely, that an annealed metal contains about 10^8 dislocations/cm.^{2} and a heavily cold-worked metal about 10^{12}/cm.^{2}

VI.—DISCUSSION

It is interesting to compare the sub-boundaries with the main grain boundaries in the same specimen. The displacement at the latter has been measured and is compared with that at the sub-boundaries in two specimens in Table VI.

TABLE VI.—*Comparison of Displacement θ at Sub-Boundaries and Main Boundaries.*

Grain-Size of Specimen	Sub-Boundaries		Main Boundaries
	Displacement, μ	Disorientation θ, degrees	Displacement, μ
Fine ($9\frac{1}{4}$ grains/mm.), after 901 hours' creep . .	≯0·2	16·6	11·0
Coarse (1 grain/mm.), after 1145 hours' creep .	≯0·1	8·7	6·3

The main grain boundaries therefore underwent more than 50 times as much displacement as the sub-boundaries. The mean orientation tilt across the main boundaries was doubtless larger than that across the sub-boundaries * but could clearly be larger only by a few times. As there is no evidence that the rate of sub-boundary displacement increases rapidly with increasing disorientation, it appears as though a difference in kind exists between sub-boundaries and main boundaries. According to the dislocation theory of polygonization, the former consist initially of rows of dislocations. An alternative structure is that of a random arrangement of atoms, and presumably the main boundaries are so composed.

Another approach leads to the same conclusion. The polygonized array of dislocations forms because it has less energy than a more or less random arrangement of dislocations. It does not follow that the polygonized array is the most stable arrangement of atoms that can accommodate the orientation differences across the sub-boundary. In fact, since there are innumerable alternative arrangements of atomic misfit, this is most unlikely. The polygonized array of dislocations is simply the arrangement first formed by migration from the initial random arrangement. The main grain boundaries, however, are formed in a different way during recrystallization and usually have more opportunity of approaching the arrangement of lowest energy. They are therefore likely to comprise some arrangement of general atomic misfit.

Given that an arrangement of general atomic misfit of lower energy than that due to polygonization exists, this will in time be attained by sub-boundaries initially polygonized; these will then have become ordinary grain boundaries and will display their properties, such as grain-boundary displacement and grain growth. The material will then have recrystallized.

The suggestion is therefore made that the sub-boundaries observed in the present work gradually underwent this transformation during creep, but that it was not complete before the tests ended. The observed results would then be produced if displacement occurred only along transformed parts of the sub-boundaries, and was there restricted by the anchoring effect of the untransformed parts.

ACKNOWLEDGEMENTS

The author desires to acknowledge the assistance given by Mr. M. H. Farmer. The work described was carried out as part of the general research programme of the National Physical Laboratory and the paper is published by permission of the Director of the Laboratory.

REFERENCES

1. C. H. M. Jenkins and G. A. Mellor, *J. Iron Steel Inst.*, 1935, **132**, 179P.
2. G. R. Wilms and W. A. Wood, *J. Inst. Metals*, 1948–49, **75**, 693.
   W. A. Wood, G. R. Wilms, and W. A. Rachinger, *ibid.*, 1951, **79**, 159.
3. E. A. Calnan and B. D. Burns, *ibid.*, 1950, **77**, 445.
4. G. B. Greenough and E. M. Smith, *ibid.*, 1950, **77**, 435.
5. G. Wyon and C. Crussard, *Rev. Mét.*, 1951, **48**, 121.
6. R. C. Gifkins, *J. Inst. Metals*, 1951, **79**, 233.
7. W. Betteridge and A. W. Franklin, *ibid.*, 1951–52, **80**, 147.
8. A. H. Cottrell and V. Aytekin, *ibid.*, 1950, **77**, 389.
9. R. W. Cahn, *ibid.*, 1949–50, **76**, 121.
10. D. McLean, *ibid.*, 1951–52, **80**, 507.
11. N. F. Mott, *Proc. Phys. Soc.*, 1951, [B], **64**, 729.
12. D. McLean, *J. Inst. Metals*, 1952–53, **81**, (3), 133.
13. D. McLean, *ibid.*, 1952–53, **81**, (6), 293.
14. R. W. K. Honeycombe, *ibid.*, 1951–52, **80**, 39.
15. W. A. Wood and W. A. Rachinger, *ibid.*, 1949–50, **76**, 237.

---

* Some measurements with the polarized-light technique on an annealed specimen gave $\alpha = 29{\cdot}4^\circ$, equivalent to a total mean tilt $\theta = 40{\cdot}6^\circ$.

GRAIN-BOUNDARY SLIP DURING CREEP OF ALUMINIUM *

By D. McLEAN,† B.Sc., MEMBER

(Communication from the National Physical Laboratory.)

SYNOPSIS

Grain-boundary displacements during creep at 200° C. have been measured in seven super-pure aluminium specimens. The tests covered a range of grain-size from 1 to 9¼ grains/mm. and of stress from about ⅓ to about ¾ ton/in.^{2}. The curves of grain-boundary displacement plotted against time resemble the corresponding extension/time curves. The fraction of the total extension due to the grain-boundary displacements was calculated. At a constant load of ½ ton/in.^{2} this increased with decrease in grain-size from about one-fiftieth for 1 grain/mm. to about one-sixth for a grain-size of 9¼ grains/mm. At constant grain-size (4½ grains/mm.) it increased with decrease in load, from about one-fiftieth for a load of 1 ton/in.^{2} to about one-fifth for a load of ¼ ton/in.^{2}. A plot of grain-boundary displacement against extension due to crystal deformation is linear for all specimens, suggesting a linear interaction between these two quantities. A model for this interaction is proposed, and calculation shows it to agree with experiment within 2 : 1.

I.—INTRODUCTION AND MAIN RESULTS

In work previously reported [1] the average grain-boundary displacement in an aluminium specimen undergoing creep extension was measured and the extension due to these movements calculated. An improved technique [2] has now been used to measure the influence of grain-size and applied stress on such grain-boundary displacements.

Earlier work had shown that the ratio of grain-boundary movement to slip-band movement is higher the lower the stress [3, 4] and that there is a tendency for fine-grained material to creep faster than coarse-grained.[3-8] These effects have been confirmed. In particular, a fine-grained specimen crept about twice as fast as a coarse-grained one. However, the grain-boundary displacements were responsible for not more than about one-sixth of the total extension, even in the fine-grained specimen, so that the more numerous grain boundaries in the latter cannot be an important cause of the greater creep rate. Although the grain-boundary slip was faster in the fine-grained specimen, the crystals themselves must also have deformed more rapidly.

The explanation of this fact may be assisted by a new observation. For all specimens, covering a range of creep rates of 30 : 1, a linear relation is found between the grain-boundary displacement and the crystal-lattice slip. From this it is inferred that an interaction occurs between the two. A model for this interaction is proposed that accords with qualitative observation and provides about the correct quantitative relation. According to this model the grain-boundary movement is linked to the process of crystal recovery.

Grain-boundary movements have been measured previously by Kê,[9] using a different method from that now employed. Since it is of interest to compare the results given by the two different methods, Section VII (p. 299) is devoted to such a comparison.

II.—EXPERIMENTAL

The preparation of the specimens and technique of measurement have been described fully in a paper [2] giving the results of measurements of slip bands in the same specimens as were used in the present experiments. Briefly, these were all of super-pure aluminium of about 99·98% purity (0·002–0·003% iron, 0·003% copper, 0·003% silicon), electropolished to avoid work effects. Different grain-sizes were obtained by annealing at different temperatures. As a precautionary measure all specimens were furnace-cooled from the various annealing temperatures to the test temperature of 200° C. in order to develop the equilibrium condition, e.g. as regards distribution of impurities between grain interior and grain boundary, pertaining to 200° C.

Seven specimens were employed. Two of these, one having 1 grain/mm. and the other 9¼ grains/mm.‡ were used to explore the influence of grain-size. They were stressed at ½ ton/in.^{2} and removed periodically for examination and measurement. The other five specimens were used to determine the influence of applied stress. They had a grain-size of 4½ grains/mm., and were tapered in width, a convenient way of obtaining a range of load extending from about ⅓ to about ¾ ton/in.^{2}. Each was loaded for a given time (1, 5, 25, 118½, or 696 hr.), removed from the machine, examined, and measured. From these measurements, curves against time for different loads from ⅓ to ¾ ton/in.^{2} could be constructed. Whereas

---

* Manuscript received 21 June 1952.

† Metallurgy Division, National Physical Laboratory, Teddington.

‡ A duplicate fine-grained specimen was also prepared and measurements made on it after 6 hours' creep. The result is included in Fig. 2 (*b*) and shows satisfactory agreement.

the first two specimens were subjected to intermittent loading with opportunities for recovery, the latter five were continuously loaded, so that the results from the two sets of specimens are not strictly comparable.

The grain-boundary displacements are given as the average total displacement, $\bar{p}$, rather than as the actual measured vertical component. The former is obtained from the latter by a simple averaging procedure.[2]

III.—MEASUREMENTS OF GRAIN-BOUNDARY DISPLACEMENT

The influence of stress on grain-boundary displacement is shown in Fig. 1, where displacement is plotted against initial stress for the five medium-grain-sized (tapered) specimens. Each curve therefore relates to a particular duration of creep. It is clear that the grain-boundary displacement increased with load faster than linearly.

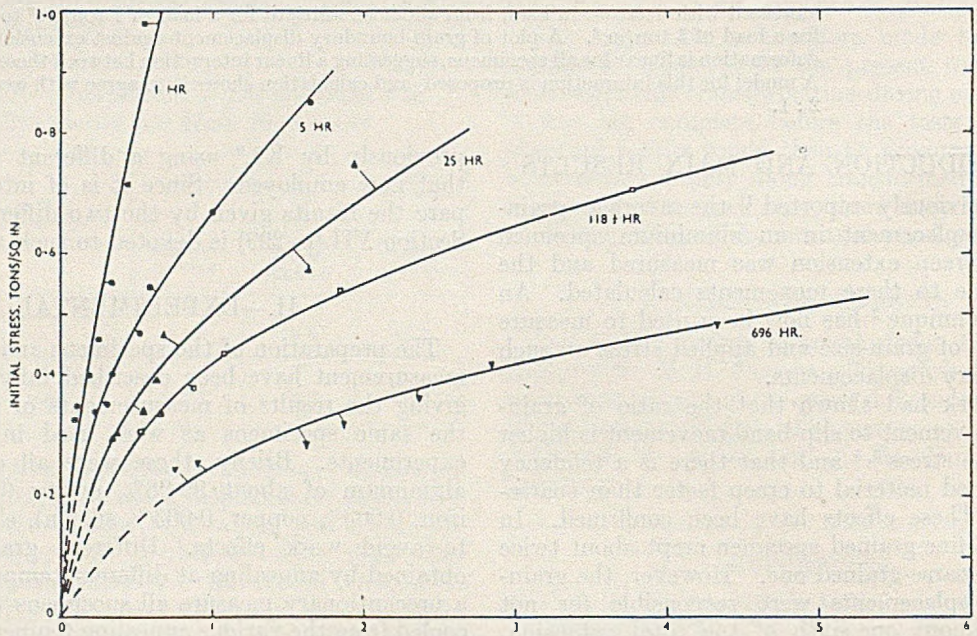


Fig. 1.—Influence of Initial Stress on Grain-Boundary Displacement for Medium-Grain-Sized Specimens. The tests were made at constant load, so that the actual stress increased with extension.

From Fig. 1 were derived grain-boundary displacement/time curves for the three initial stresses of $\frac{1}{3}$, $\frac{1}{2}$, and $\frac{3}{4}$ ton/in.², and these are given in Fig. 2 (*a*), together with the corresponding curves for the fine- and coarse-grained specimens for $\frac{1}{2}$ ton/in.². The initial portions of these curves are shown on an enlarged time scale in Fig. 2 (*b*).

The curves closely resemble a typical creep curve, exhibiting the primary and secondary stages typical of creep extension/time curves, as may be seen by comparing Fig. 2 (*a*) with the extension/time curves for these specimens given in Fig. 3. The corresponding curves show the change from primary to secondary creep at about the same time; the influence of stress on displacement and extension is similar; so, too, is the effect of difference in grain-size. The primary grain-boundary displacement and extension both increase with increase in grain-size, while the secondary rates decrease, with the result that the curves for different grain-sizes eventually cross. In this connection it should be borne in mind that the quantity plotted as grain-boundary displacement is the mean displacement at a single grain boundary, so that the smaller number of grain boundaries in the coarse-grained specimen is not responsible for the slower rate of displacement during secondary creep shown for this specimen in Fig. 2 (*a*). Two further points of resemblance between the displacement and the extension curves are that for the fine-grained specimen both show the tertiary stage of creep, and for the coarse-grained specimen both show, on the other hand, a fall in creep rate between the last two measurements. The latter behaviour seems anomalous, and the reason for it is uncertain, but that it occurred both in the grain-boundary displacement curve and the elongation curve provides additional evidence that the two quantities are connected.

The curves in Figs. 4 and 5 show how the rate of grain-boundary displacement $d\bar{p}/dt$, plotted logarithmically, changes as the displacement increases. Fig. 4 includes the curves for the fine- and coarse-grained specimens and shows the influence of grain-size for a given load; Fig. 5 reproduces the curves for the medium-grain-sized (tapered) specimens and illustrates the effect of stress for a given grain-size. The general trend is for $d\bar{p}/dt$ to fall steeply at first and then to level off; the curve for the fine-grained

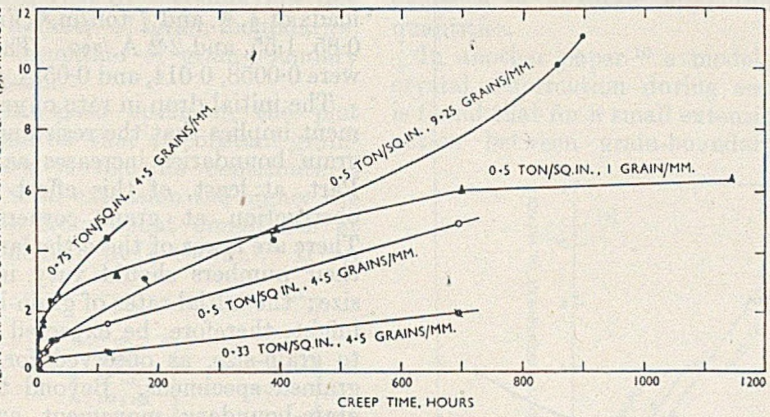


Fig. 2 (*a*).—Grain-Boundary Displacement During Creep Tests.

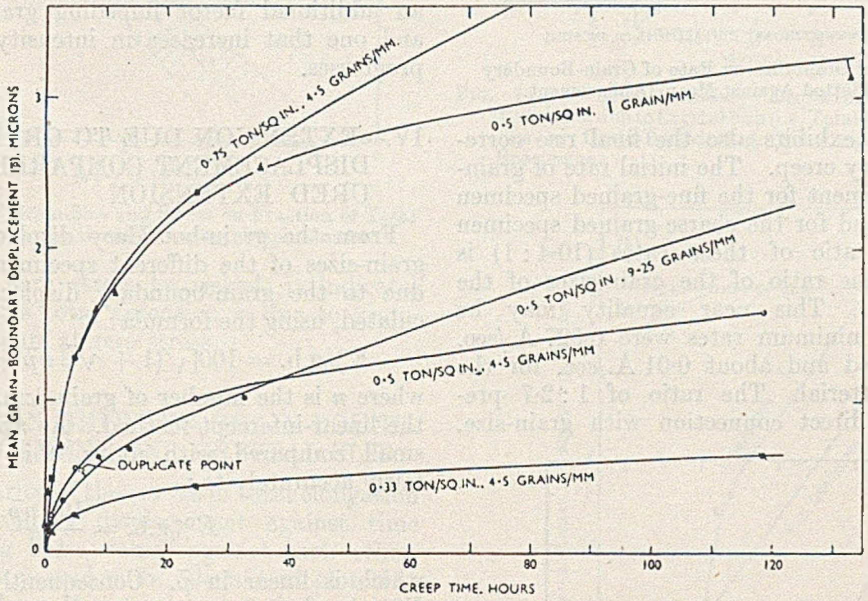


Fig. 2 (*b*).—As Fig. 2 (*a*) (on enlarged scale).

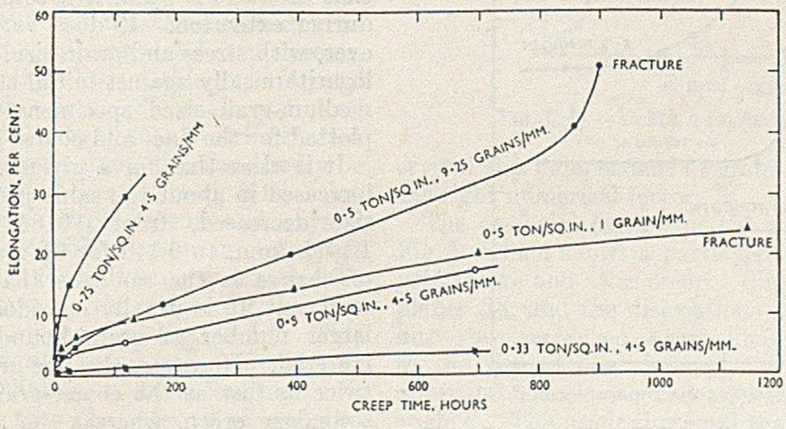


Fig. 3.—Creep Time/Elongation Curves.

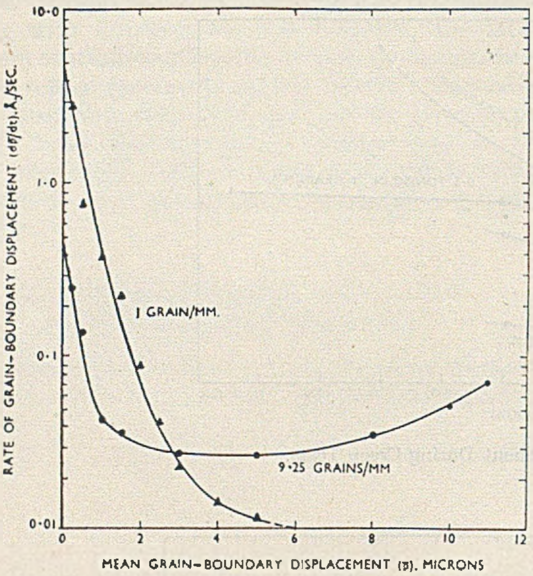


Fig. 4.—Influence of Grain-Size on Rate of Grain-Boundary Displacement Plotted Against Mean Displacement.

specimen (Fig. 4) exhibits also the final rise corresponding to tertiary creep. The initial rate of grain-boundary displacement for the fine-grained specimen was 0·45 Å./sec. and for the coarse-grained specimen 4·7 Å./sec. The ratio of these rates (10·4 : 1) is nearly equal to the ratio of the grain-sizes of the specimens ($9\frac{1}{4}$: 1). This near equality may be significant. The minimum rates were 0·027 Å./sec. for the fine-grained and about 0·01 Å./sec. for the coarse-grained material. The ratio of 1 : 2·7 presumably has no direct connection with grain-size.

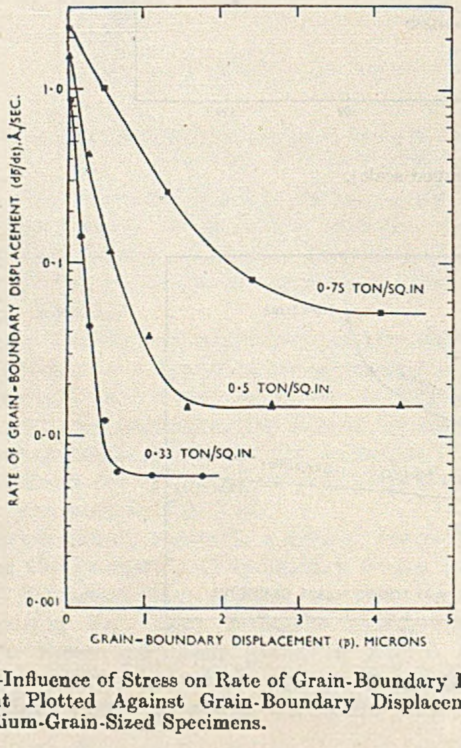


Fig. 5.—Influence of Stress on Rate of Grain-Boundary Displacement Plotted Against Grain-Boundary Displacement for Medium-Grain-Sized Specimens.

The initial rates of grain-boundary displacement for loads of $\frac{1}{3}$, $\frac{1}{2}$, and $\frac{3}{4}$ ton/in.^{2} (Fig. 5) were respectively 0·85, 1·55, and 2·2 Å./sec. The minimum creep rates were 0·0058, 0·014, and 0·05 Å./sec. for the same loads.

The initial drop in rate of grain-boundary displacement implies that the resistance to movement at the grain boundaries increases as movement continues. Part, at least, of this effect is presumably due to obstruction at grain corners and protuberances. There are fewer of these the larger the grain-size, and their numbers should vary inversely as the grain-size; the initial rates of grain-boundary displacement might, therefore, be expected to vary in proportion to grain-size, as observed for the fine- and coarse-grained specimens. Beyond this, the mechanics of grain-boundary movement are unclear. However, one possibility is that, as crystal slip occurs concomitantly with grain-boundary displacement, the boundaries may become roughened, and this may be an additional factor impeding grain-boundary slip and one that increases in intensity as deformation progresses.

IV.—EXTENSION DUE TO GRAIN-BOUNDARY DISPLACEMENT COMPARED WITH MEASURED EXTENSION

From the grain-boundary displacements and the grain-sizes of the different specimens, the extension due to the grain-boundary displacements was calculated, using the formula:[1]

$$E\%\text{g.b.} = 100[\sqrt{(1 + \sqrt{2}\, n\bar{p} + n^2\bar{p}^2)} - 1]$$

where n is the number of grains/cm. as measured by the linear-intercept method. As $n\bar{p}$ in most cases is small compared with unity, this formula reduces fairly accurately to:

$$E\%\text{g.b.} = \frac{100\, n\bar{p}}{\sqrt{2}}$$

which is linear in $\bar{p}$. Consequently, the curves of $E\%$g.b. plotted against time have the same shape as those of $\bar{p}$ against time, and therefore the same as those of total elongation against time. This means that the ratio $E\%$g.b./$E\%$ total does not vary much during extension. It does vary considerably, however, with stress and grain-size. The ratio is plotted logarithmically against initial stress in Fig. 6 for the medium-grain-sized specimens, and points are also plotted for the fine- and coarse-grained specimens.

It is clear that for a given initial stress the ratio increased in about the same proportion as the grain-size decreased, from 0·0186 for a grain-size of 1 grain/mm. to 0·148 for $9\frac{1}{4}$ grains/mm. This sense of change is the same as that found by previous workers. It is usually considered to be due to the larger number of grain boundaries in fine-grained material. However, the fine-grained specimen crept twice as fast as the coarse-grained specimen during secondary creep, whereas the additive effect of the grain boundaries could give an increase only in the ratio of 5 : 4. The faster creep rate of the fine-grained specimen in this case is therefore not due simply to the larger number of grain boundaries. Nor is it due to the formation of grain-boundary fissures, for none were seen.

For the medium-grain-sized specimen the plot against initial stress confirms that at constant grain-size the grain-boundary movements contribute a smaller amount to the total extension the higher the load. The ratio ranges from about one-fiftieth at 1 ton/in.^{2} to about one-fifth at $\frac{1}{4}$ ton/in.^{2}. If an extrapolation to zero stress be risked, it indicates that about half the total extension is due to grain-boundary movements at zero stress.

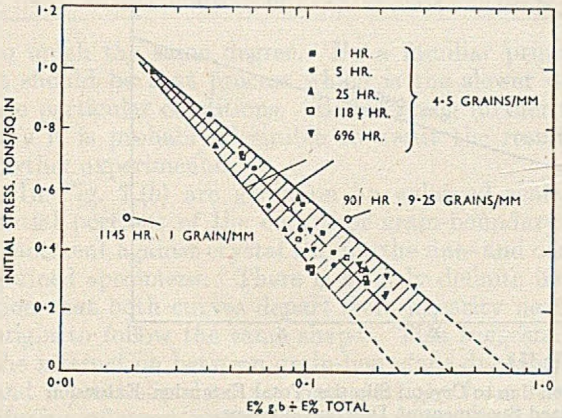


Fig. 6.—Influence of Grain-Size and Stress on Fraction of Total Creep Extension due to Grain-Boundary Displacement.

V.—INTERACTION BETWEEN GRAIN-BOUNDARY DISPLACEMENT AND CRYSTAL SLIP

The close similarity of the curves of total elongation and of grain-boundary displacement against time suggested plotting these values against each other. It subsequently appeared more relevant to plot that part of the total elongation not due to grain-boundary displacement (and presumably due to crystal slip) against the grain-boundary displacement. The two curves in Figs. 7 (*a*) and (*b*) relate to the fine- and coarse-grained specimens and the three in Fig. 8 to the medium-grain-sized specimens at different stresses.

All the curves are straight lines within experimental error, except near the origin. Since the range of creep rates covered was about 30 : 1 and different grain-sizes and stresses were used, this can hardly be accidental, but suggests rather a linear interaction between grain-boundary displacement and crystal slip. The slopes increase with decrease in stress, but do not alter appreciably for a difference in grain-size of nearly 10 : 1, as shown by the curves for the fine- and coarse-grained specimens in Fig. 7 (*a*). This requires that the interaction be virtually independent of grain-size. The actual creep rates for these two specimens differed, so that the rates of crystal slip and of grain-boundary displacement must have changed in the same ratio. This provides further evidence of a linear interaction between the two quantities.

In another paper[10] a model is proposed for the crystal deformation during second-stage creep. It is found that for a small extension this gives an interaction between grain-boundary displacement and crystal slip independent of grain-size and with about the right numerical factor.

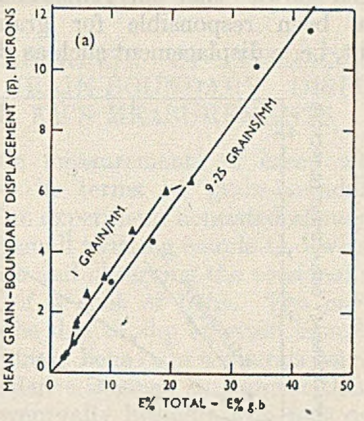


Fig. 7 (*a*).—Grain-Boundary Displacement Plotted Against Extension due to Crystal Slip (i.e. Total Extension—Extension due to Grain-Boundaries), for Large- and Fine-Grained Specimens.

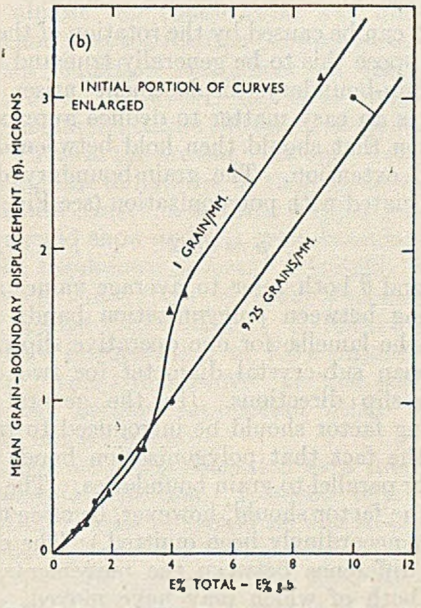


Fig. 7 (*b*).—As Fig. 7 (*a*) (on enlarged scale).

The essential feature of the model is depicted in Fig. 9, which shows a particular crystal in an aggregate before and after creep. Slip occurs on the glide planes *SS*, and the dislocations generated polygonize into rows arranged approximately at right angles to the slip planes, causing the lattice to be tilted alternately clockwise and anti-clockwise through an angle θ. The original crystal therefore splits up into sub-crystals bounded by rows of dislocations. For clarity, Fig. 9 shows the deformation caused by slip in one direction only, but the same process is presumed to happen for each operative slip direction. For one operative slip direction the crystal splits up into sub-crystals of lamellar shape, as shown in Fig. 9, but for three operative directions the resultant sub-crystals are fairly equiaxed.

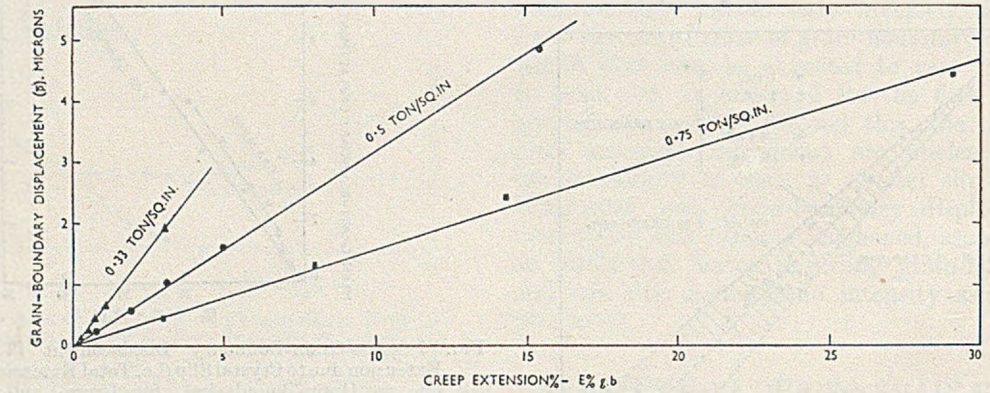


Fig. 8.—Grain-Boundary Displacement Plotted Against Extension due to Crystal Slip (i.e. Total Extension–Extension due to Grain Boundaries), for Medium-Grain-Sized Specimens at Different Stresses.

It has been shown [1] that polygonization can sometimes have been responsible for grain-boundary displacement, i.e. a displacement such as that marked p in Fig. 9 can be caused by the rotation of the lattice. Let us suppose this to be generally true and that the whole grain-boundary displacement arises in this way. It is an easy matter to deduce approximately the relation that should then hold between displacement and extension. The grain-boundary displacement associated with polygonization (see Fig. 9) is :

$$\bar{p} = d\bar{\theta}/2 \; . \quad . \quad . \quad . \quad . \quad (1)$$

where $\bar{p}$ and $\bar{\theta}$ both refer to average values and d is the spacing between polygonization bands, i.e. the width of the lamellæ for one operative slip direction or the mean sub-crystal diameter for two or more operative slip directions. (In the general case a multiplying factor should be introduced to take into account the fact that polygonization bands are not necessarily parallel to grain boundaries. The average value of the factor should, however, be close to unity, and it has accordingly been omitted.) The observed p is the difference between the movements of two crystals, both of which may have moved. If they move in opposite directions, the measured p would be twice that given by equation (1). If they move in the same direction, $p = 0$. The average measured value, $\bar{p}$, is therefore given by equation (1), and this is true whatever the direction of movement, provided that the movements are random. Using the equation connecting extension due to crystal slip with disorientation at polygonized boundaries : [1]

$$E \cong \bar{\theta}/2 \; . \quad . \quad . \quad . \quad . \quad (2)$$

where E is the fractional extension due to crystal slip, we obtain, substituting for $\bar{\theta}$ in equation (1) :

$$\bar{p} = dE \quad . \quad . \quad . \quad . \quad . \quad (3)$$

(The factor ½ enters because one direction only of slip is under consideration.) This equation has the correct form, since it gives a linear relation between grain-boundary displacement and extension which is independent of grain-size. According to this equation the measured value of d, the average sub-crystal size, should be equal to the value of $\bar{p}/E$ given by the slopes of the curves in Figs. 7 and 8. Table I shows that this is so.

Details of the measurement of d are given in another paper.[10] The agreement is as close as can reasonably be expected. It seems, therefore, that the model in Fig. 9, on which equation (3) is based, is possibly correct, and that under the conditions of test used in this work practically all the grain-boundary displacement may have been associated with polygonization in the manner depicted in Fig. 9. The existence of a relationship between the values of $\bar{p}$, which are measured on the surface of the specimen, and of E, which is a measurement of the extension of the material as a whole, suggests that the surface measurements do not seriously misrepresent the grain-boundary movements throughout the metal.

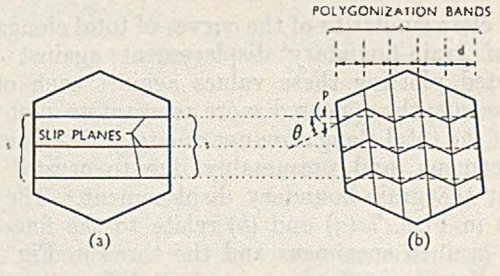


Fig. 9.—Model for Crystal Deformation and Grain-Boundary Movement During Creep. (*a*) Initial state of crystal; (*b*) State of crystal after creep.

The model in Fig. 9 for the interaction between crystal slip and grain-boundary slip suggests the probability that one of these processes controls the other, and hence the total creep rate. It is a crucial matter to decide which process predominates, since methods of influencing it will influence the creep rate

TABLE I.—*Comparison between Sub-Crystal Diameter* (d) *and* $\bar{p}/E$

Specimen and Grain-Size	Initial Stress, ton/in.^{2}	d, cm. × 10^{-3}	$\bar{p}/E$, cm./fractional extension × 10^{-3}
Fine-grained (9¼ grains/mm.)	½	2·3	2·8
Coarse-grained (1 grain/mm.)	½	2·5	2·9
Tapered, medium-grained (4½ grains/mm.)	⅓	8·0	6·6
	½	3·4	3·2
	¾	2·2	1·55

to much the same degree. By a familiar principle, it should be that process which is the slower under the particular conditions. Before going further than this it is probably desirable to await the results of further experiments.

In Fig. 7 (*b*) are given on an enlarged scale the initial portions of the curves of grain-boundary displacement against crystal slip for the fine- and coarse-grained specimens. There is a fairly definite indication that both curves depart from linearity near the origin to follow the same shape. This suggests that the interaction between grain-boundary displacement and crystal slip does not operate right from the beginning of creep, and is consistent with the view that it characterizes the second stage of creep. Since, during the primary stage, the coarse-grained specimen extended much more than the fine-grained one, the fact that the departure from linearity extended to a larger elongation for the coarse-grained specimen is understandable on this hypothesis.

VI.—GRAIN GROWTH DURING CREEP

Besides the tangential slip at grain boundaries so far discussed, growth of one grain into another also occurred, as has been observed by Wyon and Crussaid [11] and by Chang and Grant.[12] This is illustrated in Figs. 11 and 12 (Plate XLII). Fig. 11 shows a boundary that has moved about 20 μ into the neighbouring grain. Fig. 12 is of the same field at a later stage and makes clear that growth of the left-hand grain into the right-hand one is actually occurring. Attention may be drawn to the "tide marks" parallel to the boundary, which are presumably due either to the grain growth or to the boundary slip, or both, being intermittent; and also to the tendency of the tide marks to anchor to etch pits, which is suggestive of a surface-tension effect.

Grain growth was considerably more marked the finer the grain-size. A possible reason for this is that the higher temperature of anneal used to produce a coarser-grained structure gives the grain-boundary atoms more opportunity of reaching positions of a good fit, i.e. of low energy, before test, than those in finer-grained specimens. If so, this difference must have persisted throughout the 700–1000 hr. of creep of the specimens to cause the observed difference in grain growth during creep. Taking the amount of grain growth occurring during any heating period as a measure of the effectiveness of the heating in promoting grain-boundary equilibrium, this seems reasonable, because the amount of grain growth that occurred during the 1000 hr. of creep was much less than the amount that must have occurred during the annealing treatments to produce the stated grain-sizes.

VII.—COMPARISON OF PRESENT VALUES OF GRAIN-BOUNDARY DISPLACEMENT WITH KÊ'S MEASUREMENTS

Kê made measurements of creep which he has interpreted in terms of grain-boundary displacement.[9] The experiment consisted of suddenly applying a very small twisting couple to a wire at elevated temperature and observing the torsional creep by the deflection of a spot of light. The creep was considered to be due to slip at grain boundaries, which were thought to behave in a viscous way, and a good deal of evidence is given to support this view. The slip was eventually brought to a halt owing, it was thought, to the resistance set up at grain corners. It was considered that such resistance was purely elastic at the low strains involved, an assumption supported by the facts that on removing the applied couple almost the entire twist was ultimately recovered and that the relation between stress and limiting creep deflection was linear for the very small stresses used. However, for large grain-boundary displacements the elastic resistance must clearly break down. Those in the present work must have been large enough for this to happen, so that, if the assumptions involved are correct, a suitable comparison with Kê's results should show at what grain-boundary displacement the breakdown occurs. A suitable comparison is between the curves of grain-boundary velocity plotted against displacement. To obtain these quantities some manipulation of Kê's published data is necessary; this, can, however, be done without introducing any serious assumption other than those which he himself made.

Kê gives the creep deflection as a fraction of the instantaneous deflection. It seems required by his results that the latter should be due to elastic shear. For a grain-size of a cm. the elastic-shear displacement across each grain, produced by a shear stress of S per unit area, would be Sa/G, where G is the rigidity modulus. The limiting creep deflection was observed to be 50% of the instantaneous deflection, and hence :

$$\text{grain-boundary displacement} = \tfrac{1}{2}\,\frac{Sa}{G}\sqrt{2}\ . \quad (4)$$

the factor $\sqrt{2}$ being due to the grain boundaries making, on an average, an angle of 45° with the applied shear stress. The grain-size of the wires used by Kê was 0·03 cm., and the rigidity modulus of aluminium is $2{\cdot}7 \times 10^{11}$ dynes/cm.^{2}. For an applied shear stress of ¼ ton/in.^{2} ($0{\cdot}4 \times 10^8$ dynes/cm.^{2}), a value used in the present work, the limiting grain-boundary displacement therefore works out at 320 Å., and this is the displacement that would have been reached had elastic resistance at the grain corners been maintained. Since the overall shear measured by the observed deflections must be in simple proportionality to the grain-boundary displacements—there being, according to Kê's assumptions, no other process contributing to creep—the displacements at intermediate stages can be worked out from the published deflections by simple arithmetic. Results are given in Table II for observations at 200° C. The grain-boundary velocities can be obtained from a plot of displacement against time on a linear scale, and these are also included in Table II.

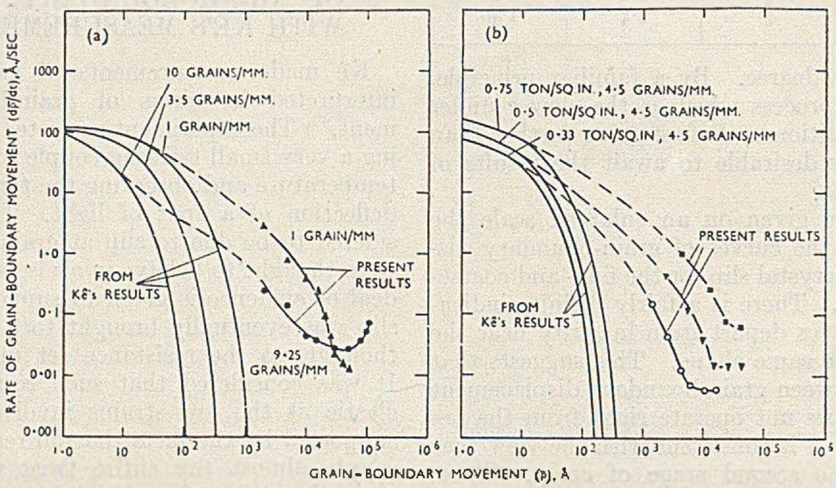


Fig. 10.—Comparison Between Results for Grain-Boundary Velocity Obtained by Kê [9] and in Present Work. (*a*) Specimens of various grain-sizes. (*b*) Medium-grain-sized specimens at different stresses.

Table II.—*Grain-Boundary Displacements and Velocities for Aluminium at* 200° *C. from Kê's Measurements.*[9]

Time, sec.	0	0·1	1·0	10	100	1000	3000
Displacement, Å.	0	13	38	73	190	280	320
Velocity, Å./sec.	120	54	11·6	1·6	0·4	0·04	0

The curve of velocity against displacement given by these results is presented in Fig. 10 (*a*) with logarithmic ordinate and abscissa. Two other curves have been drawn for grain-sizes of 0·1 and 1·0 mm. to compare with the fine- and coarse-grained specimens used in the present work. In Fig. 10 (*b*) three curves are drawn for a constant grain-size of 4½ grains/mm. and relating to stresses of ⅓, ½, and ¾ ton/in.², again plotted logarithmically, to compare with the other conditions used in the present work. The initial portions of the derived curves have been drawn according to Kê's model, which gives the initial velocity as independent of grain-size, but proportional to stress. The curves with plotted points in the two figures, show the relation found in the present work for the five different combinations of grain-size and stress, and it is clear that their position in relation to the former curves is consistent with the idea that the elastic resistance at grain corners and protuberances breaks down at a certain grain-boundary displacement, since they lie well to the right of the curves derived from Kê's results. Corresponding pairs of curves can be joined up in a variety of ways, but if this is done without violent curvature, as indicated by the broken lines in Fig. 10 (*a*) and (*b*), they signify that the elastic resistance breaks down at about 100 Å. mean displacement in the coarse-grained specimen and at about 10 Å. mean displacement in the other specimens. The former value seems reasonable, but the latter value appears to be too low to be consistent with the linearity that Kê has reported.

ACKNOWLEDGEMENTS

The author desires to acknowledge the helpful assistance given by Mr. M. H. Farmer. The work described above was carried out as part of the general research programme of the National Physical Laboratory, and the paper is published by permission of the Director of the Laboratory.

REFERENCES

1. D. McLean, *J. Inst. Metals*, 1951–52, **80**, (9), 507.
2. D. McLean, *ibid.*, 1952–53, **81**, (3), 133.
3. W. Rosenhain and J. C. W. Humfrey, *J. Iron Steel Inst.*, 1913, **87**, 219P.
4. W. A. Wood, G. R. Wilms, and W. A. Rachinger, *J. Inst. Metals*, 1951, **79**, 159.
5. D. Hanson and M. A. Wheeler, *ibid.*, 1931, **45**, 229.
6. C. L. Clark and A. E. White, *Proc. Amer. Soc. Test. Mat.*, 1932, **32**, (II), 492.
7. J. McKeown, *J. Inst. Metals*, 1937, **60**, 201.
8. K. von Hanffstengel and H. Hanemann, *Z. Metallkunde*, 1938, **30**, 41.
9. T. S. Kê, *Phys. Rev.*, 1947, [ii], **71**, 533.
10. D. McLean, *J. Inst. Metals*, 1952–53, **81**, (6), 287.
11. G. Wyon and C. Crussard, *Rev. Mét.*, 1951, **48**, 121.
12. H. C. Chang and N. J. Grant, *Trans. Amer. Inst. Min. Met. Eng.*, 1952, **194**, 619 (in *J. Metals*, 1952, **4**, (6).

INTERCRYSTALLINE CORROSION IN CAST ZINC–ALUMINIUM ALLOYS*

By C. W. ROBERTS,† B.Sc., A.I.M., MEMBER

SYNOPSIS

To determine the susceptibility of zinc–aluminium alloys to intercrystalline corrosion in an air/water-vapour atmosphere at 95° C. and to examine the effect of the presence of other elements on the incidence of this form of corrosion, alloys of various compositions within the range aluminium 0–22%, copper 0–1·5%, magnesium 0–0·09%, lead, tin, and cadmium 0–0·030%, bismuth 0–0·016%, and manganese 0–0·050% were prepared and tested in the as-cast condition.

The main conclusions drawn from the work are that: (1) intercrystalline corrosion is confined to the α (zinc-rich) phase, although attack is more severe when the β (aluminium-rich) phase is also present, as a result of the larger surface area of the α grains in the two-phase alloys; (2) the severity of attack is greatly increased by the presence of small percentages of lead, tin, cadmium, and bismuth; (3) the addition of a small amount of magnesium greatly reduces the severity of the corrosion, whether impurities are present or not, provided conditions are such that intermetallic compounds of magnesium with the impurity elements are not formed; and (4) the presence of copper increases the resistance of two-phase alloys to intercrystalline attack.

The results of the present investigation and those published by other investigators are discussed, and tentative theories are put forward to explain certain aspects of the phenomenon.

I.—INTRODUCTION

ZINC alloys have been used for pressure die-castings since the early years of the century, the alloying elements originally used being copper and tin, with smaller amounts of aluminium to prevent contamination of the alloy by iron during melting. It was later discovered that by increasing the aluminium content, the fluidity of these alloys was increased to such an extent that tin could be dispensed with altogether. By 1918 alloys containing up to 10% aluminium as the major alloying constituent had been widely adopted in the pressure die-casting industry, and large numbers of castings went into service. Serious corrosion trouble was soon encountered, however, particularly when the castings were used in humid conditions, and various cases were reported of castings distorting, swelling, and even disintegrating completely.

At that time any available grade of zinc, and even scrap metal, was used in the production of the alloys. Research showed that such failures were due to intercrystalline corrosion and that the severity of the attack was closely associated with the presence in the alloys of small quantities of certain impurities, in particular of lead. Subsequently, only the highest-purity metal then available (99·94%) was used in the manufacture of these alloys, and this resulted in a marked decrease in the number of failures of die-castings attributable to intercrystalline corrosion, although the scope of application of such castings was very limited.

About 1923 it was found that the presence of approximately 0·10% magnesium and up to 3·0% copper reduced still further the susceptibility of zinc–aluminium alloys to intercrystalline attack and when, a year or so later, zinc of purity 99·99% became available commercially and was used in the manufacture of the alloys, the number of failures of die-castings due to intercrystalline corrosion decreased to very small proportions.

With the use of high-purity metal it was found that about 0·04% magnesium was sufficient to confer on the alloys immunity from intercrystalline attack and that the presence of copper in alloying proportions was no longer necessary. Consequently, the zinc die-casting alloy most commonly used in this country became standardized at 4·1% aluminium and 0·04% magnesium (Mazak 3), with impurities limited to very small percentages. An alloy of similar composition, but containing in addition about 1·0% copper (Mazak 5), is also used to some extent, both materials being covered by British Standard No. 1004.

Castings made from these materials are very resistant to both surface and intercrystalline corrosion when exposed in normal atmospheric conditions, but if, through contamination during processing, the percentage of harmful impurities reaches a figure much in excess of the limits stated in the relevant specification, the castings may become susceptible to corrosion of the intercrystalline type, particularly if they are used in warm, humid conditions.

A considerable amount of work has been done in the past with a view to elucidating the various phenomena associated with intercrystalline corrosion in these alloys, but the results reported are to some extent conflicting. The work described below was carried out in an attempt to clarify the problem.

---

* Manuscript received 30 July 1952.

† Research Department, Imperial Smelting Corporation, Ltd., Avonmouth.

II.—PREVIOUS WORK

Brauer and Peirce,[1] in 1923, made a very comprehensive study of the subject, using gravity-die-cast test-pieces prepared from various grades of zinc (one of which was of 99·98% purity). They used an air/water-vapour atmosphere at 95° C. as the corroding medium and from the results obtained came to the following conclusions: (1) Intercrystalline attack is confined to the α phase (zinc-rich), the γ phase (aluminium-rich—now described as β) remaining unaffected. (2) The decomposed γ phase is severely attacked as a result of the very large number of grain boundaries produced by the eutectoid decomposition. (3) The grain-size of zinc–aluminium alloys decreases with increasing aluminium content, and in coarse-grained alloys (below 2% aluminium) the depth of penetration is greater, although expansion may be slight. (4) The severity of attack, as affecting the physical characteristics of a specimen, depends upon the grain-size and the presence or absence of impurities or other metals added to the alloy. (5) As for the effect of the normal impurities present on the rate and severity of attack, lead is a powerful accelerator, cadmium and tin almost as detrimental, iron is less so, and nickel and manganese are only moderately detrimental. Copper to the extent of 0·5% is a powerful retarding agent and in amounts up to 5·0% is more or less beneficial.

Seven years later, Anderson[2] carried out similar steam-ageing tests on nine zinc die-casting alloys of various compositions and found that one only was free from intercrystalline corrosion. This particular alloy had the composition: aluminium 4·0%, copper 3·0%, magnesium 0·13%, and iron 0·023%, the manganese, lead, cadmium, and tin each being below 0·01%.

Using two grades of metal, described respectively as electrolytic zinc (99·98% purity) and refined zinc (98·86% purity), Guertler and his co-workers[3] in 1935 carried out steam-ageing experiments with zinc alloys of the following range of compositions: (*a*) aluminium 0–4%, copper 2·0–8·0%, and (*b*) aluminium 2·0%, copper 4·0%, manganese 1·0%.

It was found that all alloys showed poor resistance to corrosion by salt solution and that all underwent deterioration when tested in steam at 95°–100° C., those made from "refined zinc" being most seriously affected in both series of tests.

In 1940 Petrich and Wolf[4] published results obtained from tests on two series of zinc alloys (one containing 9·0% aluminium and 0·009–0·31% lead, the other 0·09% lead and 0·28–9·95% aluminium), which were examined after exposure for six days to water vapour at 95° C. Both series were exposed in the unstrained and compressed conditions, corrosion being more pronounced in the latter case. Even in the alloy with the lowest lead content, hairline cracks were visible after six days' steam-ageing, and it was observed that, in general, corrosion followed the grain boundaries with (presumably electrolytic) solution of the aluminium-rich phase.

Between 1941 and 1947 Löhberg published the results of his researches on the question in six papers.[5–10] The purity of the zinc used in preparing the alloys was not always quoted, but in the first series of experiments there was present, in addition to the alloying elements, 0·02% lead, 0·002% cadmium, and 0·03% iron. The alloys tested were within the range 1·1–21·6% aluminium, 0·65–1·1% copper, 0·03–0·2% magnesium, and 0·02%–0·70% lead, but some of them presumably also contained small quantities of cadmium, thallium, tin, calcium, and sodium.

The standard steam-ageing testing procedure was used, and the conclusions reached may be summarized as follows: (1) The depth of penetration of the corrosion was found to be dependent upon the structure, being greatest with those of coarse grain. (2) Corrosion takes place only when the aluminium-rich (β) phase is present; it affects only this phase, and is intercrystalline only when this phase is concentrated along the grain boundary. (3) The degree of corrosion increases with increasing aluminium content, the maximum increase in weight and length as a result of corrosion being shown by alloys of eutectoid composition. (4) Corrosion is increased by the presence of lead, cadmium, thallium, and particularly tin, and is decreased by copper, calcium, sodium, and magnesium. The adverse effect of the first group of impurities is alleviated to some extent by the addition of magnesium.

Experimenting later with alloys of eutectoid composition containing in addition 0·4 and 0·7% lead, with and without 0·2% magnesium, Löhberg found that the material of higher lead content was completely disintegrated after two days' steam-ageing, whereas the material with 0·4% lead took three or four days to reach a similar state. He stated that the corrosion-resistance of the magnesium-containing alloy was greater than that of the magnesium-free alloy and suggested that this was due to the conversion of the lead to the compound Mg_2Pb, which, although intercrystalline, was more finely dispersed and evenly distributed. He also concluded that the decrease in grain-size produced by the magnesium additions contributed to the improvement in corrosion-resisting properties of these alloys.

Löhberg's fifth paper,[9] which is on the subject of hydrogen evolution during the process of corrosion, throws little further light on the subject, except for a remark to the effect that if magnesium is present in the alloy it has an inhibiting effect due to the formation of a protective layer on the zinc.

The last paper in the series[10] concerns the effect of temperature on the rate of corrosion both by water and by water vapour. The results obtained showed that the rate of corrosion increases (1) with increasing aluminium content of the alloy, (2) with increase in temperature—at 95° C. it is 100–200 times that at room temperature, and (3) with increasing grain-size.

In a more recent paper,[11] Löhberg has reported on the action of steam and water on compounds of aluminium with antimony, copper, and magnesium and of magnesium with lead, tin, silicon, bismuth, antimony, cadmium, zinc, and copper, some of which compounds may occur in Mazak alloys. Magnesium plumbide (Mg_2Pb), which was found to be much more rapidly attacked than the other magnesium compounds, was completely decomposed in 4 hr. by steam at 95° C., while in air at 20° C. it was completely oxidized in 49 days. The compound $MgZn_2$ showed no change in weight when exposed to air for 287 days.

It was concluded that an intermetallic compound is attacked by water or steam only if the components have widely different electrochemical potentials and that the attack can be halted by a protective oxide layer or by a protective layer of the nobler metal left after selective removal of the other component from the surface.

A number of other papers on the intercrystalline corrosion of zinc die-casting alloys have been examined, but the information contained in them was not found to be very relevant to the present investigation.

III.—EXPERIMENTAL WORK

The materials used in the present investigation were: Special high-purity zinc (99·999+ % pure); Crown Special zinc (B.S. 1003) (99·99+ % pure, the impurities being lead 0·0015%, cadmium 0·001%, iron 0·001%, and copper a trace); and aluminium of 99·98% purity. The other metals were the purest available.

The object of the research was to investigate generally the occurrence of intercrystalline corrosion in zinc–aluminium alloys. The scope of the investigation may be summarized thus: (*a*) to determine whether single-phase zinc-rich alloys are susceptible to intercrystalline corrosion; (*b*) to determine the extent to which two-phase alloys are attacked; (*c*) to examine the effect of other elements in alloying proportions on the susceptibility of these alloys to intercrystalline attack; (*d*) to examine the effect of impurity elements on the rate and intensity of intercrystalline attack; and (*e*) to assess the efficiency of magnesium in suppressing the intercrystalline attack in such alloys.

Preliminary experiments made in these laboratories [12] indicated that some intercrystalline corrosion occurs with single-phase zinc-rich alloys when there are no impurities present to act as cathode (assuming that the phenomenon is electrochemical in character), and that no corrosion takes place when the single-phase aluminium-rich alloy is tested under similar conditions.

To confirm these findings, and to extend the investigation, alloys numbered 1 to 10, of composition shown in Table I, were prepared and gravity die-cast in an iron mould to produce bars $\frac{3}{8}$ in. square. The test-pieces thus prepared were subjected to the standard steam test, i.e. ten days in an air/water-vapour atmosphere maintained at 95° ± 2° C. When this treatment was completed, the test-pieces were withdrawn and the surface condition noted. Two complete cross-sections were cut from each sample, prepared in the normal manner, and examined at high magnification. The observations made on them are reported in column 3 of Table I.

In the light of the information obtained from this examination, alloys numbered 11–26 and 32–101, of composition shown in Table I, were prepared and subjected to the same steam-ageing treatment. Several commercial pressure die-castings in Mazak 3 were also included (27–31 of Table I). The information obtained from the subsequent metallographic examination of sections cut from all these test specimens also appears in Table I, column 3.

Figs. 1–9 (Plates XLIII and XLIV) show the effects of impurities and alloy composition on the character and extent of the corrosion. Fig. 1 depicts a typical field at the edge of alloy 2 (0·075% aluminium), the path of the corrosion round the grains being comparatively narrow. Alloy 6 contained 0·003% lead, the aluminium content being similar to that of alloy 2. The path of the corrosion in this case was somewhat wider, as shown in Fig. 2, but the depth of penetration was very little greater.

The corrosion at the edge of alloy 9 (0·04% magnesium, 0·013% lead) is shown in Fig. 3; attack appears to have been very intense and the path of the corrosion comparatively wide.

The type of intercrystalline corrosion observed in alloy 26 (4·44% aluminium, 0·048% magnesium), shown in Fig. 4, is very similar in appearance to that which can be induced in the standard pressure-die-cast Mazak 3 (Fig. 5).

Figs. 6 and 7 illustrate the effects of copper and copper with magnesium on the extent of the intercrystalline attack. The former is of a field in a specimen of alloy 52, containing 4·0% aluminium and 0·3% copper; the latter shows a field in a specimen of alloy of similar composition but containing in addition 0·05% magnesium (alloy 57).

The effect on the intercrystalline attack of adding 0·01% cadmium to an alloy consisting of 4% aluminium in high-purity zinc (alloy 68) is shown in Fig. 8. This should be compared with Fig. 9, which is of a similar alloy free of cadmium (alloy 38).

The photomicrographs also illustrate the degree and mode of occurrence of the intercrystalline attack in the various samples, as indicated in column 3 of Table I. The terms " severe " and " intense " indicate the type of attack illustrated in Figs. 3 and 5. Fig. 5 also illustrates intercrystalline corrosion occurring as a continuous layer, as opposed to individual patches as shown in Fig. 4. Fig. 6 shows corrosion of average intensity, and in Fig. 9 the corrosion is about midway between average and severe. The term " spine " is used to describe corrosion running along a single grain-boundary path; the beginning of a prominent example is shown in the lower centre of Fig. 2.

Table I.—*Summary of Experimental Results.*

Alloy No.	Composition, %	Occurrence, Severity, and Depth of Penetration of Intercrystalline Corrosion
1	Zinc (99·999%)	Nil.
2	0·075 Al	General attack to 0·008 in.; individual spines to 0·024 in.
3	0·15 Al	General attack to 0·016 in.; individual spines to 0·040 in.
4	0·27 Al	General attack to 0.040 in.
5	0·003 Pb	Nil.
6	0·075 Al, 0·003 Pb	To a general depth of 0·013 in.; individual spines to 0·040 The corrosion in these samples was of a more intense nature than that observed in the corresponding lead-free samples 2, 3, and 4.
7	0·165 Al, 0·011 Pb	To a general depth of 0·060 in.
8	0·27 Al, 0·0075 Pb	To a general depth of 0·060 in.
9	0·04 Mg, 0·013 Pb	Very intense to 0·040 in.; less intense to 0·20 in.
10	0·038 Mg	Very occasional spines to 0·002 in.
11	96·8 Al, 3·2 Zn	Nil.
12	79·97 Al, 20·03 Zn	Nil.
13	80·27 Al, 19·73 Zn, 0·001 Pb	Nil.
14	0·24 Al, 0·016 Mg	Occasional patches to 0·004 in.
15	0·28 Al, 0·014 Mg, 0·0063 Pb	To a general depth of 0·10 in.; less intense than that observed in No. 9.
16	2·00 Al, 0·051 Mg, 0·006 Pb	General attack to 0·002 in.
17	0·96 Al, 0·045 Mg, 0·0064 Pb	General to 0·004 in.; individual spines to 0·024 in.
18	0·095 Al, 0·047 Mg	Little or no corrosion observed.
19	0·046 Al	General to 0·012 in.
20	22·26 Al, 0·0063 Pb	General to 0·008 in.; patches to 0·10 in.; hairline cracks visible on surface of test-piece.
21	0·61 Al	Attack to 0·007 in.; to 0·020 in. in places.
22	1·83 Al	More intense to 0·010 in.; up to 0·016 in. in places.
23	4·48 Al	Fairly intense to 0·010 in.; to 0·028 in. in places.
24	22·40 Al	Intense to 0·010 in.; one patch to 0·080 in.
25	22·52 Al, 0·045 Mg	Intense to 0·005 in.; one patch to 0·100 in.
26	4·44 Al, 0·048 Mg	Patches of fairly intense intercrystalline corrosion to 0·003 in. Certain areas of the surface unattacked.
27	Gravity die-cast Mazak 3	Very occasional patches to 0·003 in.
28	Pressure die-cast Mazak 3, stored one year	Very frequent patches to 0·004 in.
29	Pressure die-cast Mazak 3, stored six months	Less frequent patches to 0·004 in.
30	Die-casting in Mazak 3	Fairly general to 0·003 in.
31	Die-casting in Mazak 3	Patches to 0·003 in.; somewhat more intense than No. 30.
32	S.P. Zn + 0·08 Al + 0·021 Mg	Fairly frequent spines to 0·001 in.; occasional spines to 0·004 in.
33	S.P. Zn + 0·08 Al + 0·010 Mg	Fairly frequent spines to 0·005 in.; a few to 0·020 in.
34	S.P. Zn + 0·04 Al + 0·003 Mg	Somewhat less frequent spines to 0·006 in.; a few to 0·012 in.
35	S.P. Zn + 0·04 Al + 0·0015 Mg	Somewhat less frequent spines to 0·006 in.; no deeper penetration observed.
36	S.P. Zn + 0·02 Al	Occasional spines to 0·002 in.; a few to 0·005 in.
37	S.P. Zn + 0·005 Al	Nil.
38	S.P. Zn + 4 Al	Layer to 0·015 in. max.; average 0·007 in.
39	S.P. Zn + 4 Al + 0·001 Mg	Layer to 0·009 in. max.; average 0·005 in.
40	S.P. Zn + 4 Al + 0·009 Mg	Layer to an average depth of 0·001 in.; moderately frequent patches to 0·005 in.
41	S.P. Zn + 4 Al + 0·026 Mg	Moderately frequent patches to 0·005 in. Much of the surface unattacked.
42	S.P. Zn + 4 Al + 0·047 Mg	Moderately frequent patches to 0·005 in. Much of the surface unattacked.
43	S.P. Zn + 4 Al + 0·057 Mg	As No. 41.
44	S.P. Zn + 4 Al + 0·081 Mg	Moderately frequent patches to a depth of 0·003 in.
45	S.P. Zn + 4 Al + 0·090 Mg	Moderately frequent patches to a max. depth of 0·004 in.
46	S.P. Zn + 4 Al + 0·05 Mn	Intense to 0·005 in. Frequent spines to 0·050 in.
47	S.P. Zn + 4 Al + 0·05 Mn + 0·05 Mg	Comparatively infrequent patches to 0·003 in.
48	C.S. Zn + 4 Al	Intense to 0·026 in.
49	C.S. Zn + 4 Al + 0·05 Mg	Fairly frequent patches to 0·003–0·004 in.
50	C.S. Zn + 4 Al + 0·05 Cu	Intense to 0·040 in.
51	C.S. Zn + 4 Al + 0·10 Cu	Very frequent (almost continuous) patches to 0·010 in.
52	C.S. Zn + 4 Al + 0·30 Cu	Very frequent (almost continuous) patches to 0·005–0·006 in.
53	C.S. Zn + 4 Al + 0·80 Cu	Very frequent (almost continuous) patches to 0·008 in.
54	C.S. Zn + 4 Al + 1·5 Cu	Very frequent patches to 0·006–0·008 in.
55	C.S. Zn + 4 Al + 0·05 Mg + 0·05 Cu	Comparatively frequent patches to 0·003 in. Pits to the same depth also observed.
56	C.S. Zn + 4 Al + 0·05 Mg + 0·10 Cu	Fairly frequent pits and patches to 0·001 in.; a few to 0·003–0·004 in.
57	C.S. Zn + 4 Al + 0·05 Mg + 0·30 Cu	Fairly frequent pits and patches to 0·001 in.; a few to 0·003–0·004 in.
58	C.S. Zn + 4 Al + 0·05 Mg + 0·80 Cu	Fairly frequent pits and patches to 0·001 in.; a few to 0·003–0·004 in.
59	C.S. Zn + 4 Al + 0·05 Mg + 1·5 Cu	Fairly frequent pits and patches to 0·001 in.; a few to 0·003–0·004 in.
60	C.S. Zn + 0·004 Pb + 0·05 Mg	Intense throughout the sample.
61	C.S. Zn + 0·005 Sn + 0·05 Mg	Attack of considerably less intensity to 0·10 in.
62	C.S. Zn + 0·005 Cd + 0·05 Mg	Attack to a depth of 0·025 in.; intensity as No. 61.
63	S.P. Zn + 4 Al + 0·05 Mg + 0·002 Cd	Occasional patches to 0·001 in.
64	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Cd	Frequent pits and patches to 0·002 in.
65	S.P. Zn + 4 Al + 0·05 Mg + 0·008 Cd	Frequent patches to 0·005 in. max.
66	S.P. Zn + 4 Al + 0·05 Mg + 0·014 Cd	A layer to 0·003 in.

TABLE I.—*continued.*

Alloy No.	Composition, %	Occurrence, Severity, and Depth of Penetration of Intercrystalline Corrosion
67	S.P. Zn + 4 Al + 0·05 Mg + 0·025 Cd	A layer to 0·006 in.
68	S.P. Zn + 4 Al + 0·010 Cd	Attack to 0·012 in.
69	S.P. Zn + 4 Al + 0·05 Mg + 0·001 Sn	Patches to 0·002 in. average.
70	S.P. Zn + 4 Al + 0·05 Mg + 0·002 Sn	Patches to 0·002 in. average.
71	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Sn	Patches to 0·003 in. average.
72	S.P. Zn + 4 Al + 0·05 Mg + 0·008 Sn	Patches to 0·003 in. average.
73	S.P. Zn + 4 Al + 0·05 Mg + 0·014 Sn	Frequent patches to 0·007 in.; more severe than Nos. 70–72.
74	S.P. Zn + 4 Al + 0·05 Mg + 0·025 Sn	A layer to 0·007 in. max.; more severe than Nos. 70–72.
75	S.P. Zn + 4 Al + 0·010 Sn	Intense to 0·050 in.
76	S.P. Zn + 4 Al + 0·05 Mg + 0·0015 Pb	Patches to 0·004 in.
77	S.P. Zn + 4 Al + 0·05 Mg + 0·003 Pb	Patches to 0·004 in.
78	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Pb	More frequent patches to 0·004 in.
79	S.P. Zn + 4 Al + 0·05 Mg + 0·030 Pb	Frequent patches to 0·006 in.; more severe than Nos. 76–78.
80	S.P. Zn + 4 Al + 0·05 Mg + 0·020 Pb	Patches to 0·004 in.; considerably less frequent than No. 79 but more severe than Nos. 76–78.
81	S.P. Zn + 4 Al + 0·008 Pb	Intense to 0·040 in.
82	S.P. Zn + 0·020 Cd	Nil.
83	S.P. Zn + 0·020 Sn	Nil.
84	S.P. Zn + 0·5 Cd	Nil.
85	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Sn + 0·004 Pb	Fairly general to 0·001 in.; comparatively frequent patches to 0·006 in.
86	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Sn + 0·005 Cd	Fairly general to 0·001 in.; less frequent patches to 0·004 in.
87	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Sn + 0·005 Bi	Comparatively infrequent patches to 0·003–0·004 in.
88	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Pb + 0·005 Cd	Occasional patches to 0·003 in.
89	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Pb + 0·005 Bi	Very frequent patches to 0·001–0·002 in.; occasional patches to 0·005 in.
90	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Cd + 0·005 Bi	Fairly frequent patches to 0·001–0·002 in.; occasionally to 0·005 in.
91	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Sn + 0·003 Pb + 0·004 Cd	Somewhat less frequent patches to 0·003–0·004 in.
92	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Sn + 0·003 Pb + 0·003 Bi	Fairly infrequent patches to 0·003 in.
93	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Sn + 0·004 Cd + 0·003 Bi	Very frequent patches to 0·006 in.
94	S.P. Zn + 4 Al + 0·05 Mg + 0·003 Pb + 0·004 Cd + 0·003 Bi	Very frequent patches to 0·001–0·002 in.; one large patch 0·050 in. deep.
95	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Sn + 0·004 Pb + 0·004 Cd + 0·004 Bi	Frequent patches to 0·005–0·006 in.
96	S.P. Zn + 4 Al + 0·05 Mg + 0·002 Bi	Comparatively infrequent patches to 0·004 in. max.
97	S.P. Zn + 4 Al + 0·05 Mg + 0·004 Bi	Comparatively infrequent patches to 0·004 in. max.
98	S.P. Zn + 4 Al + 0·05 Mg + 0·005 Bi	Comparatively infrequent patches to 0·004 in. max.; one patch 0·010 in. deep.
99	S.P. Zn + 4 Al + 0·05 Mg + 0·010 Bi + 0·020 Pb	Patches decidedly more frequent and intense than No. 96. Depth of penetration not much greater.
100	S.P. Zn + 4 Al + 0·05 Mg + 0·012 Bi	Attack somewhat less intense than that found in No. 99.
101	S.P. Zn + 4 Al + 0·05 Mg + 0·016 Bi	Attack somewhat less intense than that found in No. 99.

In view of the suggestion made by Löhberg[8] that the action of magnesium is to "fix" the lead as the intermetallic compound Mg_2Pb and because the alloy (No. 9 in the present series), which probably contained this compound as an intercrystalline constituent, was found to offer poor resistance to corrosion, a lead–magnesium alloy of composition corresponding to the compound was prepared and cast. It was found that the material was extremely brittle and disintegrated rapidly when exposed to moist air at room temperature.

IV—SUMMARY OF RESULTS

From a consideration of the results obtained in this series of experiments the following points emerge :

(1) Single-phase zinc–aluminium alloys of aluminium content between 0·08 and 0·02% are susceptible to intercrystalline corrosion when exposed to the air/water-vapour atmosphere at 95° C. The depth of penetration of the attack decreases with decreasing aluminium content, and it appears that no attack occurs when the aluminium present is much below about 0·01%. The corrosion becomes much more severe when lead is present. In the absence of aluminium, zinc, either pure or contaminated with lead, is immune from intercrystalline attack. (Alloys 2, 5, 6, 19, 36, and 37.)

(2) The intercrystalline attack in these single-phase zinc–aluminium alloys appears to be almost completely suppressed by the addition of about 0·05% magnesium, but some attack takes place if the amount of magnesium present is much less than 0·03%. (Alloys 18 and 32–35.)

(3) The aluminium-rich (β) solid solution is not susceptible to intercrystalline attack by water vapour at 95° C. (Alloys 11, 12, and 13.)

(4) In two-phase zinc–aluminium alloys free from magnesium, it appears that the depth of penetration of the intercrystalline attack does not increase greatly with increasing aluminium content, but the intensity of attack is greater in the alloys of higher aluminium content. The presence of small percentages of lead (0·006–0·013%) in these alloys increases the severity

of the intercrystalline corrosion, but the simultaneous presence of about 0·05% magnesium causes a considerable reduction in the severity of the attack. The behaviour of alloy 15 was, however, anomalous. (Alloys 3, 4, 7, 8, 15, 16, 17, 20, 21, 22, 23, 24, and 25.)

(5) The resistance to intercrystalline attack of two-phase zinc–aluminium alloys is improved slightly by the addition of as little as 0·001% magnesium. Maximum resistance is obtained by the addition of 0·02–0·03% magnesium, but further additions of this element up to 0·09% appear to confer no significantly greater benefit. (Alloys 14, 25, 26, and 38–45.)

(6) Although Mazak die-castings of composition within the limits prescribed by British Standard 1004 show intercrystalline corrosion when subjected to steam-ageing treatment as defined in that specification, the depth of penetration is very small and is not likely to be greater than 0·004 in. This, though quite detectable, is inappreciable from a practical standpoint, since the conditions to which die-castings are normally exposed in service are far less severe than those in the steam tank. (Alloys 27–31.)

(7) The addition of up to 0·30% copper to zinc–4% aluminium alloys is beneficial with regard to resistance to intercrystalline corrosion. No further improvement is obtained with increasing percentages up to 1·5%. (Alloys 48 and 50–54.)

(8) When 0·05% magnesium is also present in these alloys, the addition of 0·10–1·5% copper slightly improves the resistance to intercrystalline corrosion, but in no case is the attack completely suppressed. (Alloys 49 and 55–59.)

(9) No intercrystalline corrosion occurs when small quantities of magnesium, lead, tin, or cadmium are present alone in pure zinc. When, however, a small percentage of magnesium is present, together with a small percentage of lead, tin, or cadmium, severe intercrystalline corrosion occurs, the severity of the attack being greatest with lead and least with cadmium. (Alloys 5, 9, 10, 60, 61, 62, 82, 83, and 84.)

(10) The presence in alloys of Mazak 3 composition of up to about 0·005% lead, tin, cadmium, or bismuth does not appear to modify greatly the susceptibility of these alloys to intercrystalline attack. The severity of the attack increases somewhat with increasing amounts of these impurities up to about 0·025%, the maximum figure used in the present series of experiments.

In the absence of magnesium the presence of small amounts of lead or tin greatly increases the severity and depth of penetration of the attack, whereas with a similar amount of cadmium present alone the severity of the attack is increased to a much lesser extent. (Alloys 63–81 and 96–101.)

(11) In the same respect, the presence (in alloys containing 4% aluminium + 0·05% magnesium) of small percentages of lead, tin, cadmium, and bismuth in various combinations has an effect similar in magnitude to that of the most potent element when present in amounts equal to the total of the impurity elements. (Alloys 85–95.)

(12) When manganese is present in the alloy as an impurity element, its effect on the incidence of intercrystalline corrosion is deleterious, but less so than that of lead or tin. The evidence is incomplete however. (Alloys 46–47.)

(13) The intermetallic compound Mg_2Pb is unstable in moist air at room temperature.

In the group of alloys 32–35 (single-phase zinc–aluminium alloys with decreasing amounts of magnesium), the aluminium content of the first two alloys was 0·08%, whereas in the latter two it was 0·04%. The results obtained from this series, therefore, are not directly comparable. Alloys 48–62 were prepared from " Crown Special " zinc; the remainder from special high-purity metal (99·999%). Alloy 50 contained 0·0005% tin; in the remainder of the alloys in that group tin was not detected.

V.—DISCUSSION OF RESULTS

An air/water-vapour atmosphere maintained at 95° ± 2° C. has been used for some thirty years as a medium for testing the susceptibility of zinc die-casting alloys to intercrystalline corrosion. The results obtained should, however, be viewed in proper perspective, since the conditions involved are far more searching than those to which commercial die-castings are exposed in normal use. Nevertheless, apart from the intergranular attack, the general corrosive action of steam and boiling water on zinc and zinc–aluminium alloys is comparatively mild, the surface appearance of many of the test-pieces that suffered from intercrystalline corrosion being reasonably good.

Intercrystalline corrosion of a single-phase, or ostensibly single-phase, alloy is usually explained as an electrochemical phenomenon. Briefly and simply, the small amounts of insoluble impurities present segregate at the grain boundaries, and, if widely separated from the base metal or solid solution in electrochemical characteristics, they are liable to become poles of a galvanic couple.

In mildly corrosive conditions (insufficient to attack rapidly, if at all, the base metal or solid solution), the galvanic couples become active and, depending on the polarity, the intercrystalline film of impurities or the outer layer of the adjacent crystal is dissolved, with the formation of a void between neighbouring crystals.

The evidence obtained from the present investigations conflicts with the view that this mechanism is the primary cause of intercrystalline corrosion in zinc–aluminium alloys. Zinc free from aluminium but containing appreciable amounts of impurities does not suffer intercrystalline attack, whereas when aluminium is present in solid solution and the total impurity content is less than 0·001% the incidence of intercrystalline corrosion is considerable. Furthermore, the small amount of aluminium present in the zinc does not endow it with an electrode potential very different from that of pure zinc.

It appears, therefore, that the susceptibility of the single-phase alloy to intercrystalline attack is associated with the presence of aluminium in solution. The solubility of aluminium in zinc at normal temperature is low; it decreases from about 0·62% at 250° C. to 0·12% at 95° C. and 0·03% at 25° C. The solid solution that first forms on freezing is not stable, and the excess aluminium is precipitated at room temperature, rapidly in the initial stages and then at a decreasing rate, so that after about two years' ageing the aluminium content of the zinc-rich phase is reduced to about 0·03%, irrespective of the original aluminium content of this phase. The fact that precipitation takes place at room temperature suggests that the aluminium atoms within the zinc lattice possess considerable mobility, and since under ideal conditions precipitation takes place at grain boundaries, it seems probable that a tendency exists for the aluminium atoms to diffuse in the direction of grain boundaries.

Aluminium as an element has a high affinity for oxygen, but when exposed in bulk in an oxidizing environment, the strongly adherent film of oxide which forms immediately on the surface protects the remainder of the metal from further attack. In the case of dilute solutions of aluminium in zinc, the aluminium atoms near the surface are susceptible to attack by oxygen, though they are too widely spaced to allow of the formation of a protective oxide film. Further, such a film, if formed, would be mechanically weak, since the cubic Al_2O_3 would not fit closely on to the exposed faces of the hexagonal zinc crystals. It appears possible, therefore, that these highly reactive atoms near the grain surface catalyse the oxidation of zinc to ZnO by virtue of their high affinity for oxygen and their inability to form a stable oxide film.

If the " grain-boundary-seeking tendency " of the aluminium atoms is assumed, the atoms within the grain will diffuse to the surface (to maintain even distribution), and the process of oxidation will continue until the grains near the specimen surface are surrounded by a continuous envelope of oxide. There is, however, little direct experimental evidence available in support of such a theory, and it is put forward with some reserve to explain the striking influence on the behaviour of zinc of a small proportion of aluminium in solid solution.

It having been established that intercrystalline corrosion takes place in single-phase alloys, little further need be said on the subject of the phenomenon in two-phase alloys. The increase in the intensity of the attack is probably due to the fact that the second (β) phase appears at the grain boundaries of the primary crystals as a eutectoid with the α phase. This fine-grained lamellar eutectoid increases greatly the effective area of the α-phase boundaries and consequently the intensity of the attack. It is clear, however, that the view expressed by Löhberg that " corrosion takes place only when the aluminium-rich phase is present; it affects only this phase and is intercrystalline only when this phase is concentrated along the grain boundary ", is incorrect. Intercrystalline corrosion takes place in single-phase (α) alloys; the intensity of the attack is increased, but the attack is not initiated by the presence at the grain boundaries of the aluminium-rich (β) phase.

Regarding the action of the impurity elements lead, tin, cadmium, and bismuth, it was found that, with magnesium present also, the intensity of the intercrystalline attack was increased considerably but the depth of penetration only slightly. In the absence of magnesium the effect of lead and tin was very marked; cadmium appeared to be comparatively innocuous, however.

The action of these impurities is usually explained in terms of the electrochemical theory of corrosion, the impurities acting as cathode with the resulting dissolution of adjacent areas of zinc or zinc-rich phase in the presence of a mildly corrosive medium. Since the impurities segregate at the grain boundaries, the corrosion is of the intercrystalline type.

The standard electrode potentials of the four metals zinc, lead, tin, and cadmium are as follows (hydrogen = 0·000): lead 0·126, tin 0·136, cadmium 0·402, zinc 0·762 V.[13] The comparative accelerating effect of these elements on corrosion was in the order to be expected from these figures, though the conditions are not comparable in all cases, since lead and tin are present as intercrystalline constituents whereas cadmium is in solution in the zinc. The fact that the least harmful impurity is in solution is, however, noteworthy.

Zinc itself is immune from intercrystalline corrosion even in the presence of metals such as lead which could act as cathode. When some aluminium is in solid solution, however, zinc is susceptible to intercrystalline corrosion, though when no impurities are present the rate of attack is less. The presence of lead or tin at the grain boundaries greatly accelerates the rate of attack in solid solutions of aluminium in zinc, presumably because these metals act as cathodes. In this electrochemical process, therefore, it appears that solid solutions of aluminium in zinc can act as the anode, but pure zinc cannot.

It is possible that the action of these impurity elements is also mechanical in that their presence, in eutectic form, at the grain boundaries of the α solid solution increases the effective area of the surface of this phase, and thus favours the progress of the intercrystalline attack. Since cadmium is in solid solution, it is ineffective in this respect.

It has been shown that the addition of 0·03–0·05% magnesium to zinc–aluminium alloys suppresses almost completely intercrystalline corrosion in single-phase alloys. The action of magnesium in preventing or minimizing intercrystalline corrosion in these alloys is difficult to explain. Löhberg, who assumed that lead was the primary instigator of the corrosion, suggested in one of his papers that the lead was converted to the intermetallic compound Mg_2Pb, which, although still intercrystalline, was apparently innocuous. He also stated that the reduction in grain-size caused by the magnesium addition contributed to the improved resistance to corrosion of these alloys. In a later paper[9] he states that when magnesium is present in these (zinc–aluminium) alloys it has an inhibiting effect owing to the formation of a protective layer on the zinc.

The first suggestion has been shown to be incorrect. The intermetallic compound Mg_2Pb was found to be unstable in moist air at room temperature (a fact since confirmed by Löhberg), and a specimen of high-purity zinc in which magnesium and lead were present together was severely corroded after ten days' exposure to steam.

The improved resistance to corrosion produced by a smaller grain-size is possibly due to there being available near the surface of the specimen a much larger area of grain surface. This would reduce the chances of the corrosion following a "grain-boundary path" continuously normal to the specimen surface, and consequently the corrosion would tend to occur as patches adjacent to the surface rather than as isolated spines of greater depth.

Regarding Löhberg's third suggestion, the solubility of magnesium in zinc at 360° C. is 0·11%, whereas at room temperature it decreases to about 0·002% and it is probable that the solubility of this element in the α phase at corresponding temperatures is very little higher. When the single-phase alloy cools, the excess magnesium is precipitated at the grain boundaries, probably as the intermetallic compound $MgZn_5$, and may form a fairly continuous network enclosing the grains of the zinc-rich phase. The water vapour is thus prevented from coming into contact with this phase, and it has been shown that this intermetallic compound $MgZn_5$ is itself virtually unattacked by water vapour. Magnesium is more soluble in the aluminium-rich solution, and when this phase is present the magnesium content of the α phase may be so reduced that insufficient magnesium is precipitated to form a continuous envelope. Under these conditions the α phase is susceptible to some extent to attack by water vapour, but sufficient $MgZn_5$ is available to hinder progress of the attack.

The effect of the presence of copper on the resistance of zinc–aluminium alloys to intercrystalline corrosion is quite marked. In the alloy prepared from Crown Special zinc and free of magnesium and copper, the intercrystalline attack penetrated to a depth of 0·026 in. The presence of 0·30% copper reduced the depth of penetration to 0·006 in., but no improvement was obtained by increasing further the copper content.

When the standard amount of magnesium (0·05%) was present in addition, the presence of copper up to 0·30% also improved somewhat the resistance to corrosion of these alloys, but in this case the margin available for improvement was more restricted, since in the copper-free alloy the attack penetrated only 0·003 in.

Both copper and magnesium are corrosion inhibitors, therefore, though their modes of action are probably very dissimilar. Magnesium is only slightly soluble in the α phase and is strongly electronegative, whereas copper has a substantial solubility in this phase at about 100° C. (0·7% approximately) and is electropositive to zinc. Thus, when copper is present in the zinc-rich solid solution, the potential of this phase is moved towards the positive side (i.e. nearer to that of lead and tin) and, in addition, its presence appears to increase the solubility of aluminium. The action of copper may therefore be to reduce the electrode potential between the α phase and the intercrystalline impurities—if, indeed, this circumstance is relevant—and to stabilize the phase, thus decreasing the tendency of the aluminium atoms to diffuse towards the grain boundaries.

There is one further relevant point in this connection. The presence of magnesium and copper in zinc–aluminium alloys is known to delay the eutectoid transformation, which normally takes place at 270° C. On cooling, therefore, this transformation takes place at lower temperatures and consequently the hexagonal (zinc-rich) phase separates from the β phase in granular form rather than in the normal lamellar form. These granules of the hexagonal phase are surrounded by the β phase, and are thus protected to a much greater extent from the action of the corrosive medium. The β phase (aluminium-rich) is not attacked by steam, and hence the decrease in the degree of the intercrystalline attack when either or both of these elements are present in the alloys.

The severe intercrystalline attack that takes place in pure zinc when a small amount of magnesium together with small amounts of lead or tin are present is not difficult to understand. All these elements have low solubilities in zinc and are precipitated at the grain boundaries, probably in the form of the intermetallic compounds Mg_2Pb and Mg_2Sn. The compound Mg_2Pb has been shown to be unstable in moist air, and Löhberg[11] has reported that Mg_2Sn is rapidly attacked by steam.

Cadmium, however, has considerable solubility in zinc, though the equilibrium between zinc, cadmium, and magnesium is not well known. Cadmium and magnesium are reported to form an intermetallic compound, $MgCd_3$, which if present at the grain boundaries would probably be susceptible to attack by steam. Löhberg[11] concluded that the intensity of the attack by steam on intermetallic compounds of this type varies with the difference between the electrochemical potentials of the components. The results obtained in the present experiments tend to confirm this conclusion, since the relevant data for the three alloys are as follows:

Compound Assumed to be Present	Electrode Potential Difference, V.	Depth of Penetration of Intercrystalline Corrosion, in.
Mg_2Pb . . .	2·249	0·187
Mg_2Sn . . .	2·239	0·100
$MgCd_3$ . . .	1·973	0·025
$MgZn_5$ . . .	1·613	Nil

This effect of magnesium is also of considerable interest from a practical point of view, since the accidental contamination by magnesium of commercial grades of zinc which contain small amounts of lead would probably result in rapid disintegration of the material if stored or used in damp conditions. The contingency is, however, remote, since magnesium is rarely, if ever, used alone as an alloying element in zinc.

ACKNOWLEDGEMENTS

The author wishes to express his thanks to the Directors of The Imperial Smelting Corporation, Ltd., for permission to publish the paper, and to colleagues in the Research Department for helpful advice and discussion. The analysis of the samples was carried out by Mr. G. Ludwell and Mr. R. Starr.

REFERENCES

1. H. E. Brauer and W. M. Peirce, *Trans. Amer. Inst. Min. Met. Eng.*, 1923, **68**, 796.
2. H. A. Anderson, *Proc. Amer. Soc. Test. Mat.*, 1930, **30**, (I), 318.
3. W. Guertler, F. Kleweta, W. Claus, and E. Rickertsen, *Z. Metallkunde*, 1935, **27**, 1.
4. G. Petrich and W. Wolf, *ibid.*, 1940, **32**, 412.
5. E. Andres and K. Löhberg, *ibid.*, 1941, **33**, 208.
6. K. Löhberg, *ibid.*, 1941, **33**, 213.
7. K. Löhberg, *ibid.*, 1942, **34**, 73.
8. K. Löhberg, *Metallforschung*, 1946, **1**, 65.
9. K. Löhberg, *ibid.*, 1946, **1**, 66.
10. K. Löhberg, *ibid.*, 1947, **2**, 230.
11. K. Löhberg, *Z. Metallkunde*, 1950, **41**, 56.
12. K. P. Scott, Private communication.
13. H. H. Uhlig, " The Corrosion Handbook." **1948**: New York (John Wiley and Sons, Inc.); London (Chapman and Hall, Ltd.).

NOTICE TO AUTHORS OF PAPERS FOR THE "JOURNAL"

AND CONTRIBUTORS TO DISCUSSIONS

1. **Papers will be considered for publication from non-members as well as members of the Institute.** They are accepted for publication in the *Journal* and not necessarily for presentation at any meeting of the Institute. MSS. should be addressed to The Editor of Publications, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

2. **Papers suitable for publication** may be classified as:

(*a*) Papers recording the results of original research.

(*b*) First-class reviews of, or accounts of progress in, a particular field.

(*c*) Papers descriptive of works methods, or recent developments in metallurgical plant and practice.

(*d*) Papers in classes (*a*), (*b*), and (*c*) above, previously published in languages other than English, French, German, or Italian, if of sufficient merit.

3. **Manuscripts and illustrations** should be submitted in duplicate. MSS. must be typewritten (*double-line spacing*) on one side of the paper only, and authors are requested to sign a declaration that neither the paper nor a substantial part thereof has been published elsewhere. Exceptions may be made in certain cases where a paper has been published in a language other than English, French, German, or Italian (see 2(*d*) above). MSS. not accepted are normally returned within 6 months of receipt.

In the interests of economy, all papers must be written as concisely as possible; in general, internal research reports are not in suitable form for publication as papers in the *Journal*. All but the simplest mathematical expressions should be written by hand, with capital and small letters clearly distinguished. Superscript and subscript letters should also be plainly indicated. Greek letters and special signs should be identified in the margin. For style, spelling, and abbreviations used, any recent issue of the *Journal* may be consulted.

4. **Synopsis.** Every paper must have a synopsis (not exceeding 250 words in length) which, in the case of a paper reporting original research, should state its objects, the ground covered, and the nature of the results. The synopsis will appear at the beginning of the paper, and should be in a form suitable for use by abstracting organizations. Extracts from a "Guide for the Preparation of Synopses" drawn up by the Abstracting Services Consultative Committee are reproduced below.

5. **References** must be collected at the end of the paper and must be numbered in the order in which they occur in the MS. Initials of authors must be given, and the Institute's official abbreviations for periodical titles (as used in *Metallurgical Abstracts*) should be employed, where known. References to papers should be set out in the style:

A. L. Dighton and H. A. Miley, *Trans. Electrochem. Soc.*, 1942, **81**, 321 (i.e. year, volume, page).

References to books should be in the following style:

C. Zener, "Elasticity and Anelasticity of Metals". Chicago: 1948 (University of Chicago Press).

6. **Illustrations.** Each illustration must have a number and description; only one set of numbers must be used in one paper, and it is desirable to number the half-tone illustrations consecutively, rather than to intersperse them with the line figures. The captions should be typed on a separate sheet.

The set of **line figures** sent for reproduction must be drawn (about twice the size to appear in the *Journal*) in Indian ink on smooth white Bristol board, good-quality drawing paper, co-ordinate paper, or tracing *cloth*, which are preferred in the order given. Co-ordinate paper, if used, must be blue-lined, with the co-ordinates to be reproduced finely drawn in Indian ink. Curves should be drawn boldly (i.e. at least twice the thickness of the frame). Experimental points should be indicated by open or closed circles, triangles, squares, &c. (preferably not crosses). Curves should be broken on each side of such symbols and plenty of allowance should be made for closing up in blockmaking. All lettering and numerals, &c., should preferably be in *pencil*, so that the Institute's standard lettering may be affixed, and ample margins must be left outside the framework of the figures to enable this to be done. The second set of line illustrations may be photostat copies.

Photographs must be restricted in number, owing to the expense of reproduction, and photomicrographs should be trimmed to the smallest possible of the following sizes consistent with adequate representation of the subject: 4 in. deep by 3 in. wide: 2 in. deep by 3 in. wide: 2 in. square. Magnifications of photomicrographs must be given in each case. Photographs for reproduction should be loose, not pasted down (and not fastened together with a clip, which damages them), and the figure number and author's name should be written on the back of each. Captions should be given to the photomicrographs, but these should be kept as brief as possible.

Because of the present high cost of printing and paper it is imperative that authors restrict illustrations (particularly photographs) to the absolute minimum deemed necessary to support their argument. Only in exceptional cases will illustrations be reproduced if already printed and readily available elsewhere.

7. **Tables or Diagrams.** Results of experiments, &c., may be given in the form of tables *or* figures, *but* (unless there are exceptional reasons) *not both*. Tables should bear Roman numbers, and each should have a heading that will make the data intelligible without reference to the text.

8. **Corrections.** A certain number of corrections in proof are inevitable, but any modification of the original text is to be avoided. Since corrections are very expensive, the Institute reserves the right to require authors to contribute towards their cost if the Editor deems them to be excessive. The Institute also reserves the right to require a contribution to the cost of remaking any block where this is necessitated by an error on the author's part.

9. **Reprints.** Individual authors are presented with a maximum of 25, and two or more authors with a maximum of 50 reprints from the *Journal*, without covers. Limited numbers of additional reprints can be supplied at the author's expense, if ordered before proofs are passed for press. (Orders should preferably be placed when submitting MSS.)

10. **Discussion.** Except in the case of special symposia, shorthand records of discussions are not taken at meetings. Written discussion may be submitted on any paper, preferably typewritten (*double-line spacing*). References should be given in the form of footnotes. Paragraphs 6 and 7 above are also applicable to such contributions. Reprints of discussion cannot be supplied to contributors.

GUIDE FOR THE PREPARATION OF SYNOPSES

(*As recommended by the Abstracting Services Consultative Committee*)

1. **Purpose.** The synopsis is not part of the paper; it is intended to convey briefly the content of the paper, to draw attention to all new information, and to the main conclusions. It should be factual.

2. **Style of writing.** The synopsis should be written concisely and in normal rather than abbreviated English. It is preferable to use the third person. Where possible use standard rather than proprietary terms, and avoid unnecessary contracting.

It should be presumed that the reader has some knowledge of the subject, but has not read the paper. The synopsis should therefore be intelligible in itself without reference to the paper; for example, it should not cite sections or illustrations by their numerical references in the text.

3. **Content.** The title of the paper is usually read as part of the synopsis. The opening sentence should be framed accordingly and repetition of the title avoided. If the title is insufficiently comprehensive, the opening should indicate the subjects covered. Usually the beginning of a synopsis should state the objective of the investigation.

It is sometimes valuable to indicate the treatment of the subject by such words as: brief, exhaustive, theoretical, &c.

The synopsis should indicate newly observed facts, conclusions of an experiment or argument and, if possible, the essential parts of any new theory, treatment, apparatus, technique, &c.

It should contain the names of any new compound, mineral species, &c., and any new numerical data, such as physical constants; if this is not possible, it should draw attention to them. It is important to refer to new items and observations, even though some are incidental to the main purpose of the paper; such information may otherwise be hidden, though it is often very useful.

When giving experimental results the synopsis should indicate the methods used; for new methods the basic principle, range of operation, and degree of accuracy should be given.

4. **References.** If it is necessary to refer to earlier work in the summary, the reference should always be given in full and not by number. Otherwise references should be left out.

When a synopsis is completed, the author is urged to revise it carefully, removing redundant words, clarifying obscurities, and rectifying errors in copying from the paper. Particular attention should be paid by him to scientific and proper names, numerical data, and chemical and mathematical formulæ.

THE CONSTITUTION OF CHROMIUM–MANGANESE ALLOYS BELOW 1000° C.*

1447

By W. B. PEARSON,† D.F.C., M.A., D.Phil., MEMBER, and
W. HUME-ROTHERY,‡ O.B.E., F.R.S., MEMBER

SYNOPSIS

The constitution of chromium–manganese alloys between 1000° and 525° C. has been studied by microscopical and X-ray methods. The σ phase undergoes a transformation at 980°–1005° C. (according to composition), and high-temperature X-ray photographs show that both the high-temperature, σ', and low temperature, σ, modifications have characteristic "σ" structures resembling that of the σ-iron–chromium phase. The phase appears to be stable down to 523° C., and no signs of a eutectoid decomposition could be found. The solid solubility of manganese in chromium diminishes markedly below 1000° C., the temperature of the $\sigma \rightleftharpoons \sigma'$ transformation. In the range 1000°–800° C. normal two-phase (α-Cr + σ) alloys are formed when the solubility limit is exceeded. Between 800° and 600° C., three-phase alloys are found consisting of α-Cr, σ, and small amounts of a phase denoted α'-Mn, whose crystal structure appears to be similar to that of α-manganese; the amount of this last phase diminishes gradually on prolonged annealing, and the stable equilibrium almost certainly involves two-phase (α-Cr + σ) alloys. Below 600° C., the solubility of manganese in chromium diminishes even more markedly, and three-phase (α-Cr + α'-Mn + small amounts σ) alloys are formed. With annealing periods of the order of 1–2 months at 600°–500° C., equilibrium conditions are not obtained, but it is thought that the α'-Mn phase has a composition in the region of Mn$_2$Cr. The lattice spacings of the solid solution of manganese in α-Cr have been determined.

I.—INTRODUCTION

THE equilibrium diagram of the system chromium–manganese was determined in 1949 by Carlile, Christian, and Hume-Rothery [1], and was shown to contain a wide solid solution of manganese in chromium followed by a phase denoted θ, which was shown later [2] to possess the same crystal structure as the σ phase in the system chromium–iron, and is therefore called σ in the present paper. The σ phase was shown to be formed by a peritectic reaction, and this was confirmed by Zwicker,[3] although in an earlier paper [4] this author had regarded the σ phase as involving a eutectic. The diagram of Carlile, Christian, and Hume-Rothery in the region 0–70 at.-% manganese was determined § only above 1000° C., although these authors noted that changes occurred on annealing at low temperatures, and this was confirmed by Zwicker,[3] who showed the solubility of manganese in chromium to diminish greatly below 1000° C. The two investigations were not, however, in exact agreement where they overlapped, and as the purity and exact composition of Zwicker's alloys were not indicated,‖ a further study of the system has been made, and has shown that non-equilibrium structures are produced when previously homogenized alloys are re-annealed at low temperatures. It is probable that conditions of true equilibrium will require annealing treatments of several years, and the present paper is submitted to describe the structures obtained after annealing periods of the order of 5–71 days.

II.—EXPERIMENTAL TECHNIQUE

The manganese and chromium used were high-purity, electrolytic, hydrogen-reduced metals, and were melted in thoria-lined alumina crucibles in an H.F. induction furnace, using slight modifications of the methods described by Carlile, Christian, and Hume-Rothery, and by Pearson and Hume-Rothery.[6] The ingots were homogenized by heating in hydrogen for more than 2 hr. at 15°–20° C. below the solidus. Subsequent annealing treatments were carried out by standard methods [1, 6] in sealed evacuated tubes. Powder X-ray-diffraction films were used to determine lattice spacings by standard methods. Phases were identified by drawing a file across the actual surface examined under a microscope, and where the specimen had been annealed and quenched from a high temperature, the filings were not annealed, because this resulted in loss of manganese. At lower temperatures filings were annealed in the ordinary way. Seventeen alloys in lump form, and eleven batches of annealed filings were analysed by Johnson, Matthey and

---

* Manuscript received 15 October 1952.

† Low-Temperature Solid-State Physics Department, National Research Council, Ottawa, Canada; formerly Inorganic Chemistry Laboratory, Oxford.

‡ Royal Society Warren Research Fellow, and University Lecturer in Metallurgical Chemistry, Oxford.

§ As pointed out later,[5] owing to a mistake in drawing, the phase boundaries in one diagram were extended to low temperatures where they had not been determined.

‖ The alloys do not seem to have been analysed, although loss of manganese is known to occur on melting and on annealing.

Co., Ltd., and no evidence of contamination was found.*

III.—LATTICE-SPACING DATA

The lattice spacing of pure chromium, containing 0·003% oxygen, was determined as 2·8786–2·8787 kX at 18° C., in good agreement with the value 2·8789 kX obtained by Carlile [7] for pure Hilger chromium in lump form. The lattice spacings of the α-Cr solid solution are given in Fig. 1, which shows the change of lattice spacing to be too small for the solubility curve to be determined by lattice-spacing methods. The abnormality at low concentrations is not understood, but is probably connected with the effect of traces of impurity. It is to be noted that Poole and Axon [8] have shown that a rather similar abnormality in aluminium-rich aluminium–magnesium alloys is connected with the presence of only 0·01% silicon as impurity.

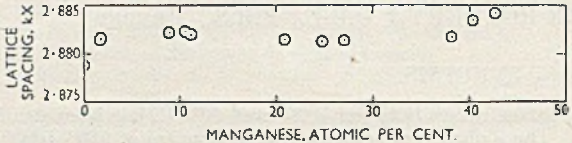


FIG. 1.—The Lattice Spacings of the Homogeneous Solid Solution of Manganese in Chromium. Alloy compositions were obtained by chemical analysis.

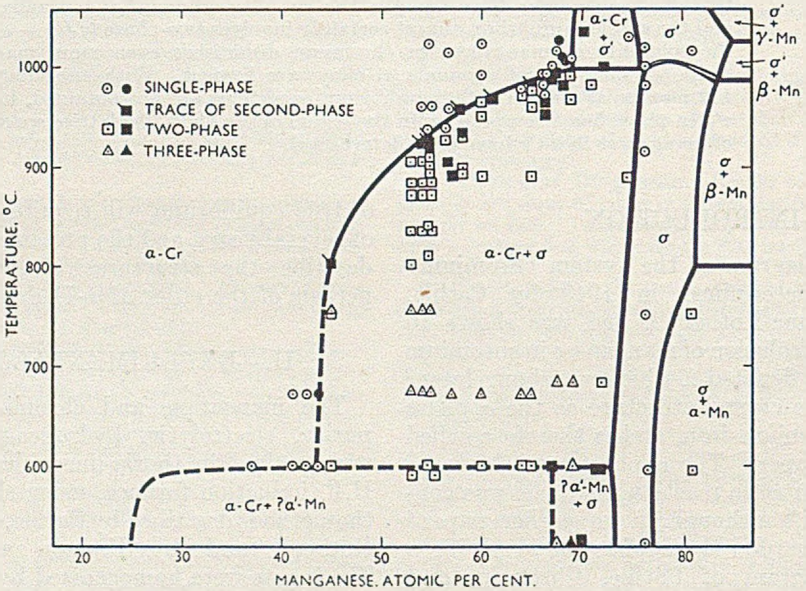


FIG. 2.—The Constitution of Chromium–Manganese Alloys. The full points refer to alloys which were analysed after the final annealing treatment. The open points refer to alloys which were not analysed, and are included only to show that the whole field was systematically examined. The compositions of these points are probably accurate to within ±1 at.-%, but they should not be used for critical comparison, because occasional larger variations were found.

IV.—EXPERIMENTAL RESULTS

The results obtained in the present work by the microscopical method are shown in Fig. 2. At 1000° C. the solubility of manganese in α-chromium is 69 at.-%, in agreement with the value of Carlile, Christian, and Hume-Rothery, but in contradiction to the value 65 at.-% given by Zwicker,[3] although the latter's microstructures could be reconciled with the present results.

Below 1000° C. the solubility of manganese in chromium diminishes markedly, owing to a transformation in the σ phase with which the α-Cr phase is in equilibrium. The transformation gives rise to well-defined thermal arrests, and the temperature horizontals on the chromium-rich and manganese-rich sides of the σ phase were determined as 998° and 981° C., respectively, the transformation rising to a maximum temperature of 1005° C. in the middle of the σ field. Photographs obtained by a high-temperature X-ray camera showed that both the σ' (high-temperature) and σ (low-temperature) phases possess typical " σ " structures, but the nearly equal scattering powers of chromium and manganese prevented any interpretation of the change.

From 1000° to 800° C. the solubility of manganese in chromium diminishes markedly, and when the solubility limit is exceeded, normal two-phase alloys of the (α-Cr + σ) type are formed. In this region of the diagram, X-ray powder photographs of 23 alloys were taken, and the results confirmed those shown in Fig. 2. From 800° to 600° C., the solubility of manganese in chromium continues to diminish, but the polyphase alloys, with very few exceptions, consisted of α-Cr, σ, and a third phase whose crystal structure was either identical with, or

---

* Both chromium and manganese were determined, and the analytical totals exceeded 99·90%, except for two batches of filings where the total was only 99·80%, owing to the accidental inclusion of particles of silica from the annealing tubes.

very closely similar to, that of α-manganese, and which is therefore referred to as the α′-Mn phase. Fig. 3 (Plate XLV) shows the structure of a normal two-phase (α-Cr + σ) alloy, and Fig. 4 (Plate XLV) that of a three-phase (α-Cr + σ + α′-Mn) alloy; the etching methods used to distinguish between the σ and α′-Mn alloys are described in the Appendix (p. 314). In the region of the diagram from 800° to 600° C., X-ray powder photographs were taken of 18 alloys, and the results confirmed the microscopical work, and permitted the identification of the α′-Mn phase. In this range of temperature, the relative proportion of the α′-Mn phase diminished on prolonged annealing, and there is little doubt that above 600° C., conditions of true equilibrium involve two-phase (α-Cr + σ) alloys. In the X-ray films, the α′-Mn phase gave rise to a set of characteristic lines which could readily be recognized, and which were identified as due to a structure of the α-manganese type. Thus in alloy 64 * the X-ray-diffraction film showed 3 lines due to the α-Cr phase, and 11 lines whose intensities and interplanar spacings agreed with the known values for α-manganese; the symbol α′-Mn is used to emphasize this resemblance.

Below 600° C., the solubility of manganese in chromium diminishes more rapidly, and in the region of the diagram between 523° and 600° C., powder X-ray photographs of 22 alloys were taken. In general, the microscopical and X-ray methods were in good agreement, although, in a few cases, alloys which had been described as two-phase on the basis of microscopical examination were found to contain traces of a third phase by the X-ray method. After annealing at 523° C. for 71 days, alloy 24·4 was homogeneous, whereas alloy 30·7 was of the two-phase (α-Cr + α′-Mn) type, and the solubility curve in Fig. 2 is drawn to agree with the X-ray data. In the composition range 30–65 at.-% manganese, alloys on annealing in the range 523°–600° C. were usually three-phase. The effect of increasing the annealing period was erratic, the proportion of the σ phase diminishing in some alloys, but remaining unchanged in others. When, however, alloys which had been annealed below 600° C., with the production of a three-phase (α-Cr + α′-Mn + σ) structure, were annealed above 600° C., the proportion of the α′-Mn phase diminished, and there seems no doubt that the equilibrium diagram contains a horizontal line at about 600° C. The proportion of the α′-Mn phase in the polyphase alloys increased with increasing manganese content up to 65 at.-%, but beyond this a change in the general structure of the alloys occurred, and they consisted of the σ phase, together with a fine structure in which both α-Cr and α′-Mn phases could be detected. This fine structure might be taken to indicate that the state of true equilibrium involves a eutectoidal decomposition of the σ phase into α-Cr and α′-Mn solid solutions, but this seems improbable for the following reasons:

(1) No combination of heat-treatment and deformation has yet caused a homogeneous σ alloy to decompose on annealing below 600° C.

(2) In this work, and in that of Carlile, Christian, and Hume-Rothery, alloys on the manganese-rich side of the σ phase showed (σ + α-Mn) structures after annealing at low temperatures. Most of these alloys were homogenized in the two-phase (σ + γ-Mn) region, and for those a low-temperature annealing treatment might perhaps transform the γ-manganese into α-manganese, and leave the σ phase unchanged owing to a sluggish transformation. The diagram of Carlile, Christian, and Hume-Rothery, however, shows one alloy homogenized in the σ region and one homogenized in the γ-Mn region, both of which consisted of (σ + α-Mn) on annealing at low temperatures, and it seems improbable that this should happen unless the σ phase persisted as a stable phase at low temperatures.

(3) In some alloys lying just outside the chromium-rich boundary of the σ phase, prolonged annealing below 600° C. gave structures which were essentially of the (σ + α′-Mn) type, the amounts of the α-Cr phase being relatively small, as though they were a remnant resulting from a failure to obtain complete equilibrium. This fact, together with the change in the nature of the alloys with more than 65 at.-% manganese, and the persistence of the homogeneous σ phase alloys at low temperatures, suggest that, under conditions of true equilibrium, the α′-Mn phase is formed by a peritectoid reaction at 600° C., and the dotted line in Fig. 2 indicates a possible composition for the α′-Mn phase. Annealing treatments have been extended up to 42 days at 600° C. and 71 days at 523° C., and under these conditions it has not been possible to obtain a homogeneous α′-Mn alloy, whilst no satisfactory correlation has been traced between the compositions of the alloys and the relative proportions of the different phases, so that the possibility that conditions of true equilibrium involve a eutectoidal decomposition of the σ phase cannot be excluded.

V—CONCLUSIONS

The diagram shown in Fig. 2 is an equilibrium diagram above 800° C. Below 800° C. it shows the limits of the chromium-base solid solution under conditions of annealing of the order of 1–2 months, and above 600° C. the solubility curve is not likely to be seriously in error. In the range 800°–600° C. when the solubility limit is exceeded three-phase alloys of the (α-Cr + σ + α′-Mn) type are produced on annealing previously homogenized specimens, and from these the α′-Mn phase gradually disappears on prolonged annealing. A horizontal almost certainly exists at 600° C., and below this temperature, three-phase alloys are again produced on annealing previously homogenized specimens, and it seems almost certain that, under conditions of true equilibrium, the

---

* Alloys are referred to by their atomic percentage of manganese.

α-Cr phase is in equilibrium with the α′-Mn phase, but the transformations are so sluggish that the composition of the latter cannot yet be determined. Thermodynamic considerations show that, as the alloys are so far from equilibrium after annealing for 42–71 days in the range 523°–600° C., it is probable that true equilibrium will be reached only after annealing treatments of some years, and the present paper is submitted to help the understanding of these difficult alloys.

ACKNOWLEDGEMENTS

The authors must express their thanks to Professor Sir Cyril Hinshelwood, F.R.S., for laboratory accommodation, and to Dr. F. M. Brewer for many other facilities which have encouraged their research.

APPENDIX

Etching Methods Used for Chromium–Manganese Alloys After Annealing at Low Temperatures

The two following reagents enable the different phases to be distinguished:

(i) *N*/10 *Hydrochloric Acid.*—In two-phase (Cr + σ) and (Cr + α′-Mn) alloys the σ or α′-Mn phase is attacked when present in a chromium matrix. In three-phase (Cr + σ + α′-Mn) alloys, the α′- Mn phase etches very readily, much more so than the σ phase. When present in a σ matrix, particles of the α′-Mn phase remain clear and white, and the σ matrix is attacked.

(ii) *Alcoholic Nitric Acid* (10–15%).—The α′-Mn phase is etched very distinctly, whether it occurs in a σ or a chromium matrix. The reagent does not, however, etch particles of the σ phase when present in a chromium matrix.

REFERENCES

1. S. J. Carlile, J. W. Christian, and W. Hume-Rothery, *J. Inst. Metals*, 1949–50, **76**, 169.
2. W. B. Pearson, J. W. Christian, and W. Hume-Rothery, *Nature*, 1951, **167**, 110.
3. U. Zwicker, *Z. Metallkunde*, 1951, **42**, 277.
4. U. Zwicker, *ibid.*, 1949, **40**, 337.
5. W. Hume-Rothery and J. W. Christian, *Monthly J. Inst. Metals* (*News Section*), **1950**, (May), 152.
6. W. B. Pearson and W. Hume-Rothery, *J. Inst. Metals*, 1952, **80**, 641.
7. S. J. Carlile, D.Phil. Thesis, Oxford: **1949**.
8. D. M. Poole and H. J. Axon, *J. Inst. Metals*, 1952, **80**, 599.

REPORT OF COUNCIL

FOR THE YEAR ENDED 31 DECEMBER 1952

The Council is pleased to be able to report that the year 1952 was a very active and successful one for the Institute. Meetings were attended by larger numbers of members than for many years and, thanks to the generosity of the contributors to the Industrial Donations Fund, it was possible not only to maintain the standard of the Institute's publications—which have a world-wide reputation—but also to overcome the arrears of printing of MSS. of papers accepted for publication.

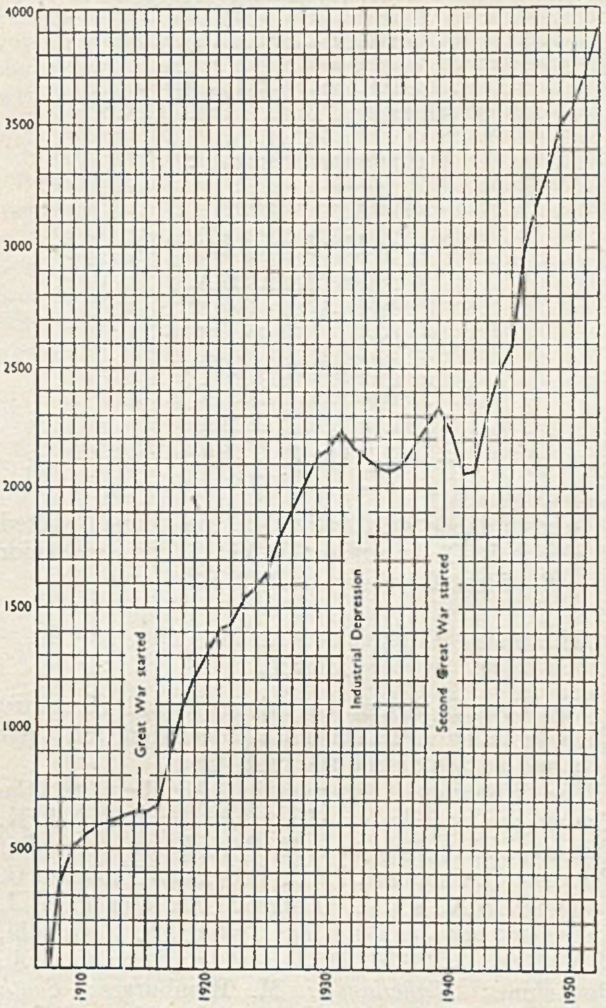
The Council's thanks are particularly due to the Chairman, Committee, and members of the Oxford Local Section for their invitation to hold the 1952 Autumn Meeting in Oxford. The arrangements made were excellent, the attendance was very good, and the generous hospitality offered to the members and their ladies was much appreciated.

Membership

The steady increase in the Institute's membership has been maintained, as will be seen from the table below. There is, however, still considerable scope for development both in the British Isles and in other countries throughout the world. The Council therefore again appeals to members to take all appropriate steps to increase the Institute's membership within their own industries and circle of acquaintances. Copies of a pamphlet on " The Institute of Metals: Particulars of its Objects, Work, and Membership " are available, on application, to all who can use them for membership development purposes.

It is particularly hoped that members resident outside the British Isles will be prepared to endeavour to increase the Institute's membership in their respective countries and branches of industry. To assist them, the Secretary will be glad not only to send them as many copies as are required of the pamphlet referred to above, but also to supply copies of lists of members in their respective countries for distribution with the pamphlets to possible members. Specimen copies of the (monthly) *Journal* and *Metallurgical Abstracts* can also be sent by the Secretary to persons likely to be interested.

At 31 December	1945	1946	1947	1948	1949	1950	1951	1952
Honorary Members . .	6	6	9	9	11	11	11	11
Fellows . .	6	6	7	6	9	10	8	8
Ordinary Members . .	2213	2414	2491	2546	2685	2815	2941	3144
Junior Members .	...	...	...	...	...	291	305	362
Associate Members . .	12	25	17	19	18	...	...	...
Student Members . .	361	529	655	746	783	452	462	423
Active List .	2598	2980	3179	3326	3506	3579	3727	3948
Suspense List .	179	58	36	55	67	97	124	95
Total .	2777	3038	3215	3381	3573	3676	3851	4043



Active Membership at 31 December 1908–1952.

Obituary

The Council deeply regrets to record the death of Sir William Griffiths, Past-President.

It also much regrets to record the deaths of the following members, which were notified during the year: Mr. H. B. Barnard, Mr. J. S. Bowden, Professor L. C. F. de Brouckère, Mr. C. H. Carder, Colonel W. C. Devereux (Member of Council 1939–43), Dr. H. D. H. Drane, Mr. I. J. Espir, Mr. G. H. Field, Mr. W. R. Franklin, Dr. F. Johnson, Sir James Lithgow, Dr. A. Morris, Mr. E. H. Munnik, Mr. E. J. Overton, Mr. F. Powell, Mr. E. W. Pritchard, Mr. M. Schrero, and Dr. G. Vanzetti.

OFFICERS OF THE INSTITUTE

The following members were declared elected to fill vacancies as honorary officers of the Institute with effect from the 1952 Annual General Meeting:

President:

C. J. SMITHELLS, M.C., D.Sc.

Vice-Presidents:

G. L. BAILEY, C.B.E., M.Sc.
S. F. DOREY, C.B.E., D.Sc., F.R.S.

Honorary Treasurer:

E. H. JONES

Ordinary Members of Council:

ALFRED BAER, B.A.
N. I. BOND-WILLIAMS, B.Sc.
N. P. INGLIS, Ph.D., M.Eng.
IVOR JENKINS, D.Sc.
A. G. RAMSAY, B.Sc., Ph.D.
H. SUTTON, D.Sc.
Major P. LITHERLAND TEED, A.R.S.M.
W. J. THOMAS

In accordance with Article 42, the Council elected Professor F. C. THOMPSON, D.Met., M.Sc., as Senior Vice-President for the year 1952–53.

HONORARY CORRESPONDING MEMBERS TO THE COUNCIL

The Council desires to express to all Honorary Corresponding Members its appreciation for their help and advice during the past year.

The Honorary Corresponding Members to the Council are as follows: *Australia:* Professor H. K. Worner, D.Sc.; *Belgium:* H. P. A. Féron; *Canada:* Professor B. Chalmers, Ph.D., D.Sc., and Professor G. Letendre, B.A., Ph.D.; *France:* Professor P. A. J. Chevenard and Jean Matter; *India:* N. P. Gandhi, M.A., B.Sc., A.R.S.M., D.I.C.; *Italy:* Leno Matteoli, Dott.chim.; *Netherlands:* M. Hamburger; *South Africa:* G. H. Stanley, D.Sc., A.R.S.M., and Professor L. Taverner, A.R.S.M., D.I.C.; *Spain:* Professor J. Orland, M.Sc., M.A., Ph.D., D.D.; *Sweden:* Professor Carl A. F. Benedicks, Fil.Dr., D.Ing.e.h., Dr.Techn.h.c., and Professor Axel Hultgren; *Switzerland:* Professor A. von Zeerleder, Dr.Ing.; *United States of America:* Professor R. F. Mehl, Ph.D., Hon.Eng.D., Hon.Sc.D., Professor C. S. Smith, Sc.D., and Dr. R. A. Wilkins.

INSTITUTE OF METALS MEDAL

The Institute of Metals (Platinum) Medal for 1952 was awarded to Mr. WILLIAM SYDNEY ROBINSON, until recently President of the Consolidated Zinc Corporation, Ltd., in recognition of his outstanding services to the non-ferrous metal industries in developing the Australian zinc–lead industry and the British zinc industry.

As Mr. Robinson was unable to come to England to receive his medal from the President, the Governor-General of Australia kindly presented the medal on behalf of the Council at a private ceremony.

W. H. A. ROBERTSON MEDAL

The W. H. A. Robertson Medal for 1951 was awarded to Mr. CYRIL ERNEST DAVIES, for his paper on "The Cold Rolling of Non-Ferrous Metals in Sheet and Strip Form", published in the *Journal*, 1950–51, vol. **78**, pp. 501–536.

ROSENHAIN MEDAL

The Rosenhain Medal for 1952 was awarded to Professor ANDRÉ GUINIER, of the Conservatoire National des Arts et Métiers, Paris, in recognition of his outstanding contributions in the field of physical metallurgy, particularly in connection with precipitation phenomena.

CAPPER PASS AWARDS

The Adjudicating Committee made no award for the year 1951 as, in its opinion, the quality and numbers of papers submitted for publication by the Institution of Mining and Metallurgy and the Institute of Metals, on the subjects for which the Capper Pass Awards are made, failed to reach a suitable standard.

It has been decided to include assaying within the fields for which the awards are made.

STUDENTS' ESSAY PRIZE

The first Students' Essay Prize of twenty guineas was awarded to Mr. J. C. WRIGHT (Student Member), Laboratory Assistant, Development and Research Department, The Mond Nickel Co., Ltd., Birmingham, for an essay on "The Metallographic Investigation of Failed High-Temperature Components". The essay was published in the *Bulletin*, 1952, vol. **1**, pp. 112–115.

This competition is open to all Student Members of the Institute and to all Associate Members of Local Sections who are eligible for Student Membership, provided that both are within the normal age-limits for Student Membership, viz. 17 to 25 years.

PRESIDENTIAL BADGE OF OFFICE

The Council has gratefully accepted a generous offer by the Directors of Johnson, Matthey and Co., Ltd., to present to the Institute a suitable badge of office for the President.

PUBLICATIONS

In view of the generous response to the appeal to industry for regular financial support of the Institute's work, the Council was able, during the year, to approve a plan submitted by the Publication Committee to overcome the arrears of printing of papers accepted by the Institute. As a result, 81 papers and lectures were published in the *Journal* in the calendar year 1952, as compared with 57 in 1951 and 61 in 1950 (the comparable figures for the financial year 1951–52 were 75 against 52 in the previous year, as mentioned in the Honorary Treasurer's Report), in addition to 13 papers published for a Symposium on " Properties of Metallic Surfaces ". By December 1952 the delay in publication of acceptable MSS. of papers had been reduced to 4 months, which is the minimum period necessary for the processes of refereeing, editing, printing, and despatch.

During the year certain economies were authorized in the production of the *Journal* and *Metallurgical Abstracts* and, in particular, the experiment has been tried of inserting plates at one place in each issue, which has effected an appreciable saving in money.

A full list of papers and lectures published during the calendar year 1952 is given in Appendix I to this Report (p. 320).

Metallurgical Abstracts has continued to maintain its high standard. Publication of abstracts of the literature is more prompt than for many years, and the number of periodicals searched has been increased.

Annual Indexes. The delays in publication of the annual indexes to *Metallurgical Abstracts*, caused by staff changes, have been overcome. The index to the volume which ended with the August 1952 issue is in the press, and should be ready for despatch to members in January 1953.

10-*Year Index.* The production of the two volumes (names and subjects) of the index to Volumes 1–10 (1934–1943) of *Metallurgical Abstracts*, which had also been delayed by staff changes, is now well advanced. The " Names " volume of the index (516 pp.) has already been printed and is being bound, and the " Subjects " volume is now going through the press. It is confidently hoped that this index will be ready for distribution by the middle of 1953. The MS. of the index to Volumes 11–20 (1944–1953) has been prepared up to Vol. 19 (1951–52), and it is intended to proceed with its publication after the completion of Volume 20, which is the current volume.

In the *Monograph and Report Series* there were published during the calendar year No. 11 : " Thermodynamics of Alloys ", by Mr. J. Lumsden ; No. 12 : " The Cold-Working of Non-Ferrous Metals and Alloys " ; No. 14 : " Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys ", and a second revised, edition of No. 3 : " Atomic Theory for Students of Metallurgy ", by Dr. W. Hume-Rothery, F.R.S. In addition, advance copies were published of the 13 papers to be included in No. 13 : " Properties of Metallic Surfaces ", the bound volume of which, containing a report of the discussion at a symposium on this subject, should be ready for publication early in 1953.

Discussions took place during the year regarding the possibility of co-operation between the British metallurgical societies in the publication of one comprehensive series of metallurgical abstracts, in lieu of the individual and independent abstracts published by three of the societies, and of reports of progress in metallurgy. Though the publication of one comprehensive series of abstracts was agreed to be attractive in principle, the Council was informed that there were special circumstances which necessitated the continuation of the abstracts of one of the Institutes in their present form. As the proposed scheme would be financially unsound if there were duplicate publication, the Council regretfully reached the conclusion that no further action on this matter is desirable at the present time.

It appeared that no agreement could be reached by the four British metallurgical societies on the joint publication of periodical reports of progress in metallurgy. The Council is, however, exploring the possibility of publishing such reports with the exclusion of subjects directly relating to extraction and ferrous metallurgy, and has appointed a sub-committee to prepare a detailed plan and estimates of cost.

GENERAL MEETINGS

On Thursday, 3 January 1952 a General Meeting of the Institute was held in Birmingham, when there was an Informal Discussion, arranged by the Metallurgical Engineering Committee, on " Tool and Die Materials for the Extrusion of Non-Ferrous Metals and Alloys ". There was a large attendance, and a valuable result of the meeting was the bringing closer together of the users and the die-steel manufacturers.

The Forty-Fourth Annual General Meeting was held in London from 24 to 27 March 1952, when Dr. C. J. SMITHELLS, M.C., was inducted into the Chair. The Forty-Second May Lecture was delivered on Monday, 24 March, at the Royal Institution, Albemarle Street, London, W.1, by Dr. J. J. P. STAUDINGER on " The Place of Plastics in the Order of Matter ". One day of the meeting was devoted to a Symposium on " Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys ", arranged by the Metallurgical Engineering Committee. An innovation was the holding of simultaneous scientific and technical sessions on widely differing subjects. The experiment was most successful, a large attendance being maintained at each session throughout the meeting. The procedure was repeated at the Autumn Meeting, when its success was confirmed. It may be assumed, therefore, that—whenever appropriate—simultaneous sessions on widely differing subjects will be arranged in connection with future General Meetings ; they have enabled the maximum number of papers to be discussed, within the time available, by members of differing, and often specialist, interests.

The Forty-Fourth Annual Autumn Meeting was held in Oxford from 15 to 19 September 1952, by invitation of the Oxford Local Section. The meeting was most successful, and about 400 members, delegates, and ladies took part. The Twenty-Third Autumn Lecture was delivered in the Sheldonian Theatre on Monday, 15 September, by Professor H. W. SWIFT, M.A., D.Sc., whose subject was "On the Foot-Hills of the Plastic Range". During the meeting there was an Informal Discussion on "Grain Boundaries", arranged, as an experiment, by the Metal Physics Committee.

On Wednesday, 19 November 1952, an all-day Symposium, arranged by the Metal Physics Committee, was held at the Royal Institution, Albemarle Street, London, W.1., on "Properties of Metallic Surfaces". Over 450 members and visitors were present, many of them from overseas. The large attendance and full discussion would appear to confirm the need for a meeting on this subject. The papers, and a report of the discussion, will be published at an early date as Monograph No. 13.

Students' Educational Tour

An Easter Vacation Educational Tour for Junior and Student members was held in South Lancashire from 31 March to 4 April 1952, inclusive. It was attended by 31 Junior and Student Members, and 7 works were visited. The Council records its gratitude to the Directors of the works visited for their co-operation in making the tour a success.

As these tours appear to meet a real need, especially in the case of young men in industry and students at technical schools, another—to be held in the Birmingham area—is being arranged for the 1953 Easter Vacation.

Local Sections and Associated Societies

The six Local Sections (Birmingham, London, Oxford, Scottish, Sheffield, and South Wales) had good programmes of meetings during the winter session. The newly formed Oxford Local Section is to be particularly congratulated on its growth and keenness. Though it had only been in existence for one session, it acted as host to the Institute at one of the most successful of provincial Autumn Meetings of the Institute.

Members continued to enjoy the privilege of free membership of three Associated Societies (Leeds Metallurgical Society, Liverpool Metallurgical Society, and Manchester Metallurgical Society), and during the year arrangements were made for similar facilities to be granted to members by the North East Metallurgical Society, which became an Associated Society.

On 2 October 1952 a joint meeting of members of the Institute with local members of the Chemical Society, Royal Institute of Chemistry, and Society of Chemical Industry took place in Bristol, when Major P. L. Teed, A.R.S.M., read a paper on "Some Metallurgical Problems Imposed by Stratospheric Flight".

Special Committees

The Metal Physics Committee met three times during the year and arranged a scientific session at the Annual General Meeting, an Informal Discussion on "Grain Boundaries" at the Autumn Meeting, and an all-day Symposium on "Properties of Metallic Surfaces" on 19 November. It also made arrangements for the presentation of papers on metal physics to Local Sections and Associated Societies.

The Metallurgical Engineering Committee met four times. It arranged an Informal Discussion on "Tool and Die Materials for Non-Ferrous Metals and Alloys" in Birmingham on 3 January and an all-day Symposium on "Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys" in connection with the Annual General Meeting. The Council has also approved a proposal by the Committee that, in connection with the 1953, 1954, and 1955 Annual General Meetings, there shall be a series of three all-day Symposia on "The Control of Quality in the Production of Wrought Non-Ferrous Metals" covering, respectively, the subjects of melting and casting (1953), working (1954), and heat-treatment and finishing (1955). The Committee has arranged an Informal Discussion on "Rolls and their Maintenance in the Non-Ferrous Metals Industry", to be held in Birmingham on 8 January 1953.

An *ad hoc* Committee is studying the question of the publication of reviews of progress in non-ferrous metallurgy.

A list of members of the main Committees of the Institute is given in Appendix III.

Industrial Donations Fund

The Council desires to record its warm appreciation of the continuing generous response to its appeal for regular financial support of the Institute's work by industry. A list of the contributions to the fund during the financial year ended 30 June 1952 is attached as Appendix II to this Report (p. 322).

Without this financial support, the Institute's work would have had to be severely restricted; with its aid, however, the Council has been able to expedite the publication of acceptable scientific and technical papers and to consider the provision of new services for science and industry.

Staff

There were no changes among the senior members of the staff during the year.

JOINT ACTIVITIES

Joint Library and Information Department

Members, Government Departments, Research Laboratories, Universities, and other teaching establishments have continued to make great use of the facilities offered by the Library and Information Department. During 1952 13,448 publications were borrowed; in 1951 the figure was 11,617. The number of text-books acquired was 475, and the Council offers its thanks to the donors who presented copies of their works to the Library.

Use of the Lending Library is a valuable privilege of membership. Books and periodicals can be borrowed, both from the Joint Library and also, on application through the Librarian, from the Science Library and the National Central Library. Under certain conditions photocopies of documents can be obtained for members, both at home and abroad; in 1952 130 photostat copies and 27 microfilms were supplied to members.

The Information Department, an important part of the service, is prepared to answer scientific and technical enquiries from members, but it is not its function to give the type of advice which comes within the field of the metallurgical consultant.

JOINT COMMITTEE FOR NATIONAL CERTIFICATES IN METALLURGY

The Joint Committee has approved five new schemes for National Certificates in Metallurgy during 1952; other schemes have been revised, and schemes are also being organized in districts not yet covered, with a view to suitable courses being started as soon as possible.

During 1952 the number of courses in operation has been:

Ordinary Certificate Courses	30
Contributory Centres with First- or First- and Second-year Courses	6
Higher Certificate Courses	18

Final examinations for the Ordinary National Certificate have been held at 28 Technical Colleges and for the Higher Certificate at 16 Technical Colleges.

The lists of those who have qualified for Certificates and for receipt of prizes are published in the Joint Committee's annual report. This report also contains reports on the final examinations held during the year.

JOINT COMMITTEE ON METALLURGICAL EDUCATION

The Joint Committee has published a report on "The Education and Training of Metallurgists". The following is a summary of the recommendations:

(1) The universities are at present, are likely to be, and should continue to be, the chief source of high-grade technologists. Universities should expand their output of technologists to the greatest extent they can achieve.

(2) Employers should finance the attendance of suitable employees at post-graduate courses in universities.

(3) Formally established methods should be adopted for training graduates immediately they enter industry.

(4) Since more technologists are needed than the universities can supply, the remainder coming from colleges of technology, a few colleges of technology should be upgraded and equipped for the education of high-grade technologists.

(5) In high-grade, technical teaching institutions, instruction in technical and elementary subjects should be physically separated from that in higher technological subjects.

(6) The present standard of staffing in technical colleges and colleges of technology should be raised.

(7) There should be active collaboration in various fields between colleges of technology and employers.

(8) The whole of the present examinations in metallurgy for non-university students should be considered with a view to their co-ordination and rationalization and the establishment of the Associateship of the Institution of Metallurgists as the normal goal for almost all such students.

(9) Further provision for sandwich courses should be made.

(10) Since it appears that operatives in the metallurgical industries are insufficiently responsive to the opportunities offered for education and training, the greater use of incentives should be examined; employers should take into account any examination successes by their employees when considering promotion and the filling of vacancies in operational departments.

(11) To achieve a balance in the education of metallurgists, all teaching establishments—including the universities—and employers are recommended to instruct students and employees in the new humanities.

The Committee welcomes comments on the recommendations made in the report, and these will be given careful attention. Copies of the report can be obtained on application to the Secretary to the Joint Committee for Metallurgical Education at 4 Grosvenor Gardens, London, S.W.1.

MOND NICKEL FELLOWSHIPS COMMITTEE

During 1952 Mr. E. H. JONES replaced Mr. W. A. C. NEWMAN as representative of the Institute of Metals on the Committee.

The following awards were made for 1952:

A. G. DUCE (Joseph Lucas (Gas Turbine Equipment), Ltd.) to study in the United Kingdom, the Continent, the U.S.A., and Canada, the metallurgy and testing of materials, especially in sheet form, developed for high-temperature service, and the techniques employed in the manufacture of the combustion systems of gas-turbine engines.

F. G. HORTON (National Foundry College, Wolverhampton) to study casting production methods in the United Kingdom and on the Continent, with special reference to shell moulding, centrifugal casting, and sand-cement moulding.

G. P. KEMPSON (Henry Wiggin and Co., Ltd.) to study the development, application, and control of ferrous and non-ferrous melting and ingot-casting processes in the United Kingdom, on the Continent, the U.S.A., and Canada.

H. A. LONGDEN (Steel, Peech, and Tozer) to study metallurgical control methods in the United Kingdom, on the Continent, the U.S.A., and Canada, with special reference to open-hearth slag and temperature control.

APPENDIX I

List of Papers Published

The following is a complete list of the papers and lectures published by the Institute during the calendar year 1952:

1342. The Production and Properties of Oxide-Reduced Copper Powder. By E. C. Ellwood, Ph.D., A.I.M., and W. A. Weddle, B.Sc.

1343. Unrelated Simultaneous Interdiffusion and Sintering in Copper–Nickel Compacts. By J. M. Butler, M.A., Ph.D., and T. P. Hoar, M.A., Ph.D., F.I.M.

1344. Heat-Treatment of Titanium-Rich Titanium–Iron Alloys. By H. W. Worner, M.Sc.

1345. The Solid Solutions of Zinc in Aluminium. By E. C. Ellwood, Ph.D., A.I.M.

1346. Metal Economics. I.—Primary Resources of Ferrous and Non-Ferrous Metals:
    Introduction. By Professor A. J. Murphy, M.Sc., F.I.M.
    The World Supply of Non-Ferrous Metals, Including the Light Metals. By R. Lewis Stubbs.
    Metals as Natural Resources. By Professor S. Zuckerman, C.B., F.R.S.
    World Demand and Resources of Iron Ore. By T. P. Colclough, C.B.E., D.Sc.

Metal Economics. II.—Scrap Reclamation, Secondary Metals, and Substitute Metals:
    The Scope for Conservation of Metals, Ferrous and Non-Ferrous. By C. A. Bristow, B.Sc., A.R.S.M., F.I.M., A. J. Sidery, Assoc.Met., and H. Sutton, D.Sc., F.I.M.
    Economy by Standardization of Alloys and of the Method of Reclamation of Scrap Metals. By C. Dinsdale, M.Sc., F.I.M.
    The Influence of Specifications on Productivity and the Economic Utilization of Ferrous and Non-Ferrous Metals. By F. Hudson, F.I.M.
    Secondary Heavy Metals. By E. H. Jones, F.I.M.
    Secondary Aluminium and Magnesium. By Colonel W. C. Devereux, C.B.E.

1347. Electric Furnaces for the Thermal Treatment of Non-Ferrous Metals and Alloys. By C. J. Evans, A.M.I.E.E., P. F. Hancock, B.A., F.I.M., F. W. Haywood, Ph.D., B.Sc., F.R.I.C., F.I.M., and J. McMullen, A.I.M.

1348. Gas Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys. By J. F. Waight, B.Sc.(Eng.), M.Inst.Gas E.

1349. Batch and Continuous Annealing of Copper and Copper Alloys. By Edwin Davis, M.Sc., F.I.M., and S. G. Temple, M.Sc., A.I.M.

1350. Bright Annealing of Nickel and Its Alloys. By H. J. Hartley, M.Sc., F.I.M., and E. J. Bradbury, M.Eng., A.M.I.Mech.E., A.I.M.

1351. Batch Thermal Treatment of Light Alloys. By C. P. Paton, B.Eng.

1352. Flash Annealing of Light Alloys. By R. T. Staples.

1353. Continuous Heat-Treatment of Aluminium Alloys of the Duralumin Type. By Marcel Lamourdedieu.

1354. The Sintering of Copper–Zinc Powder Compacts. By D. D. Howat, B.Sc., Ph.D., F.R.I.C., R. L. Craik, B.Sc., A.R.T.C., and J. P. Cranston, B.Sc., Ph.D., A.R.T.C.

1355. The Effect of the Elements of the First Long Period on the $\alpha \rightleftharpoons \beta$ Transformation in Titanium. By A. D. McQuillan, Ph.D., B.Sc.

1356. Determination of Elastic Constants and Stress/Strain Relationship to Fracture of Sintered Tungsten Carbide–Cobalt Alloys. By E. Lardner, B.Sc., A.I.M., and N. B. McGregor.

1357. The Constitution of the Copper-Rich Copper–Zinc–Germanium Alloys. By P. Greenfield, Ph.D., B.Sc., and Professor G. V. Raynor, M.A., D.Sc.

1358. Micrographic Aspects of the Diffusion of Zinc and Aluminium in Copper. By H. Bückle, Dr.-Ing., and J. Blin.

1359. Note on Sub-Crystal Structure in Cold-Rolled Aluminium. By A. E. L. Tate, A.I.M., and D. McLean, B.Sc.

1360. Fundamental Reactions in the Vacuum-Fusion Method and Its Application to the Determination of O_2, N_2, and H_2 in Mo, Th, U, V, and Zr. By H. A. Sloman, M.A., F.R.I.C., F.I.M., and C. A. Harvey. With an Appendix by O. Kubaschewski, Dr. phil. nat. habil.

1361. The Opaque-Stop Microscope as a Means of Studying Surface Relief. By W. M. Lomer, M.Sc., B.A., and P. L. Pratt, B.Sc.

1362. A Modified Dew-Point Method for Vapour-Pressure Measurements of Lead–Mercury Alloys. By B. R. Burgan, R. C. Hall, and R. F. Hehemann.

1363. The Effect of Grain-Size on the Structural Changes Produced in Aluminium by Slow Deformation. By W. A. Rachinger, M.Sc.

1364. Observations on the Structure and Properties of Wrought Copper–Aluminium–Nickel–Iron Alloys. By Maurice Cook, D.Sc., Ph.D., F.I.M., W. P. Fentiman, B.Sc., A.I.M., and Edwin Davis, M.Sc., F.I.M.

1365. The Solid Solubilities of Cadmium, Indium, and Tin in Aluminium. By H. K. Hardy, Ph.D., M.Sc., A.R.S.M., A.I.M.

1366. The Constitution of the Aluminium–Chromium–Zinc Alloys at Low Chromium Contents. By A. R. Harding, B.Sc., Ph.D., and Professor G. V. Raynor, M.A., D.Sc.

1367. The Aluminium-Rich Alloys of the System Aluminium–Chromium–Iron. By J. N. Pratt, B.Sc., Ph.D., and Professor G. V. Raynor, M.A., D.Sc.

1368. The Equilibrium Diagram of the System Copper–Gallium in the Region 30–100 At.-% Gallium. By J. O. Betterton, Jr., D.Phil., B.S., and William Hume-Rothery, O.B.E., F.R.S.

1369. Presidential Address. By C. J. Smithells, M.C., D.Sc., F.I.M.

1370. A Study of Some Factors Influencing the Young's Modulus of Solid Solutions. By A. D. N. Smith, B.A.

1371. The Ageing Characteristics of Ternary Aluminium–Copper Alloys with Cadmium, Indium, or Tin. By H. K. Hardy, Ph.D., M.Sc., A.R.S.M., A.I.M.

1372. Residual Stresses in Chill-Cast and Continuously Cast Aluminium Alloy Billets. By R. A. Dodd, M.Sc., Ph.D., A.I.M., A.R.I.C.

1373. Stress-Recovery in Aluminium. By W. A. Wood, D.Sc., and J. W. Suiter, B.Sc.

1374. Creep Processes in Coarse-Grained Aluminium. By D. McLean, B.Sc.

1375. Electron-Microscopic Studies of Slip in Aluminium During Creep. By J. Trotter.

1376. Allotropic Transformation in Titanium–Zirconium Alloys. By Professor Pol Duwez, D.Sc.

1377. The Place of Plastics in the Order of Matter. Forty-Second May Lecture. By J. J. P. Staudinger, Dr.-Ing.

1378. Boundary Slip in Bicrystals of Tin. By K. E. Puttick, B.Sc., and Ronald King, B.Sc.

1379. X-Ray Diffraction Studies in Relation to Creep. By G. B. Greenough, Ph.D., (Mrs.) Catherine M. Bateman, B.Sc., and (Mrs.) Edna M. Smith, B.A.

1380. The Formation of the Ni_3Al Phase in Nickel–Aluminium Alloys. By R. W. Floyd, B.Sc., A.I.M.

1381. The Effect of Metal/Mould Reaction on 85 : 5 : 5 : 5 Leaded Gun-Metal Sand Castings. By N. B. Rutherford, B.Sc., A.I.M.

1382. Segregation of Iron and Phosphorus at the Grain Boundaries in 70 : 30 Brass During Grain Growth. By H. M. Miekk-oja, Sc.D.
1383. The Constitution of Nickel-Rich Alloys of the Nickel–Chromium–Titanium System. By A. Taylor, Ph.D., F.I.M., F.Inst.P., and R. W. Floyd, B.Sc., A.I.M.
1384. The Constitutional Diagram of the Chromium–Tungsten System. By H. T. Greenaway, B.Met.E.
1385. The Effect of Zirconium on the Properties and Structure of Superduralumin, with Particular Reference to Forgings. By M. Tournaire and M. Renouard.
1386. Lattice-Spacing Relationships in Aluminium-Rich Solid Solutions of the Aluminium–Magnesium and Aluminium–Magnesium–Copper Systems. By D. M. Poole, B.Sc., and H. J. Axon, D.Phil.
1387. Factors Affecting Equilibrium in Certain Aluminium Alloys. By E. C. Ellwood, Ph.D., F.I.M.
1388. The Factors Affecting the Formation of 21/13 Electron Compounds in Alloys of Copper and of Silver. By W. Hume-Rothery, O.B.E., F.R.S., J. O. Betterton, Jr., D.Phil., B.S., and J. Reynolds, B.A.
1389. The Lattice Spacings and Densities of Gold–Nickel Alloys at 25° C. By E. C. Ellwood, Ph.D., F.I.M., and K. Q. Bagley, Ph.D.
1390. The Metallography of Uranium. By B. W. Mott, M.A., and H. R. Haines.
1391. The Application of Polarized Light to the Examination of Various Anisotropic Metals and Intermetallic Phases. By B. W. Mott, M.A., and H. R. Haines.
1392. The Equilibrium Diagram of the System Copper–Indium in the Region 25–35 At.-% Indium. By J. Reynolds, B.A., W. A. Wiseman, and W. Hume-Rothery, O.B.E., F.R.S.
1393. The Constitution and Structure of Nickel–Vanadium Alloys in the Region 0–60 At.-% Vanadium. By W. B. Pearson, D.F.C., M.A., and W. Hume-Rothery, O.B.E., F.R.S.
1394. Twin Accommodation in Zinc. By P. L. Pratt, B.Sc., Ph.D., and S. F. Pugh, M.A., A.I.M.
1395. The Cubic–Tetragonal Transformation in Manganese–Copper Alloys. By Z. S. Basinski, B.A., and J. W. Christian, M.A., D.Phil.
1396. Specialized Microscopical Techniques in Metallurgy. By Professor S. Tolansky, D.Sc., F.R.S.
1397. Radioisotopes in the Study of Metal Surface Reactions in Solutions. By Massoud T. Simnad, Ph.D.
1398. The Influence of Machining and Grinding Methods on the Mechanical and Physical Condition of Metal Surfaces. By Peter Spear, B.Eng., Ian R. Robinson, B.Sc., and K. J. B. Wolfe, M.Sc.
1399. The Effect of Lubrication and Nature of Superficial Layer After Prolonged Periods of Running. By F. T. Barwell, Ph.D., D.I.C., B.Sc.(Eng.), Wh.Sch., M.I.Mech.E., A.M.I.E.E.
1400. The Crystalline Character of Abraded Surfaces. By P. Gay, M.A., Ph.D., and P. B. Hirsch, M.A., Ph.D.
1401. The Effect of Surface Conditions on the Mechanical Properties of Metals, Mainly Single Crystals. By Professor E. N. da C. Andrade, F.R.S.
1402. The Effect of Surface Condition on the Strength of Brittle Materials. By Professor C. Gurney, M.A., D.Sc., F.Inst.P.
1403. The Influence of Surface Condition on the Fatigue Strength of Steel. By R. J. Love, Wh.Sch., A.M.I.Mech.E.
1404. The Influence of Surface Films on the Friction and Deformation of Surfaces. By F. P. Bowden, Sc.D., F.R.S., and D.Tabor, Ph.D.
1405. Diffusion Coatings. By D. M. Dovey, M.A., A.R.I.C., A.I.M., I. Jenkins, D.Sc., F.I.M., and K. C. Randall, B.Sc.
1406. The Nature and Properties of the Anodic Film on Aluminium and Its Alloys. By H. W. L. Phillips, M.A., F.R.I.C., F.Inst.P., F.I.M.
1407. Chemical Behaviour as Influenced by Surface Condition. By U. R. Evans, Sc.D., M.A., F.R.S.
1408. The Effect of Method of Preparation on the High-Frequency Surface Resistance of Metals. By R. G. Chambers, M.A., Ph.D., and A. B. Pippard, M.A., Ph.D.
1409. Some Friction Effects in Wire Drawing. By G. D. S. MacLellan, M.A., Ph.D.
1410. The Viscosity of Molten Tin, Lead, Zinc, Aluminium, and Some of Their Alloys. By T. P. Yao, B.Sc., Ph.D., and V. Kondic, B.Sc., Ph.D.
1411. The Constitution of Nickel-Rich Alloys of the Nickel–Titanium–Aluminium System. By A. Taylor, Ph.D., F.Inst.P., and R. W. Floyd, B.Sc., A.I.M.
1412. Relative Grain Translations in the Plastic Flow of Aluminium. By W. A. Rachinger, M.Sc.
1413. The Influence of Primary Particles on the Grain-Size of Cast Magnesium–Aluminium Alloys. By W. A. Baker, B.Sc., F.I.M., Myriam D. Eborall, B.A., and A. Cibula, M.A., A.I.M.
1414. The Young's Modulus, Poisson's Ratio, and Rigidity Modulus of Some Aluminium Alloys. By N. Dudzinski, Dipl.Ing.(Chem.).
1415. Distribution Equilibria in Some Ternary Systems Me_1–Me_2–B and the Relative Strength of the Transition-Metal–Boron Bond. By Professor G. Hägg, Ph.D., and R. Kiessling, Ph.D.
1416. The Recovery of Polycrystalline Aluminium. By J. A. Ramsey, M.Sc.
1417. The Solid Solubility of Silver in Aluminium. By L. Rotherham, M.Sc., F.Inst.P., F.I.M., and L. W. Larke, A.F.R.Ae.S., L.I.M.
1418. The Constitution of Tantalum–Titanium Alloys. By D. Summers-Smith, B.Sc., A.R.T.C.
1419. Residual Stresses in Aluminium Alloy Sand Castings. By R. A. Dodd, M.Sc., Ph.D., A.I.M., A.R.I.C.
1420. The Creep/Time Relationship Under Constant Tensile Stress. By S. Bhattacharya, B.Sc., Ph.D., W. K. A. Congreve, B.A.Sc., Ph.D., and Professor F. C. Thompson, D.Met.
1421. The Temperature Dependence of Transient and Secondary Creep of an Aluminium Alloy to British Standard 2L42 at Temperatures Between 20° and 250° C. and at Constant Stress. By A. E. Johnson, D.Sc., M.Sc.Tech., M.I.Mech.E., and N. E. Frost, B.Sc., A.M.I.Mech.E.
1422. On the Foot-Hills of the Plastic Range. Twenty-third Autumn Lecture. By Professor H. W. Swift, M.A., D.Sc., M.I.Mech.E.
1423. The Embrittlement of Copper–Antimony Alloys at Low Temperatures. By D. McLean, B.Sc.
1424. The Gold–Platinum System. By A. S. Darling, B.Sc.Eng., A.M.I.Mech.E., R. A. Mintern, and J. C. Chaston, Ph.D., A.R.S.M., F.I.M.
1425. Crystal Slip in Aluminium During Creep. By D. McLean, B.Sc.
1426. The Viscosity of Aluminium and Binary Aluminium Alloys. By Professor W. R. D. Jones, D.Sc., and W. L. Bartlett, B.Sc., Ph.D.
1427. Plastic Deformation of Coarse-Grained Aluminium. By (Mrs.) V. M. Urie, B.Sc., and H. L. Wain, B.Met.E.
1428. Direct Examination of Solid Surfaces Using a Commercial Electron Microscope in Reflection. By J. W. Menter, M.A., Ph.D., A.Inst.P.
1429. A Study of Order–Disorder and Precipitation Phenomena in Nickel–Chromium Alloys. By A. Taylor, Ph.D., F.Inst.P., and K. G. Hinton, B.Sc.
1430. Deformation of Magnesium at Various Rates and Temperatures. By J. W. Suiter, B.Sc., and W. A. Wood, D.Sc.
1431. The Effect of Certain Solute Elements on the Recrystallization of Copper. By V. A. Phillips, A.R.S.M., B.Sc., D.Eng., A.I.M., and Professor Arthur Phillips, D.Eng.
1432. Equilibrium Relations at 460° C. in Aluminium-Rich Alloys Containing 0–7% Copper, 0–7% Magnesium, and 1·2% Silicon. By H. J. Axon, B.Met., D.Phil.
1433. An Example of Strain-Relief In Powder Specimens. By J. Gordon Parr, B.Sc.
1434. The Sub-Grain Structure in Aluminium Deformed at Elevated Temperatures. By J. A. Ramsey, M.Sc.
1435. The High-Temperature Oxidation of Some Cobalt-Base and Nickel-Base Alloys. By Professor A. Preece, M.Sc., F.I.M., and G. Lucas, Ph.D.

APPENDIX II

Contributors to the Industrial Donations Fund in the Financial Year ended 30 June 1952

Donor	Gross, after Recovery of Tax by the Institute		
	£	s.	d.
United States Copper and Brass Industry: individual subscriptions totalling $1950, by the following nine companies:			
American Brass Co., The	692	9	9
Bridgeport Brass Co.			
Bristol Brass Corp., The			
Chase Brass and Copper Co.			
Chicago Extruded Metals Co.			
New Haven Copper Co., The			
Revere Copper and Brass, Inc.			
Scovill Manufacturing Co.			
Wolverine Tube Division, Calumet and Hecla Consolidated Copper Co., Inc.			
*Enfield Rolling Mills, Ltd. (incl. Enfield Copper Refining Co., Ltd.; Enfield Rolling Mills (Aluminium), Ltd.; Holloway Metal Roofs, Ltd.; and London Zinc Mills, Ltd.)	523	16	1
*Mond Nickel Co., Ltd., The (incl. Birlec, Ltd.; Henry Wiggin and Co., Ltd., and associated companies in the United States and Canada)	523	16	1
*Consolidated Zinc Corporation, Ltd., The (incl. The Broken Hill Corporation, Ltd.; Imperial Smelting Corporation, Ltd.; The National Smelting Co., Ltd.; New Broken Hill Consolidated, Ltd.; Northern Smelting and Chemical Co., Ltd.; Sulphide Corporation Ltd.; and The Zinc Corporation, Ltd.)	476	3	9
*Goodlass Wall and Lead Industries, Ltd.: Associated Lead Industries, Ltd. £190 9 6 Fry's Metal Foundries, Ltd. (incl. Antifriction Bearing Co., Ltd., The; Atlas Metal and Alloys, Ltd.; and The Eyre Smelting Co., Ltd.) £95 4 9	285	14	3
*Mufulira Copper Mines, Ltd.	285	14	3
*Roan Antelope Copper Mines, Ltd.	285	14	3
*Imperial Chemical Industries, Ltd., and its subsidiary companies	†250	0	0
Aluminium Laboratories, Ltd. (incl. Aluminium Union, Ltd.; Northern Aluminium Co., Ltd.; and Stand, Ltd.)	200	0	0
General Motors, Ltd. (incl. A.C. Sphinx Spark Plug Co., Ltd.; Delco-Remy-Hyatt, Ltd.; and Frigidaire, Ltd.)	200	0	0
Tube Investments, Ltd. (incl. The Chesterfield Tube Co., Ltd.; Mersey Cable Works, Ltd.; Reynolds Light Alloys, Ltd.; Reynolds Rolling Mills, Ltd.; Simplex Electric Co., Ltd.; South Wales Aluminium Co., Ltd.; and T.I. Aluminium, Ltd.)	200	0	0
*British Aluminium Co., Ltd., The (incl. Aluminium Corporation, Ltd.; William Mills, Ltd.; and North British Aluminium Co., Ltd.)	190	9	6
*McKechnie Brothers, Ltd.	190	9	6
*Magnesium Elektron, Ltd. (incl. F. A. Hughes and Co., Ltd.)	190	9	6
*Metallo-Chemical Refining Co., Ltd.	190	9	6
Nchanga Consolidated Copper Mines, Ltd.	150	0	0
Rhodesia Broken Hill Development Co., Ltd., The	150	0	0
Rhokana Corporation, Ltd.	150	0	0
Allen (W. H.) and Sons, Ltd.	105	0	0
Vickers-Armstrongs, Ltd., and Vickers, Ltd. (incl. A.B.C. Motors, Ltd.; Robert Boby, Ltd.; Cooke, Troughton and Simms, Ltd.; Ioco, Ltd.; George Mann and Co., Ltd.; Palmers Hebburn Co., Ltd.; G. J. Worssam and Son, Ltd.; and Powers-Samas Accounting Machines, Ltd.)	105	0	0
Booth (James) and Co., Ltd. (incl. John Wilkes, Sons and Mapplebeck, Ltd.)	100	0	0
Capper Pass and Son, Ltd. (incl. George Pizey and Co., Ltd.; The Tyne Solder Co., Ltd.; and Victor G. Stevens, Ltd.)	100	0	0
Colvilles, Ltd.	100	0	0
Davy and United Engineering Co., Ltd.	100	0	0
Johnson, Matthey and Co., Ltd.	100	0	0
*Manganese Bronze and Brass Co., Ltd., The	†100	0	0
Rylands Brothers, Ltd., and The Whitecross Co., Ltd.	100	0	0
*Consolidated Tin Smelters, Ltd. (incl. The Cornish Tin Smelting Co., Ltd.; Eastern Smelting Co., Ltd.; The Penpoll Tin Smelting Co., Ltd.; and Williams, Harvey and Co., Ltd.)	95	4	9
*Enthoven (H. J.) and Sons, Ltd.	95	4	9
*High Duty Alloys, Ltd.	95	4	9
*Stone (J.) and Co. (Charlton), Ltd.	†95	4	9
*Venesta, Ltd.	95	4	9
*Whiley (Geo. M.), Ltd.	95	4	9
Rubery, Owen and Co., Ltd.	75	6	6
Mallory Metallurgical Products, Ltd.	52	10	0
Pyrotenax, Ltd.	52	10	0
*Murex, Ltd. (incl. Murex Welding Processes, Ltd.)	52	7	8
*Associated Electrical Industries, Ltd., on behalf of the A.E.I. Group of Companies	50	0	0
Birmingham Battery and Metal Co., Ltd., The	50	0	0
Birmingham Small Arms Company, Ltd., on behalf of the B.S.A. Group of Companies	50	0	0
*Brown (David) and Sons (Huddersfield), Ltd.	†50	0	0
*Dale (John), Ltd.	50	0	0
Danks (Edwin) and Co. (Oldbury), Ltd.	50	0	0
Fairey Aviation Co., Ltd., The	50	0	0
G.K.N. Group	50	0	0
Lucas (Joseph), Ltd. (inc. Rotax, Ltd.)	50	0	0
*Morgan Crucible Co., Ltd.	50	0	0
Rolls-Royce, Ltd.	50	0	0
*Simon-Carves, Ltd.	50	0	0
*Tennant (C.), Sons and Co., Ltd.	†50	0	0
Weir (G. and J.), Ltd. (incl. Argus Foundry, Ltd.)	50	0	0
*Hopkinsons, Ltd.	49	11	5
*A.P.V. Co., Ltd.	47	12	5
*Birmetals, Ltd.	47	12	5
*Birmingham Aluminium Casting (1903) Co., Ltd.	47	12	5
*Bristol Aeroplane Co., Ltd., The	47	12	5
*British Metal Corporation, Ltd., The	47	12	5
*Chloride Electrical Storage Co., Ltd., The	47	12	5
*Essex Aero, Ltd.	47	12	5
*London and Scandinavian Metallurgical Co., Ltd.	47	12	5
*Rotol, Ltd.	47	12	5
*Star Aluminium Co., Ltd. (incl. Anglo-Swiss Aluminium Co., Ltd.)	47	12	5
*Sterling Metals, Ltd.	47	12	5
*Wolverhampton Metal Co., Ltd., The (incl. James Bridge Copper Works, Ltd.)	47	12	5
Wednesbury Tube Co., Ltd., The	42	0	0
*Hughes-Johnson Stampings, Ltd., The	40	0	0
*Light Metal Forgings, Ltd.	40	0	0
*Ferranti, Ltd.	38	1	11
*Gibbons Brothers, Ltd. (incl. The Thermic Equipment and Engineering Co., Ltd.)	38	1	11
*Parkinson Stove Co., Ltd., The	†38	1	11
Aluminum Company of America ($100)	35	14	10
*British Tin Investment Corporation, Ltd.	30	0	0
*Derby and Co., Ltd.	30	0	0
*London Electric Wire Company and Smiths, Ltd., The (incl. Liverpool Electric Cable Co., Ltd.)	†30	0	0

Donor	Gross, after Recovery of Tax by the Institute		
	£	*s.*	*d.*
*Barker and Allen, Ltd.	28	11	5
*Bolton (Thomas) and Sons, Ltd.	28	11	5
*British Lead Mills, Ltd.	28	11	5
English Electric Co., Ltd., The (incl. D. Napier and Son, Ltd.)	26	5	0
Esso Petroleum Co., Ltd.	26	5	0
Head, Wrightson and Co., Ltd.	26	5	0
*Holroyd (John) and Co., Ltd.	26	3	9
Arkinstall Brothers, Ltd.	25	0	0
Braby (Frederick) and Co., Ltd.	25	0	0
British Timken, Ltd. (incl. Fischer Bearings Co., Ltd.)	25	0	0
British United Shoe Machinery Co., Ltd., The	25	0	0
Copper and Alloys, Ltd.	25	0	0
Gardner (Henry) and Co., Ltd.	25	0	0
Messina (Transvaal) Development Co., Ltd., The	25	0	0
Plessey Co., Ltd., The	25	0	0
Pressed Steel Co., Ltd.	25	0	0
Roe (A. V.) and Co., Ltd.	25	0	0
Thompson (John) (Wolverhampton), Ltd.	25	0	0
Metal Box Co., Ltd., The	22	1	0
Almin, Ltd.	21	0	0
Bull's Metal and Melloid Co., Ltd.	21	0	0
Delta Metal Co., Ltd., The (incl. Heaton and Dugard, Ltd.)	21	0	0
Leigh and Sillavan, Ltd.	21	0	0
Monotype Corporation, Ltd., The	21	0	0
Philips Electrical, Ltd.	21	0	0
*International Alloys, Ltd.	20	19	0
*Allen (Edgar) and Co., Ltd.	20	0	0
Brightside Foundry and Engineering Co., Ltd., The	20	0	0
*British Tin Smelting Co., Ltd., The	20	0	0
Deloro Stellite, Ltd.	20	0	0
*General Electric Co., Ltd., The	20	0	0
Hall and Pickles, Ltd.	20	0	0
Hard Metal Tools, Ltd.	20	0	0
Park Gate Iron and Steel Co., Ltd., The	20	0	0
*Phosphor Bronze Co., Ltd., The	20	0	0
Renfrew Foundries, Ltd.	20	0	0
*Saunders-Roe, Ltd.	20	0	0
Société Anonyme pour l'Industrie de l'Aluminium (Lausanne, Switzerland)	20	0	0
Wickman, Ltd.	20	0	0
*Chase Non-Ferrous Metal Co., Ltd.	19	0	6
*Curran (Edward) Engineering, Ltd.	19	0	6
*Rollet (H.) and Co., Ltd.	19	0	6
*Wolverhampton Die-Casting Co., Ltd., The	19	0	6
Marconi's Wireless Telegraph Co., Ltd.	15	15	0
Perry Barr Metal Co., Ltd.	15	0	0
*Scottish Non-Ferrous Tube Industries, Ltd.	15	0	0
*Loewy Engineering Co., Ltd., The	14	5	9
Brotherhood (Peter), Ltd.	12	12	0
Bridge Foundry Co., Ltd., The	10	10	0
Carborundum Co., Ltd., The	10	10	0
Clifford (Charles) and Son, Ltd.	10	10	0
Crabtree (J. A.) and Co., Ltd.	10	10	0
Fry's Diecastings, Ltd.	10	10	0
G. W. B. Electric Furnaces, Ltd.	10	10	0
Glynn Brothers, Ltd.	10	10	0
Harrison (Birmingham), Ltd.	10	10	0
Highton and Son, Ltd.	10	10	0
Hills (West Bromwich), Ltd.	10	10	0
Hoover, Ltd.	10	10	0
Incandescent Heat Co., Ltd., The (incl. Controlled Heat and Air, Ltd.; Metal Porcelains, Ltd.; Metalectric Furnaces, Ltd.; and Selas Gas and Engineering Co., Ltd.)	10	10	0
Johnson (Richard), Clapham and Morris, Ltd.	10	10	0
Lawley (Thomas), Ltd., and Jones and Rooke (1948), Ltd.	10	10	0
Maudslay Motor Co., Ltd., The	10	10	0
Metal Information Bureau, Ltd.	10	10	0

Donor	Gross, after Recovery of Tax by the Institute		
	£	*s.*	*d.*
Metal Products Co. (Willenhall), Ltd., The	10	10	0
New Metals and Chemicals, Ltd.	10	10	0
Newey and Tayler, Ltd. (incl. Newey Brothers, Ltd., and D. F. Tayler and Co., Ltd.)	10	10	0
North Thames Gas Board	10	10	0
Platers and Stampers, Ltd.	10	10	0
Rover Co., Ltd., The	10	10	0
Vauxhall Motors, Ltd.	10	10	0
Wright, Bindley and Gell, Ltd.	10	10	0
*Central Marine Engine Works (William Gray and Co., Ltd.)	10	0	0
Crittall Manufacturing Co., Ltd., The	10	0	0
Electric Furnace Co., Ltd. (incl. Electro Chemical Engineering Co., Ltd.)	10	0	0
Electric Resistance Furnace Co., Ltd.	10	0	0
Elliott Brothers (London), Ltd.	10	0	0
*Harland Engineering Co., Ltd., The	10	0	0
Newton, Chambers and Co., Ltd.	10	0	0
Rothschild (N. M.) and Sons	10	0	0
Strebor Diecasting Co., Ltd.	10	0	0
West Yorkshire Foundries, Ltd. (Subsidiary of Leyland Motors, Ltd.)	10	0	0
*Wild-Barfield Electric Furnaces, Ltd.	10	0	0
*Winfields Rolling Mills, Ltd.	10	0	0
*Bound Brook Bearings (G.B.), Ltd.	9	10	6
*Hoyt Metal Company of Great Britain, Ltd., The	9	10	6
*Hunt and Mitton, Ltd.	9	10	6
*Jenkinson (W. G.), Ltd.	9	10	6
*Shaw, Son and Greenhalgh, Ltd.	9	10	6
*Stein (John G.) and Co., Ltd.	9	10	6
*Betts and Co., Ltd.	8	0	0
*Glenfield and Kennedy, Ltd.	8	0	0
United Wire Works (Birmingham), Ltd.	6	6	0
Acton Bolt, Ltd.	5	5	0
Barnard (H. B.) and Sons, Ltd.	5	5	0
Bawn (W. B.) and Co., Ltd.	5	5	0
Blackwells Metallurgical Works, Ltd.	5	5	0
Cheswick and Wright, Ltd.	5	5	0
Cohen (Leonard), Ltd.	5	5	0
Corfield-Sigg, Ltd.	5	5	0
Dennison Watch Case Co., Ltd.	5	5	0
Easdale (R. M.) and Co.	5	5	0
Electro-Alloys, Ltd.	5	5	0
Electroflo Meters Co., Ltd.	5	5	0
Headley, Birch and Co., Ltd.	5	5	0
Jacks (William) and Co., Ltd.	5	5	0
Kemp (A.) and Son, Ltd.	5	5	0
Lead Wool Co., Ltd., The	5	5	0
Metal Supplies, Ltd.	5	5	0
Miles (John) and Partners (London), Ltd.	5	5	0
Oakland Metal Co., Ltd.	5	5	0
Ratcliffs (Great Bridge), Ltd.	5	5	0
Rigby (John) and Sons, Ltd.	5	5	0
Templewood Engineering Co., Ltd.	5	5	0
Vincent Engineering Co., Ltd.	5	5	0
Wilkinson (John) and Sons (Saltley), Ltd.	5	5	0
Cambridge Instrument Co., Ltd.	5	0	0
Langley Alloys, Ltd.	5	0	0
Blakeborough (J.) and Sons, Ltd.	4	4	0
Kincaid (John G.) and Co., Ltd.	4	4	0
*Acorn Anodising Co., Ltd.	†3	16	2
*Platt Metals, Ltd.	3	16	2
Mining and Chemical Products, Ltd.	3	3	0
Walterisation Co., Ltd., The	3	3	0
Beryllium and Copper Alloys, Ltd.	2	2	0
Carobronze, Ltd.	2	2	0
Follsain-Wycliffe Foundries, Ltd.	2	2	0
Gascoignes Non-Ferrous Foundries, Ltd.	2	2	0
Premier Cooler and Engineering Co., Ltd., The	2	2	0
Sheffield Testing Works, Ltd., The	2	2	0
Tungum Sales Co., Ltd.	2	2	0
Jenks (E. P.), Ltd.	1	1	0

* Annual Donation, Under Covenant, for Not Less Than 7 Years.

† Tax recoverable, but not actually recovered in the financial year 1951–52.

APPENDIX III

Committees

The main committees of the Institute which have served during the year were constituted as follows at 31 December 1952:

FINANCE AND GENERAL PURPOSES COMMITTEE

Baer, Mr. Alfred (*Chairman*).
Bailey, Mr. G. L.
Bolton, Mr. E. A.
Dorey, Dr. S. F.
Graham, Mr. A. B.
Murphy, Professor A. J.
Smout, Sir Arthur.
Tasker, Mr. H. S.

Ex-officio :

Smithells, Dr. C. J. (*President*).
Thompson, Professor F. C. (*Senior Vice-President*).
Jones, Mr. E. H. (*Honorary Treasurer*).
Smith, Mr. Christopher (*Chairman, Publication Committee*).

LOCAL SECTIONS COMMITTEE

Bolton, Mr. E. A. (*Chairman*).
Ashton, Mr. A. B.
Garside, Dr. J. E.
Kennett, Dr. S. J.
Walton, Mr. J. S.

Ex-officio :

Smithells, Dr. C. J. (*President*).
Thompson, Professor F. C. (*Senior Vice-President*).
Jones, Mr. E. H. (*Honorary Treasurer*).
Symonds, Mr. H. H. (*Chairman, Birmingham Local Section*).
Matthews, Mr. A. W. (*Honorary Secretary, Birmingham Local Section*).
Ransley, Dr. C. E. (*Chairman, London Local Section*).
Rhodes, Dr. E. C. (*Honorary Secretary, London Local Section*).
Finniston, Dr. H. M. (*Chairman, Oxford Local Section*).
Frost, Dr. B. R. T. (*Honorary Secretary, Oxford Local Section*).
Fowler, Mr. E. A. (*Chairman, Scottish Local Section*).
Hay, Mr. Matthew (*Honorary Secretary, Scottish Local Section*).
Hallett, Mr. M. M. (*Chairman, Sheffield Local Section*).
MacDougall, Mr. A. J. (*Honorary Secretary, Sheffield Local Section*).
Spring, Mr. K. M. (*Chairman, South Wales Local Section*).
Grenfell, Mr. W. H. (*Honorary Secretary, South Wales Local Section*).

MEDAL COMMITTEE

President (*Chairman*).
Senior Vice-President.

and

Not more than four Institute of Metals (Platinum) Medallists who are, or have been, Members of Council (to be selected by the President), with power to the President to co-opt not more than two other persons.

METAL PHYSICS COMMITTEE

Quarrell, Professor A. G. (*Chairman*).
Bowden, Dr. F. P.
Cottrell, Professor A. H.
Finniston, Dr. H. M.
Frank, Dr. F. C.
Hignett, Mr. H. W. G.
Hume-Rothery, Dr. W.
King, Mr. Ronald.
Lomer, Mr. W. M.
McLean, Mr. D.
Nabarro, Mr. F. R. N.
Nutting, Dr. J.
Oliver, Mr. D. A. (representing the Iron and Steel Institute and the British Iron and Steel Research Association).
Raynor, Professor G. V.
Richards, Dr. T. Ll.
Rotherham, Mr. L.

Ex-officio :

Smithells, Dr. C. J. (*President*).
Smith, Mr. Christopher (*Chairman, Publication Committee*).

METALLURGICAL ENGINEERING COMMITTEE

Thomas, Mr. W. J. (*Chairman*).
Baker, Mr. W. A.
Bolton, Mr. E. A.
Bowman, Mr. W. H.
Campbell, Mr. D. F.
Davies, Mr. C. E.
Ford, Professor H.
Miller, Mr. H. J.
Paton, Mr. Charles.
Salter, Mr. J.
Singer, Dr. A. R. E.
Swindells, Dr. N.
Walton, Mr. J. S.
Wilkinson, Mr. R. G.

Ex-officio :

Smithells, Dr. C. J. (*President*).
Smith, Mr. Christopher (*Chairman, Publication Committee*).

NOMINATIONS COMMITTEE

Smithells, Dr. C. J. (*President*) (*Chairman*).
Murphy, Professor A. J. (*Past-President*).
Tasker, Mr. H. S. (*Past-President*).
Thompson, Professor F. C. (*Senior Vice-President*).

PUBLICATION COMMITTEE

Smith, Mr. Christopher (*Chairman*).
Baker, Mr. W. A.
Bolton, Mr. E. A. (representing the Local Sections Committee).
Ford, Professor H.
Fox, Dr. F. A.
Gadd, Mr. E. R.
Inglis, Dr. N. P.
Jenkins, Dr. Ivor.
Parker, Dr. R. T.
Phillips, Mr. H. W. L.
Powell, Mr. A. R.
Raynor, Professor G. V.
Thompson, Professor F. C.
Turner, Mr. T. H.

Ex-officio :

Smithells, Dr. C. J. (*President*).
Baer, Mr. Alfred (*Chairman, Finance and General Purposes Committee*).
Jones, Mr. E. H. (*Honorary Treasurer*).
Quarrell, Professor A. G. (*Chairman, Metal Physics Committee*).
Thomas, Mr. W. J. (*Chairman, Metallurgical Engineering Committee*).

REPORT OF THE HONORARY TREASURER

FOR THE YEAR ENDED 30 JUNE 1952

THE accounts for the year show the continued growth of the Institute and its activities.

It will be seen that excess of expenditure over normal income increased from £2406 in 1950–51 to £6917 in 1951–52. This is some £1000 less than was forecast in the official estimates for the year under review; a direct comparison of the estimates with the accounts shows this figure as £2039, but the estimates provided for the printing of two books, at a cost of £975, which were delayed. The excess expenditure has been met by transfer from the Industrial Donations Fund. Costs of meetings and miscellaneous activities rose by £1941, while direct expenditure on publications (excluding salaries and overheads) increased by £7066. Printing trade charges for composition, machining, and binding were $17\frac{1}{2}$ per cent. higher than in the previous year. The printing order rose from 54,000 to 57,000 copies for the financial year 1951–52 and the *Journal* was enlarged to allow the printing of the *Bulletin*, a more ambitious form of the old News and Announcements Section. The change of format cost an estimated £800 in the year, but this sum was much more than recovered in increased revenue arising from the change.

Some £3000 of the increase in expenditure was due to the cost of a campaign to publish an accumulation of accepted papers, which was causing delay unfair to the authors. This operation has gone according to plan, the number of papers printed in the financial year 1951–52 being 75, compared with 52 in the financial year 1950–51, but it will take about half the present financial year to complete the scheme. The current year will show a further increase in general expenses; the salaries of the staff have been raised to make them more comparable with equivalent positions in other London offices; on the other hand, paper has fallen in price to some extent offsetting this. Analysis of the expenditure with the assistance of an outside accountant has confirmed the estimate of the permanent officers of the Institute that approximately 85 per cent. is attributable to costs of publication.

Income has increased by £4496. Of this amount, subscriptions and meeting receipts accounted for £1722 and publication sales and advertisements for £2737; the remainder, £37, came from higher investment yield. The chief sources of revenue continue to show increases up to the present date. The change in format of the *Journal* has been justified; membership has not fallen, and the anticipated increase from sales and advertising revenue is growing.

Apart from the Capper Pass and Mond Nickel Fellowships Funds, the total resources and current surplus, including the Endowment Fund and the Industrial Donations Fund, amount to £51,095, against £46,155 in the previous year. However, the raising of the bank rate lowered the market value of investments by some £7600, but as the intention is to keep these until redemption, the loss will not be realized and any investment of new money will yield more than before the change.

The Industrial Donations Fund, so successfully launched during the presidency of Mr. Tasker, is producing about £10,000 per year at present. It is likely that this generous contribution of industry will somewhat diminish over the next six years, at which time donations are due to cease. However, the Institute may hope to achieve, with reasonable prudence, very nearly if not quite a balanced budget by then.

THE INSTITUTE OF METALS

BALANCE SHEET AS AT 30 JUNE 1952

£]	30.6.51 £	£				
			RESERVES AND ACCUMULATED FUND			
	10,000		General Reserve		10,000	
	750		Special Publications Reserve		750	
	1,000		Investment Redemption Reserve		1,250	
	4,478		Entrance Fees		4,478	
16,228						16,478
			Accumulated Fund :			
	2,526		Balance as at 30 June 1951		2,276	
		516	Amount transferred from Industria Donations Fund		6,917	
		1,890	Amount transferable from War-Time Emergency Fund]		—	
	2,406					
	4,932				9,193	
		2,406	*Less* Excess of Expenditure over Income for the Year	6,917		
		250	,, Transfer to Investment Redemption Reserve	250		
	2,656				7,167	
2,276						2,026
18,504						18,504
			CURRENT LIABILITIES			
	1,778		Sundry Creditors		682	
	1,021		Subscriptions in Advance		2,654	
	6.356		Due to Industrial Donations Fund		585	
	—		Due to W. H. A. Robertson Fund		165	
9,155						4,086
21,295			ENDOWMENT FUND			21,295
6,356			INDUSTRIAL DONATIONS FUND			11,297
23,837			MOND NICKEL FELLOWSHIPS FUND			26,527
325			CAPPER PASS FUND			534
£79,472						£82,243

£	30.6.51 £	£				
			FIXED ASSETS			
			Office Furniture and Equipment :			
	1		Nominal Value as at 30 June 1947		1	
	642		Additions since, at cost		686	
	643				687	
	338		*Less* Depreciation to date		475	
	305				212	
			Library Books, &c. :			
	1		Nominal value as at 30 June 1947		1	
306						213
			INVESTMENTS			
			General Fund :			
17,792			Securities at cost or otherwise, as per schedule (Market value £15,183)			17,792
			CURRENT ASSETS			
	1		Publications at nominal value		1	
	250		Stock of stationery at valuation		500	
	1,890		Due from War-Time Emergency Fund		—	
			Debtors :			
		262	Subscriptions	250		
		25	*Less* Provision for doubtful subscriptions	50		
		237		200		
		3,003	Sundry debtors and payments in advance	3,079		
		32	W. H. A. Robertson Fund	—		
	3,272				3,279	
		4,089	Lloyds Bank, Ltd.	803		
		59	Cash in hand	2		
	4,148				805	
9,561						4,585
			ENDOWMENT FUND			
	21,293		Securities at cost, per schedule (Market value £16,361)		21,293	
	2		Lloyds Bank, Ltd., Deposit Account		2	
	21,295				21,295	
	—		*Less* Due to General Fund		—	
21,295						21,295
			INDUSTRIAL DONATIONS FUND			
	6,356		Due from General Fund		585	
	—		Securities at cost, per schedule (Market value £1,745)		1,950	
	—		Lloyds Bank, Ltd.		3,561	
	—		Post Office Savings Bank		500	
	—		Payments in advance for publications		4,701	
6,356						11,297
			MOND NICKEL FELLOWSHIPS FUND			
	16,650		Securities at cost, per schedule		17,300	
	1,562		Post Office Savings Bank		2,607	
	16		London Trustee Savings Bank		2,536	
	5,462		Lloyds Bank, Ltd., Deposit Account		3,708	
	21		Lloyds Bank, Ltd., Current Account		401	
	126		Interest accrued due		78	
	23,837				26,630	
	—		*Less* Creditor outstanding		103	
23,837						26,527
			CAPPER PASS FUND			
	320		Post Office Savings Bank		529	
	5		Interest accrued due		5	
325						534
£79,472						£82,243

REPORT TO THE MEMBERS OF THE INSTITUTE OF METALS

We have audited the above Balance Sheet dated 30 June 1952, and the annexed Income and Expenditure Account for the year ended 30 June 1952, and report that we have obtained all the information and explanations which to the best of our knowledge and belief were necessary for the purposes of our audit.

In our opinion proper books of account have been kept by the Institute, so far as appears from our examination of those books.

The above mentioned Balance Sheet and annexed Income and Expenditure Account are in agreement with the books of Account. In our opinion, and to the best of our information, and according to the explanations given us, the said Accounts give the information required by the Companies Act 1948, in the manner so required, and give a true and fair view, in the case of the Balance Sheet, of the state of the Institute's affairs as at 30 June 1952, and, in the case of the Income and Expenditure Account, of the excess of Expenditure over Income for the year ended 30 June 1952.

23 *October* 1952.

Poppleton and Appleby,
Chartered Accountants, Birmingham and London.

Approved on behalf of the Council: O. J. Smithells, *President.* A. Baer, *Chairman, Finance and General Purposes Committee.*
E. H. Jones, *Honorary Treasurer.* S. C. Guillan, *Secretary.*

THE INSTITUTE OF METALS

INCOME AND EXPENDITURE ACCOUNT FOR THE YEAR ENDED 30 JUNE 1952

30.6.51 £					£	£
			ESTABLISHMENT EXPENSES			
	1,011		Rent, Rates, Services, and Insurance		1,039	
	102		Lighting and Heating		190	
	86		Repairs and Renewals		123	
1,259						1,352
			ADMINISTRATIVE AND PRODUCTION EXPENSES			
	8,654		Salaries, Wages, and National Insurance		9,056	
	687		Superannuation Fund		769	
	403		Postages, Despatch, and Telephone		626	
	455		Printing and Stationery		498	
	109		Sundry Expenses		137	
	27		Bank Charges		18	
	53		Audit Fee		53	
	176		Professional Charges		157	
	265		Staff Travelling and Secretarial Expenses		365	
	104		Council and Committee Expenses		194	
10,933						11,873
722			*PENSIONS TO FORMER STAFF AND PAST SERVICE PREMIUMS*			783
			MEETINGS AND LOCAL SECTIONS			
	996		Meeting Expenses (Gross)		1,838	
			Local Sections and Associated Societies:			
	428		Grants, Programmes and Travelling		497	
	—		Students Essay Prize		21	
1,424						2,356
			JOINT ACTIVITIES			
	70		Official Entertaining		87	
	500		Joint Library Contribution		500	
	150		Joint Committee on National Certificates in Metallurgy		38	
	50		Joint Committee on Metallurgical Education		25	
	53		Subscriptions to other Bodies		52	
823						702
22			*BAD AND DOUBTFUL SUBSCRIPTIONS*			50
129			*DEPRECIATION: FURNITURE AND EQUIPMENT*			137
15,312						17,253
			LESS PROPORTION ATTRIBUTABLE TO PUBLICATIONS			
	5,228		Direct Salaries, Superannuation, and Expenses		5,670	
	7,011		Indirect Salaries, Superannuation, and Overheads		7,387	
2,239						13,057
3,073			**EXPENDITURE ATTRIBUTABLE TO ACTIVITIES OTHER THAN PUBLICATIONS**			4,196
			PUBLICATIONS			
			Journal and Metallurgical Abstracts:			
	13,102		Production Costs		18,349	
	4,429		Direct Salaries, Superannuation, and Expenses		4,743	
	5,971		Indirect Salaries, Superannuation, and Overheads		6,199	
	180		Sales Despatch		210	
	23,682				29,501	
			Special Publications:			
		979	Production Costs and Royalties	2,673		
		799	Direct Salaries, Superannuation, and Expenses	927		
		1,040	Indirect Salaries, Superannuation, and Overheads	1,188		
		79	Sales Despatch	174		
	2,897				4,962	
26,579			**EXPENDITURE ATTRIBUTABLE TO PUBLICATIONS**			34,463
£29,652						£38,659

£	30.6.51 £		£	
12,537		*MEMBERS' SUBSCRIPTIONS AND DONATIONS*		13,391
		INTEREST ON INVESTMENTS		
	496	General Fund	533	
	769	Endowment Fund	769	
1,265				1,302
726		*MEETING RECEIPTS*		1,594
		PUBLICATIONS		
		Journal and Metallurgical Abstracts:		
	4,531	Sales, net	5,051	
	5,191	Advertisements, *less* general charges	6,904	
	—	Postages recovered	279	
	9,722		12,234	
		Special Publications:		
	2,737	Sales, net	3,109	
	259	Postages recovered	112	
12,718				15,455
2,406		*EXCESS OF EXPENDITURE OVER INCOME FOR THE YEAR*		6,917
£29,652				38,659

THE INSTITUTE OF METALS

FUNDS ACCOUNTS FOR THE YEAR ENDED 30 JUNE 1952

ENDOWMENT FUND

30.6.51 £		£	30.6.51 £		£
769	To Transfer to General Fund—Investment Interest	769	21,295	By Balance at 30 June 1951	21,295
21,295	,, Balance at 30 June 1952	21,295	769	,, Interest on Investments and Bank Interest	769
£22,064		£22,064	£22,064		£22,064

INDUSTRIAL DONATIONS FUND

£		£	£		£
220	To Cost of Appeals, Administration and Stamp Duty	354	—	By Balance at 30 June 1951	6,356
—	,, 10-Year Index to Metallurgical Abstracts	21	7,092	,, Donations Received (including Tax recovered)	10,006
516	,, Amount transferred to General Fund, being Excess of Expenditure over Income for the year ended 30 June 1952	6,917	—	,, Income Tax Recovered (previous year)	2,188
6,356	,, Balance at 30 June 1952	11,297	—	,, Interest on Investments	39
£7,092		£18,589	£7,092		£18,589

MOND NICKEL FELLOWSHIPS FUND

£		£	£		£
3,723	To Grants to Fellows, including travelling	4,398	20,557	By Balance at 30 June 1951	23,837
34	,, Printing, Stationery, Postages, and Publicity	124	7,000	,, Donations received	7,000
4	,, Bank Charges	3	318	,, Interest on Investments and Bank Interest	465
27	,, Professional Charges	—			
250	,, Secretarial Expenses	250			
23,837	,, Balance at 30 June 1952	26,527			
£27,875		£31,302	£27,875		£31,302

CAPPER PASS FUND

£		£	£		£
200	To Grants	—	316	By Balance at 30 June 1951	325
325	,, Balance at 30 June 1952	534	200	,, Donations received	200
			9	,, Bank Interest	9
£525		£534	£525		£534

W. H. A. ROBERTSON FUND

£		£	£		£
30	To Balance at 30 June 1951 due to The Institute of Metals	32	—	By Donations received	200
2	,, Expenditure during year	3	32	,, Balance at 30 June 1951 due to The Institute of Metals	—
—	,, Balance at 30 June 1952 owing by The Institute of Metals	165			
£32		£200	£32		£200

SCHEDULE OF SECURITIES, 30 JUNE 1952

£	£		£	£
		General Fund :		
	5,500	3% Savings Bonds, at cost	5,500	
	1,010	2½% Treasury Stock, at cost	1,010	
	4,734	4% Consolidated Stock, at cost	4,734	
	5,057	2½% National War Bonds, at cost	5,057	
	500	2½% Defence Bonds, at cost	500	
	162	£1 War Savings (see Note 1)	162	
	829	4% Consolidated Loan (see Note 2)	829	
17,792				17,792

Note 1. At cost plus accrued interest to 22.11.50.
Note 2. Transferred at Market Value on 22.11.50 from the War-Time Emergency Fund to the General Fund.

17,792 carried forward			Carried forward	17,792

£	£			£	£
17,792 brought forward				*Brought forward*	17,792
			Endowment Fund :		
	525	£525	2½% Defence Bonds, at cost	525	
	1,285	£1,285	3% Savings Bonds, at cost	1,285	
	19,483	£17,926 5*s.* 0*d.*	4% Consolidated Loan, at cost	19,483	
21,293					21,293
			Industrial Donations Fund :		
	950	£950	3% Funding Stock, at cost	950	
	—	£1,012 9*s.* 9*d.*	4% Funding Stock 1960/90, at cost	1,000	
950					1,950
			Mond Nickel Fellowships Fund :		
	6,900		Woolwich Equitable Building Society	7,550	
	1,250		Co-operative Permanent Building Society	1,250	
	3,500		Halifax Building Society	3,500	
	5,000		Abbey National Building Society	5,000	
16,650					17,300
£56,685					£58,335

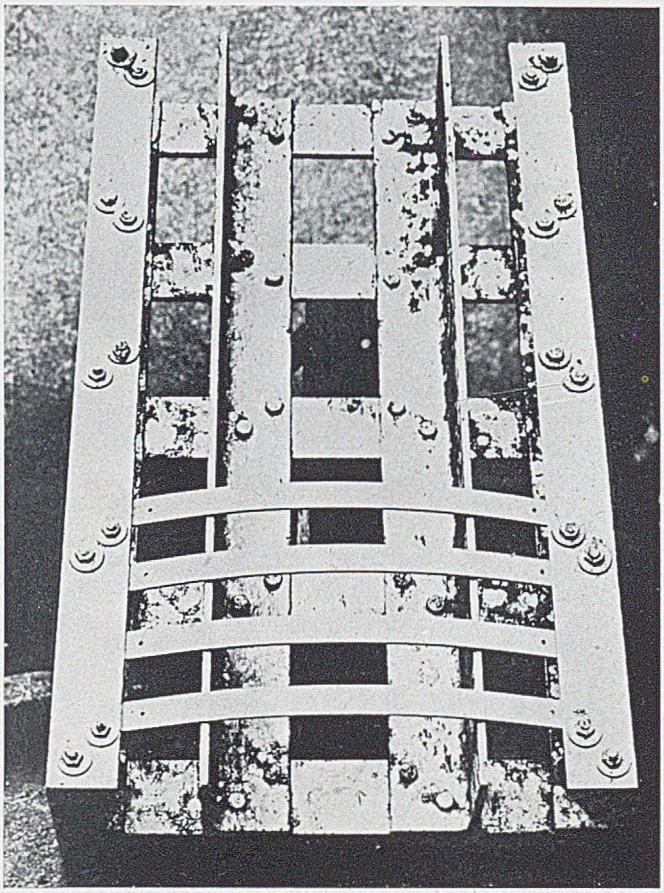
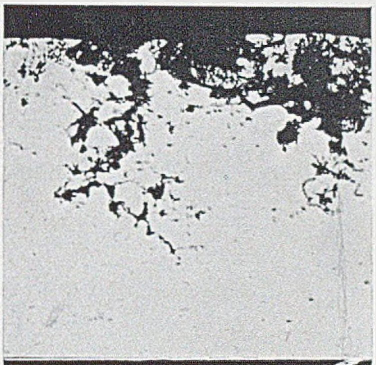


FIG. 1.—Stress-Corrosion Testing Frame for Field Tests.

MICROSECTIONS OF H10-WP ALLOY SPECIMENS AFTER 2 YEARS' EXPOSURE. × 100.



2

3

FIG. 2.—Site I. Industrial atmosphere at Sheffield.

FIG. 3.—Site II. Semi-industrial atmosphere at London.

MICROSECTIONS OF H10-WP ALLOY SPECIMENS AFTER 2 YEARS' EXPOSURE. × 100.

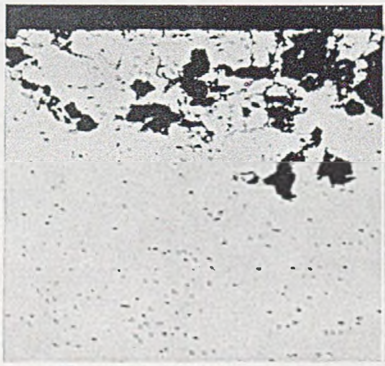


FIG. 4.—Site III. Marine atmosphere at Hayling Island.

FIG. 5.—Site IV. Rural atmosphere at Stoke Poges.

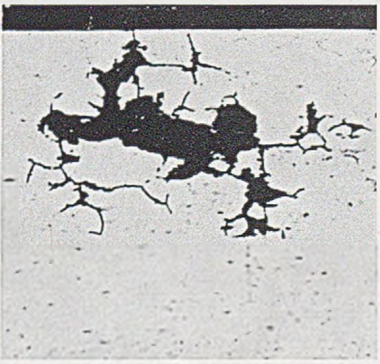
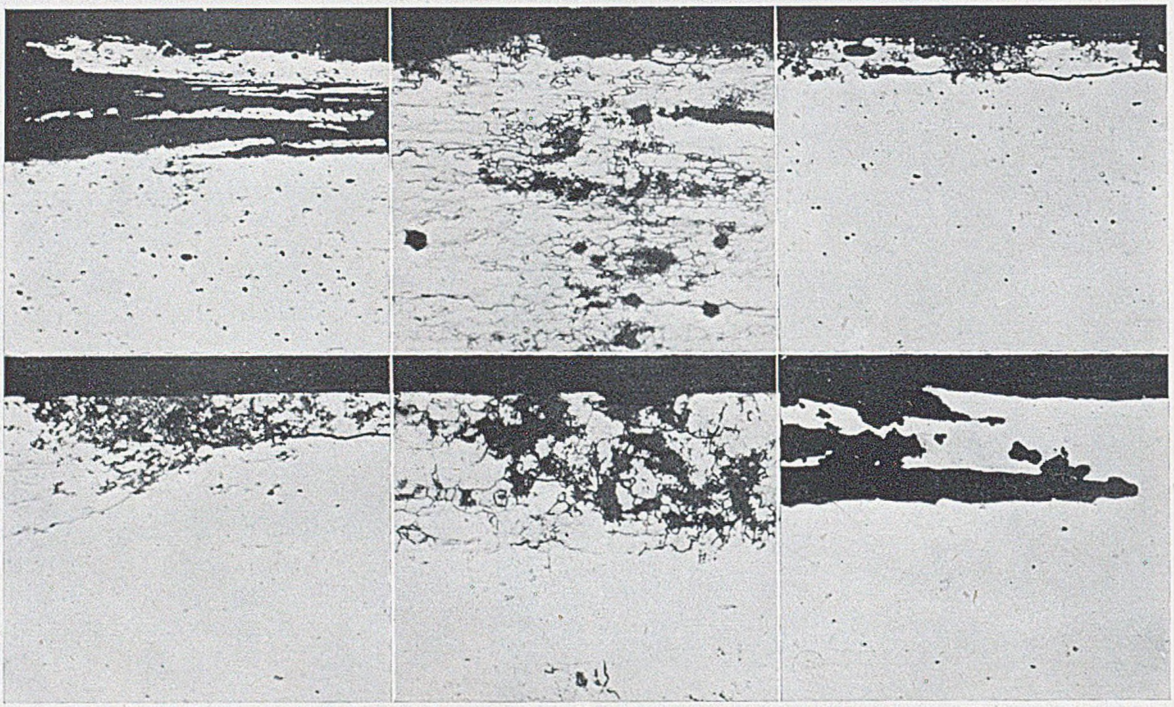


FIG. 6.—Site VII. Total immersion in fresh water at Christchurch.

MICROSECTIONS OF H15-WP ALLOY SPECIMENS AFTER 2 YEARS' EXPOSURE.

7 8 9



10 11 12

FIG. 7.—Site I. Industrial atmosphere at Sheffield. × 100.
FIG. 8.—As Fig. 7. Showing attack of sub-grain network. × 500.
FIG. 9.—Site II. Semi-industrial atmosphere at London. × 100.
FIG. 10.—Site III. Marine atmosphere at Hayling Island. × 100.
FIG. 11.—As Fig. 10. Showing attack of sub-grain network. × 500.
FIG. 12.—Site VIII. Intermittent immersion in fresh water at Christchurch. × 100.

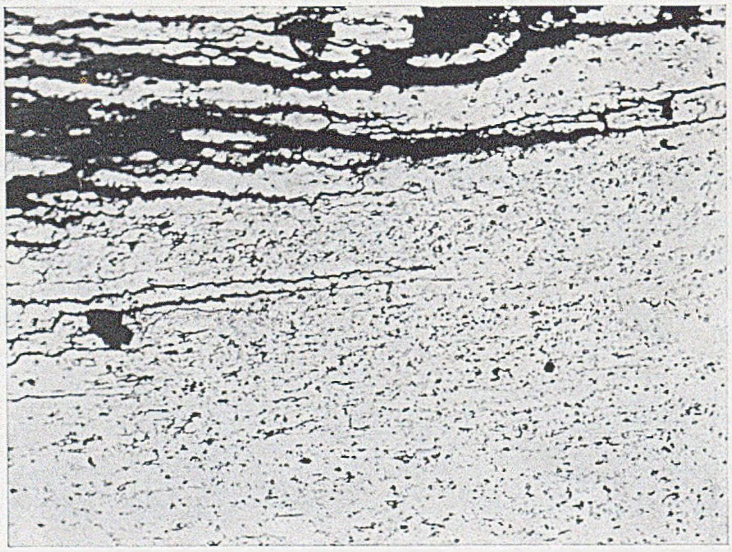


FIG. 13.—Foliation of an H15-WP Extrusion Specimen Taken from a Brine-Tank Frame. × 250.

Crystal Fragmentation in Aluminium During Creep.

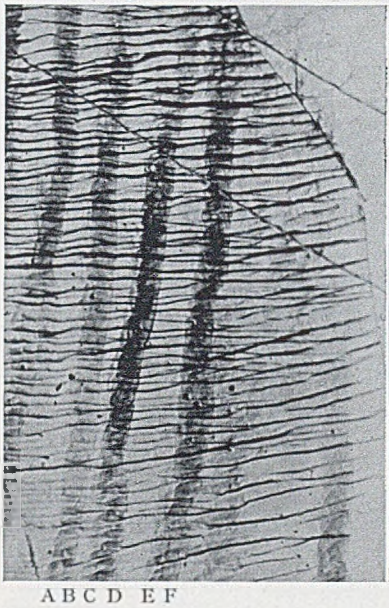


Fig. 1.—Banded Structure on Surface of Coarse-Grained Specimen Extended 13·8%. Photographed in focus, oblique lighting. × 50.

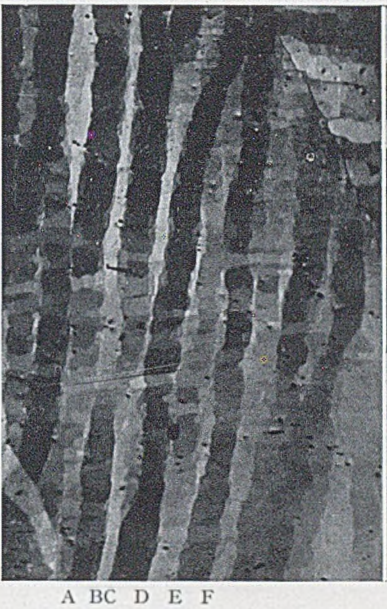


Fig. 2.—Same Field After Polishing Flat and Anodizing. Specimen now extended 23·6%. Photographed between crossed polaroids. × 50.

Fig. 3.—Relatively Unbanded Structure in Fine-Grained Specimen Extended 50%. × 100.

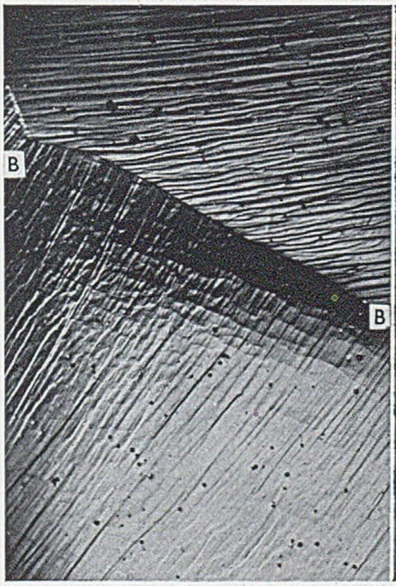


Fig. 4.—Polygonization in Coarse-Grained Specimen, Beginning First Near Grain-Boundary *BB*. The polygonized boundaries are those nearly parallel to *BB*; the other lines are slip bands. Specimen extended 4·02%. × 50.

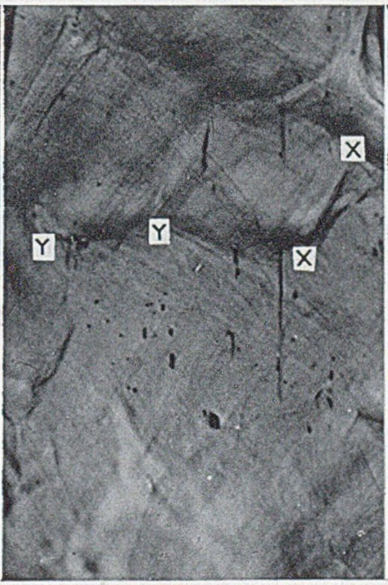


Fig. 5.—Displacement *XX* and *YY* at Sub-Crystal Boundaries in Fine-Grained Specimen Extended 69%. Photographed in focus. × 750.

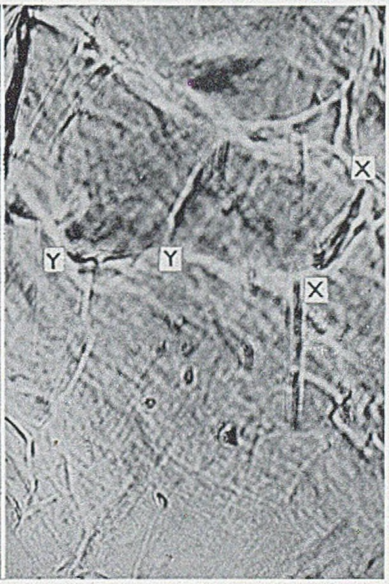


Fig. 6.—Same Field as Fig. 5, Stopped Down and Photographed Inside Focus to Show Polygonized Boundaries (Network of White Lines). × 750.

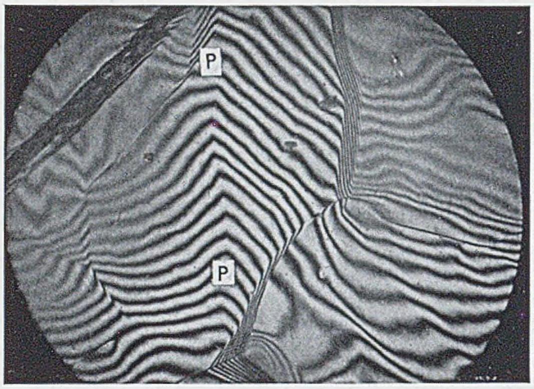


FIG. 7.—Surface Tilt at Sub-Boundary *PP* in a Fine-Grained Specimen Extended 11·9%. Interference micrograph. × 310.

GRAIN-BOUNDARY MIGRATION DURING CREEP OF ALUMINIUM.

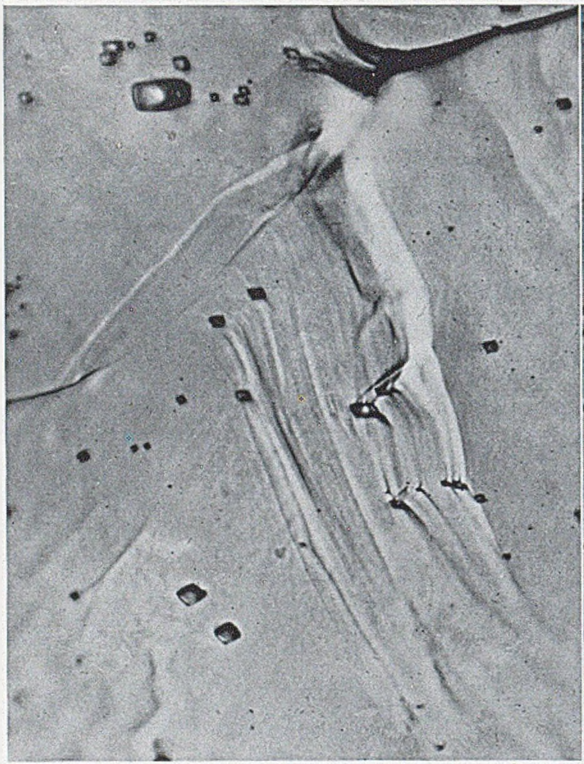


FIG. 11.—Grain Growth During Creep of Fine-Grained Specimen Extended 11·9%. The etch pits show that the left-hand grain has grown into the right-hand one. × 1250.

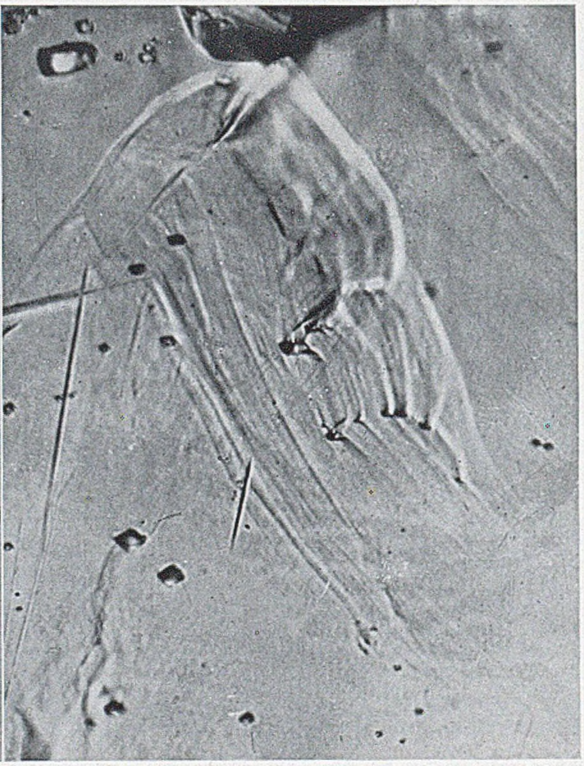
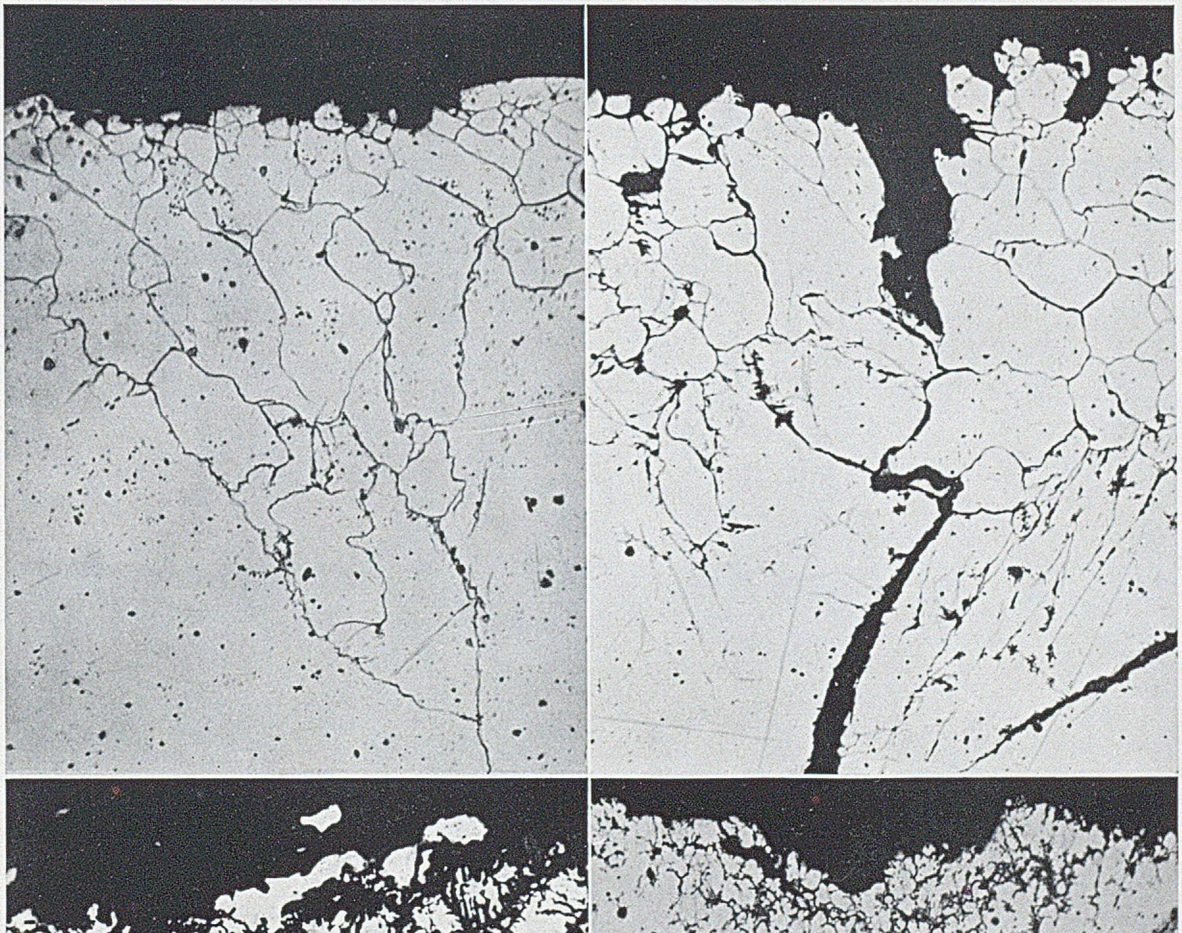


FIG. 12.—Same Field After 19·8% Extension. × 1250.

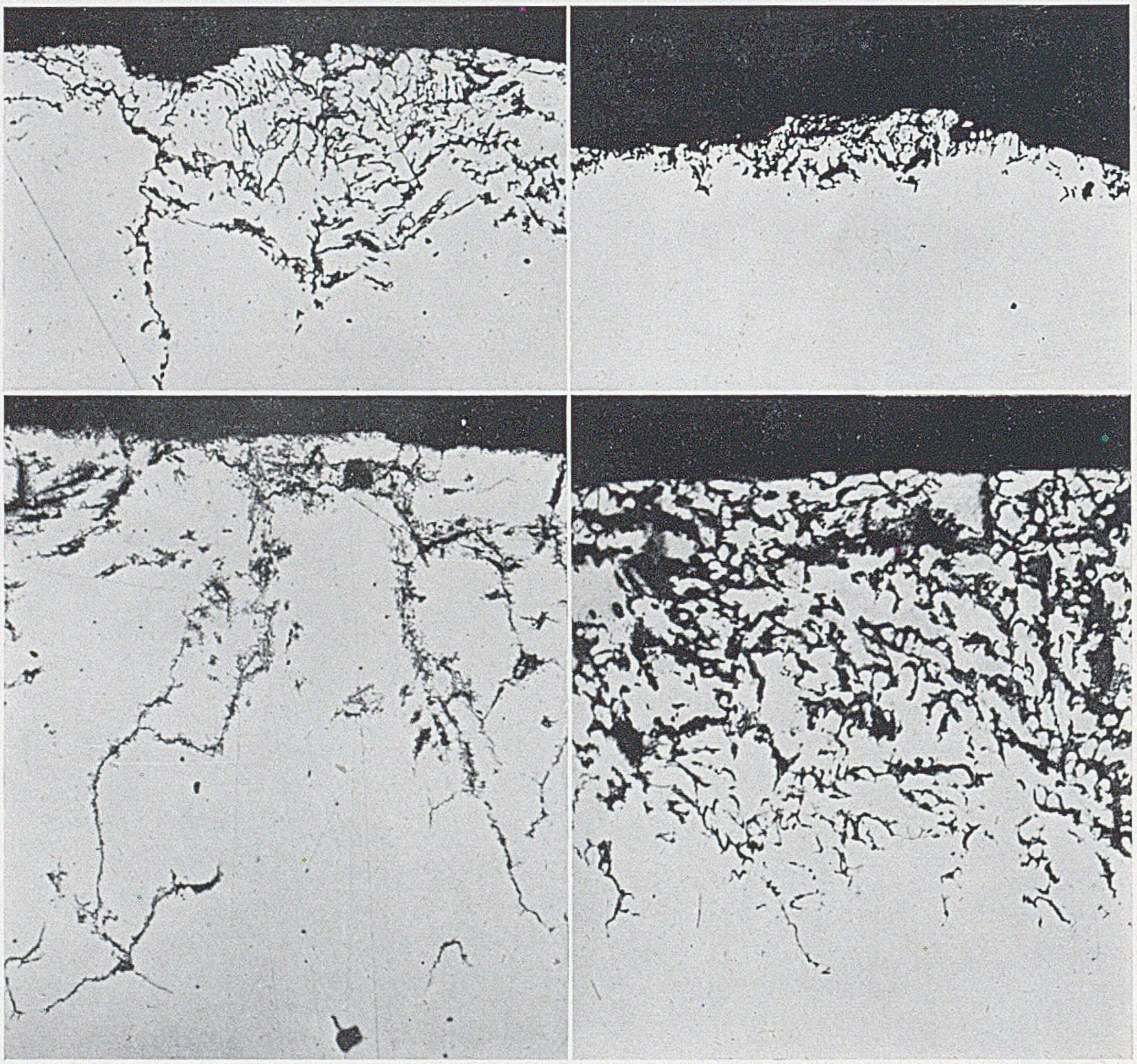
EXAMPLES OF INTERCRYSTALLINE CORROSION IN ZINC–ALUMINIUM ALLOYS.



FIGS. 1–5.—Cast Specimens After Exposure for 10 Days to an Air/Water-Vapour Atmosphere at 95° C. Unetched. × 250.

FIG. 1.—Alloy 2. S.H.P. Zinc with 0·075% Aluminium.
FIG. 2.—Alloy 6. S.H.P. Zinc with 0·075% Aluminium and 0·003% Lead.
FIG. 3.—Alloy 9. S.H.P. Zinc with 0·013% Lead and 0·04% Magnesium.
FIG. 4.—Alloy 26. S.H.P. Zinc with 4·44% Aluminium and 0·048% Magnesium.
FIG. 5.—Alloy 30. C.S. Zinc with 4·0% Aluminium and 0·04% Magnesium. (Pressure-die-cast Mazak.)

Examples of Intercrystalline Corrosion in Zinc–Aluminium Alloys.



Figs. 6–9.—Cast Specimens After Exposure for 10 Days to an Air/Water-Vapour Atmosphere at 95° C. Unetched. × 250.

Fig. 6.—Alloy 52. C.S. Zinc with 4% Aluminium and 0·3% Copper.
Fig. 7.—Alloy 57. C.S. Zinc with 4% Aluminium, 0·3% Copper, and 0·05% Magnesium.
Fig. 8.—Alloy 68. S.H.P. Zinc with 4% Aluminium and 0·01% Cadmium.
Fig. 9.—Alloy 38. S.H.P. Zinc with 4% Aluminium.

MICROSTRUCTURES OF CHROMIUM–MANGANESE ALLOYS.

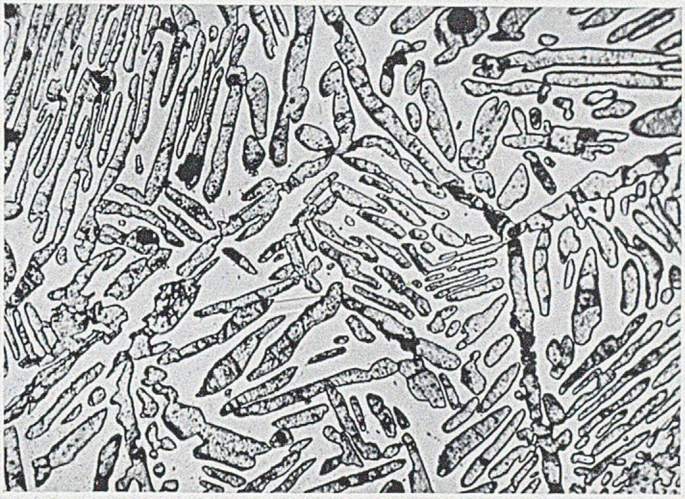


FIG. 3.—Alloy 60 After Annealing for 403 Hr. at 690° C. and Quenching. Particles of σ in a matrix of α-Cr solid solution. × 300.

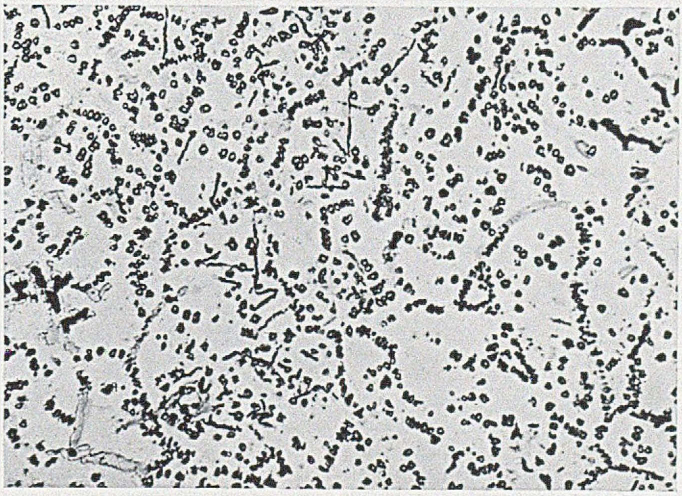


FIG. 4.—Alloy 53 After Annealing for 1018 Hr. at 675° C. and Quenching. A matrix of α-Cr, dark-etching crystals of α'-Mn, and light-etching crystals of σ. The distribution of α'-Mn particles in this alloy was uneven, and this photograph shows the region in which their relative number was greatest. × 200.

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 20 FEBRUARY 1953 PART 6

1 — PROPERTIES OF METALS

***Thermochemistry of the Rare Earths. I.—Cerium and Neodymium.** F. H. Spedding and Carl F. Miller (*J. Amer. Chem. Soc.*, 1952, **74**, (16), 4195–4198).—Heats of formation of Ce^{+++} and Nd^{++}, calculated from heats of soln. of the metals and chlorides, were −167·43 and −163·27 kg.cal./mole, resp. The resp. free energies of formation were −161·54 and 155·40 kg.cal./mole.—J. R.

***A Preliminary Study of the Oxidation and Vapour Pressure of Chromium.** (Gulbransen and Andrew). See col. 417.

***The Magnetic Susceptibility of Chromium.** T. R. McGuire and C. J. Kriessman (*Phys. Rev.*, 1952, [ii], **85**, (3), 452–454).—Using the body-force method, the susceptibility of 99·9% pure Cr has been found to increase from $3{\cdot}42 \times 10^{-6}$ at −195° C. to $4{\cdot}30 \times 10^{-6}$ e.m.u./g. at 1440° C. About ¼ of this increase occurred in the 1300°–1400° C. range with a temp. hysteresis, and is attributed to a possible phase transition. The results are discussed in terms of the various theories of the electron structures of the transitional metals. 20 ref.—P. C. L. P.

***Solubility of Nitrogen in Liquid Chromium and Melts of Chromium and Silicon.** V. S. Mozgovoi and A. M. Samarin (*Izvest. Akad. Nauk S.S.S.R.*, **1950**, [Tekhn.], (10), 1529–1536).—[In Russian]. The H.F. furnace used is described, and results of tests are given which show that the solubility of N in metallic Cr decreases as the temp. increases. The interaction of liq. Cr and N is accompanied by an exothermic reaction : $x(\mathrm{Cr}) + \frac{1}{2}\mathrm{N}_2 \rightleftharpoons (\mathrm{Cr}_x\mathrm{N})$, $\Delta H < 0$. The thermodynamic function of this reaction is $\Delta H = -7594{\cdot}5$ cal.; $\Delta F^\circ = -7594{\cdot}5 + 1{\cdot}27\,T$; $\lg K = (1660/T) - 0{\cdot}2782$. X-ray investigation indicated the presence of Cr_2N in alloys with high N content. It can be assumed that interaction of liq. Cr and N is accompanied by the reaction : $2\mathrm{Cr} + \frac{1}{2}\mathrm{N}_2 \rightleftharpoons \mathrm{Cr}_2\mathrm{N}$, $\Delta H = 7600$ cal. The solubility of N in molten Cr and Si ∝ Cr content, but diminishes, at first sharply, as the Si content of the molten metal is increased.—H. W.

***Sintered Copper with Extremely High Conductivity.** F. Pawlek (*Powder Met. Bull.*, 1951, **6**, (2), 83).—The elect. conductivity of pure sintered Cu was found to be between 0·606 and 0·613 microhm.$^{-1}$·cm.$^{-1}$, thus confirming previous values reported by Kieffer and Hotop. 5 ref.—W. A. M. P.

***Some Characteristics of the Mechanism of Sintering [of Copper].** G. C. Kuczynski (*Rev. Mét.*, 1952, **49**, (10), 733–735; discussion, 735–736).—K. discusses the growth of the contact zone when fine Cu wires wound round Cu cylinders are sintered for 25 hr. at 1050° C. (cf. *J. Appl. Physics*, 1950, **21**, 632; *M.A.*, **18**, 487) and shows that, in addn. to diffusion, some other factor, although only to a small extent, causes the higher rates of sintering of small particles, and that viscous flow plays only a negligible part. With two different metals, the mechanism of sintering appears to be also a matter of diffusion, but the effect of osmotic pressure cannot be neglected. In this case diffusion along the grain boundaries assumes greater importance.—J. H. W.

***Measurement of the Internal Energy in Copper Introduced by Cold Work.** Benjamin Welber (*J. Appl. Physics*, 1952, **23**, (8), 876–881).—The retained energy in twisted Cu bars was measured in a calorimeter basically similar to Quinney and Taylor's (*Proc. Roy. Soc.*, 1937, [A], **163**, 15; *M.A.*, **5**, 37). The ratio of retained energy, E, to work expended in twisting, W, remained const. at ~0·03 for $W = 8$–12 cal./g. This is lower than Quinney and Taylor's ratio (0·09).—R. W. C.

***Properties of Thermally Produced Acceptors in Germanium.** C. S. Fuller, H. C. Theuerer, and W. van Roosbroeck (*Phys. Rev.*, 1952, [ii], **85**, (4), 678–679).—A letter. The effect of thermal treatment in shifting the position of a p–n junction in Ge has been investigated and conclusions drawn about the diffusion rates of acceptors. Measurements of the variation of the acceptor concentration developed by prolonged heating at temp. in the range 560°–760° C. have been used to estimate the activation energy of formation.—P. C. L. P.

***Zinc as an Acceptor in Germanium.** W. C. Dunlap, Jr. (*Phys. Rev.*, 1952, [ii], **85**, (5), 945–946).—A letter. Small amounts of Zn in a single crystal of Ge were found to increase the activation energy for the excitation of carriers much more than similar amounts of In. Preliminary diffusion experiments show that Zn diffuses much more rapidly in Ge than does In, Ga, or Al.—P. C. L. P.

***Changes in Work-Functions of Vacuum-Distilled Gold Films.** Chung Fu Ying and H. E. Farnsworth (*Phys. Rev.*, 1952, [ii], **85**, (3), 485–486).—Cf. Clarke and F., *ibid.*, p. 484; *M.A.*, **20**, 388. A letter. The work-function, ϕ, of a Au film freshly prepared by evaporation in a high vacuum was found to be lower than that of the clean polycryst. Au on to which it was deposited. The variation of ϕ with film thickness, and the way in which ϕ attained the value characteristic of massive Au on subsequent heating were investigated and found to be similar to the behaviour of Ag films.
—P. C. L. P.

***Internal Friction of Iron and Nickel. II.—Internal Friction and Plastic Deformation of Ferromagnetic Pure Metals.** Makoto Osawa (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 15–18).—[In Japanese]. The internal friction of pure Fe and Ni subjected to plastic deformation by tensile and simple and repeated torsion was measured by means of the resonance of longitudinal oscillations. O. found that: (1) the internal friction decreased rapidly with plastic deformation at a very small degree of working and (2) there was a peak friction. It is suggested that (1) was due to ferromagnetism and (2) was due to dislocations.—AUTHOR.

***The Atomic Heat of Lead in the Region of Its Transition to Supraconductivity.** J. R. Clement and E. H. Quinnell (*Phys. Rev.*, 1952, [ii], **85**, (3), 502–503).—A letter. Measurement of the atomic heat of 99·996% pure Pb showed a discontinuity at the supraconducting transition temp., contrary to the indications obtained by previous workers. The magnitude of the discontinuity agrees with that deduced from the variation of the transition temp. with magnetic field. —P. C. L. P.

***Wave-Function of Ionized Lithium.** P. J. Luke, R. E. Meyerott, and W. W. Clendenin (*Phys. Rev.*, 1952, [ii], **85**, (3), 401–409).—A method is developed for the accurate calculation of the wave-function for the $1s2s\ {}^3S$ state of singly ionized Li. It leads to agreement with the observed ionized energy of the $2s$ electron.—P. C. L. P.

***Heat-Transfer to Mercury [Under Turbulent Flow Conditions].** E. R. Gilliland, R. J. Musser, and W. R. Page (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 402–404).—The necessity of extracting large amounts of heat from nuclear reactors has directed attention to the heat-transfer characteristics of liq. and molten metals. The forced-convection heat-transfer to Hg under turbulent flow conditions was studied in a closed circuit in which the Hg was both heated and cooled. Results for the heating section are alone discussed. The results are presented in the form of a graph of Nusselt numbers of the flow as ordinates against the corresponding Péclet numbers as abscissæ. The results agree fairly well with those found by Trefethen (*U.S. Atomic Energy Commission Publ.*, 1950, (**NP-1788**)) and English (*Atomic Energy Research Establishment Rep.*, 1950, (**E/R547**), but are somewhat lower than those of Isakoff (*U.S. Atomic Energy Commission Publ.*, 1951, (**AECU-1199**); *M.A.*, **19**, 819) and cast some doubt on the validity of a correlating method for heat transfer to liq. metals which omits to take into account the effect of cross-sectional d variation. 7 ref.—J. S. G. T.

***Heat and Momentum Transfer in Turbulent Flow of Mercury.** Sheldon E. Isakoff and Thomas B. Drew (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 405–409).—Velocity and temp. profiles and coeff. of heat transfer were determined for Hg flowing in an elect.-heated vertical tube of dia. 1·5 in. and length 18·6 ft., and eddy diffusivities were computed from the profiles. The Reynolds modulus, N, of the Hg flow was 36,700-373,000. The ratio of the thermal eddy diffusivity : the momentum diffusivity $\propto N^{0{\cdot}46}$, and is a function of the relative position in the cross-section that is independent of N. The velocity profiles for Hg agree essentially with those for other fluids; temp. profiles differ from the predictions of Martinelli (*Trans. Amer. Soc. Mech. Eng.*, 1947, **69**, 947; *M.A.*, **16**, 116), and the heat-transfer coeff. though agreeing approx. with those calculated by means of the Martinelli–Lyon equation, increase more rapidly with increase of N than those equations predict. 16 ref. —J. S. G. T.

***Heat-Transfer to Mercury.** W. F. Cope (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 453–458).—A theory of heat transfer to a stream of Hg, taking into account a turbulent core in which molar transport dominates, a laminary layer near the wall in which molecular transport dominates, and a buffer layer in which both methods of transport are of importance, is developed and discussed, and calculated results are compared with experimental results obtained with streams of water, a gas, and Hg. The vorticity transfer theory disagrees by a factor of 2 (or more) : 1 with measurements made on water and gases, and cannot be used to predict the heat-transfer coeff. of Hg. The several formulæ proposed differ widely in their prediction of the transfer coeff. of Hg. Full wetting of the flow tube is tacitly assumed in deriving the various heat-transfer formulæ, but is difficult to ensure. —J. S. G. T.

***Heat-Transfer Properties of Mercury.** D. English and T. Barrett (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 458–460).—A survey of literature relating to heat transfer to streams of Hg revealed a lack of knowledge under conditions of high heat flow (500 W./cm.^{2}); the present work was undertaken to extend the range of experimental data. The heat was developed elect. in the thin wall of the flow tube by a heavy A.C. The results are shown in the form of graphs in which Stanton numbers ($h/\rho V c_p$) as ordinates are plotted against Reynolds numbers as abscissæ, and relate to Hg flow in stainless steel and Cu-plated Ni tubes. No limit was found to the heat transfer, except when the inside temp. of the tube exceeded the b.p. of Hg; a vapour lock then occurred and melted the tube locally. Successful heat transfer depended on inner surface cleanliness and a good bond between the inner tube and its Cu coating. The plots of Stanton number against Reynolds number for stainless steel agreed to within 10% with those for clean Ni. Runs for as short a time as 35 hr. in Ni tubes gave marked indications of pitting, loss of polish, possible fouling, and halving of the heat-transfer coeff. Similar decreases in the heat-transfer coeff. occurred in stainless-steel tubes after 100–200 hr., without noticeable pitting or fouling. The true useful effect of the wetting agent (traces of Mg or Ti) used in Hg boiler work is the removal of dirt or oxide films rather than actual wetting. —J. S. G. T.

***Electrical Conductivity of Thin Deposits of Platinum Evaporated *in Vacuo* on Dielectric Films.** Charles Feldman (*Compt. rend.*, 1952, **234**, (19), 1858–1860).—As a continuation of the work of Mostovetch and Vodar (*Semi-Conducting Materials* (Butterworths Scientific Publications), **1951**, 260; *M.A.*, **19**, 636) on the elect. conductivity of very thin films of metals evaporated *in vacuo*, preliminary results are given on the deposition of Pt on surfaces of KBr, CaF_2, and SiO_2, as part of an investigation into more complex films. The relation : $R = R_0 T^{-n} \exp(\theta - \theta' F^{\frac{1}{2}}/T)$ is in general confirmed (V. and M., *Compt. rend.*, 1948, **226**, 167; *M.A.*, **18**, 753; Darmois, M., and V., *ibid.*, 1949, **228**, 992; *M.A.*, **16**, 769). The differences between films on glass and films on glass + CaF_2 or KBr are a large increase of R_0 and higher values of θ and θ' with the latter. If Pt is evaporated on CaF_2 or KBr simultaneously forming a mixture, a very high value of θ is obtained. The explanations of these observations are discussed.—J. H. W.

***The Allotropic Transformation of Metallic Selenium.—I.–II.** Sueo Yamamori (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 1–6; 6–10).—[In Japanese, with graphs and diagrams in English]. [I.—] Qual. measurements have been made of the temp. dependence of conductivity, and differential thermal curves of specimens crystallized from vitreous Se, and the elect. properties of the deposited monoclinic Se have been obtained and the results studied. The following conclusions were reached: (1) the reversible thermal change and allotropic transformation occurs at ~209° C.; (2) hexagonal Se, stable below ~209° C., has a positive temp. coeff. of conductivity, whereas monoclinic Se, stable above ~209° C., shows a tendency to have a negative temp. coeff., especially on supercooling; and (3) the deposited monoclinic Se is intrinsically a good conductor, but because of occluded gases, probably CS_2, it behaves like an insulator. [II.—] The conductivity of metallic Se converted from the vitreous state was qual. investigated as a function of temp. From the results, Y. concludes that there are various types of elect. conductivity/temp. curves, the main difference between them being due to the existence of the monoclinic form mixed with the hexagonal Se. Metallographic examination confirmed that specimens having the microstructure which first appears on crystn. of vitreous Se give conductivity/temp. curves characteristic of hexagonal Se, but if this structure is broken down minutely by raising the temp. of crystn. or by increasing the heating time the hexagonal form transforms to the monoclinic even at a lower temp. than the transformation point (~209° C.). It is supposed that this microstructural change is due to internal strain resulting from the crystn.—AUTHOR.

***Observations on the Photoelectric Work-Functions and Low-Speed Electron Diffraction from Thin Films of Silver on the (100) Face of a Silver Single Crystal.** Edward N. Clarke and H. E. Farnsworth (*Phys. Rev.*, 1952, [ii], **85**, (3), 484–485).—A letter. Deposition of Ag on to the (100) face of a Ag single crystal lowers its work-function if the crystal temp. is below 65° C., but this reduction can be removed by heating to 80°–200° C.—P. C. L. P.

***The Drift Mobility of Electrons in Silicon.** J. R. Haynes and W. C. Westphal (*Phys. Rev.*, 1952, [ii], **85**, (4), 680).—A letter. Transit-time measurements indicate that the mobility of electrons injected into a *p*-type Si single crystal is $\sim$1210 cm.^{2}/sec. The much larger mobility obtained than that found by Pearson and Bardeen (*ibid.*, 1949, [ii], **75**, 865; *M.A.*, **16**, 635) is attributed to the inhomogeneities produced by the crystal boundaries in their samples.—P. C. L. P.

***Calculation of the *g* Factor of Metallic Sodium.** Y. Yafet (*Phys. Rev.*, 1952, [ii], **85**, (3), 478).—A letter. The effect of spin–orbit coupling on the electronic *g* factor deduced from spin resonance experiments on Na, is calculated to make $g_{\max.}$ $\sim$2·0019 for the strongest absorption.—P. C. L. P.

***Resistivity of Evaporated Tellurium Films.** Takemaro Sakurai and Seiji Munesue (*Phys. Rev.*, 1952, [ii], **85**, (5), 921).—A letter. The resistivity of Te films 200–5000 Å. thick which had been evaporated on to a SiO_2 substrate, was reversible with temp. only at temp. lower than that of the substrate during evaporation. At higher temp. or on electron bombardment the resistivity decreases, the deposit becoming more cryst. If the substrate is kept above 180° C., the elect. properties of the deposited film are similar to those of massive Te.—P. C. L. P.

***Experiments with Audiofrequencies on Supraconductors [Tin].** B. Serin and C. A. Reynolds (*Phys. Rev.*, 1952, [ii], **85**, (5), 938–939).—A letter. The account of the work given by S., R., Feldmeier, and Garfunkel (*ibid.*, 1951, [ii], **84**, 802; *M.A.*, **20**, 72) is amplified to answer Galkin and Bezuglyi's (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, 1145; *M.A.*, **20**, 236) criticisms of their conclusions, as previously reported in abstract form. Further experimental data and arguments are given to show that at sufficiently high frequencies the effective resistance of the samples is greater than normal. —P. C. L. P.

Metals for Tomorrow: Titanium. D. I. Brown (*Iron Age*, 1952, **170**, (15), 260–279; (16), 105–113; **Molybdenum,** (15), 280–281; **Silicon,** 282–283; **Selenium,** 285; **Zirconium,** 286–287; **Cerium,** 288; **Germanium,** 289; **Lithium,** 290).—The availability and prospects of the various metals are discussed. Ti is dealt with in considerable detail, especially as regards prodn. and casting, the remainder briefly.—J. H. W.

Republic Steel: New Titanium Producer. V. W. Whitmer and T. E. Perry (*Iron Age*, 1952, **170**, (8), 117–122).—W. and P. describe the prodn. and properties of pure (99·5%) Ti and two Ti alloys, one contg. Fe 2·0 and Cr 4·0%, and the other Mn 7%. The latter alloy has a yield strength after heat-treatment of 120,000–160,000 lb./in.^{2}.—J. H. W.

The Metallurgy of Uranium. E. W. Colbeck (*Metal Ind.*, 1952, **81**, (19), 361–363; (20), 387–389).—Read before the London Local Section of the Institute of Metals. C. discusses the occurrence of mineral U in economic deposits and the methods used for its prepn. in metallic form, and reviews its phys., chem., and mech. properties. 30 ref.—J. H. W.

***Anelasticity of Zinc.** Shuichiro Takahashi (*J. Appl. Physics*, 1952, **23**, (8), 866–868).—Single-crystal bars or very coarse-grained polycrystals of Zn were subjected to transverse vibrations in the strain range 10^{-7}–10^{-4}. Up to a limiting driving stress the behaviour was purely elastic and the internal friction const. Beyond this, the max. strain amplitude increased more slowly with driving stress, the resonance curve became asymmetric and the internal friction went up. Moreover, there were comparatively slow, roughly periodic variations of strain amplitude (periods of some sec.), which T. calls "gasping". This behaviour occurs when the stress somewhat exceeds the limiting stress mentioned above, which is taken to be the elastic limit for slip.—R. W. C.

Extremely High-Temperature Materials: Analysis of Their Properties and Techniques of Manufacture. Bernard P. Planner (*Western Metals*, 1952, **10**, (5), 43–45).—A review.—T. G.

The Properties of Metallic Coatings Produced by Evaporation and Sputtering. S. Tolansky (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (3), 11 pp.).—The prodn. of thin and highly uniform metallic coatings by cathodic sputtering and by thermal evaporation is described. The characteristics of the deposits produced are discussed, and so are some of the appn. of these techniques.—G. T. C.

The Theory of Breaking Strength. I.—Static Strength. Hideji Suzuki (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 19–23).—[In Japanese]. By thermodynamic consideration of a model, a formula connecting the breaking strength of a crystal with its thermodynamic functions was obtained. In the case of a simple tensile test of an internal-stress-free polycryst. specimen at 0° K., the formula agrees with that of Fürth (*Proc. Roy. Soc.*, 1941, [A], **177**, 217; *M.A.*, **8**, 130). The formula is successfully applied to tensile tests on a specimen with internal stress or under hydrostatic pressure, but it fails to evaluate the strength in the compression test.—AUTHOR.

***Variation of Elastic Moduli with Temperature for Various Steels and Pure Metals.** G. T. Harris and M. T. Watkins (*Iron Steel Inst. Special Rep.*, 1952, (43), 185–188; discussion, 355–367).—Young's modulus was determined for 12 gas-turbine steels and for pure Fe, Al, and Ni at room temp. and from 100° to 800° C. by steps of 100° C. Static load/deflection and dynamic-vibration-frequency detn. methods were employed, and showed general agreement to 1%. The shear modulus was determined up to 800° C. for two steels and the value of Poisson's ratio deduced from the two elastic moduli. 16 ref.—D. M. P.

Fatigue at High Temperatures. H. J. Tapsell (*Iron Steel Inst. Special Rep.*, 1952, (43), 169–174; discussion, 355–367).—The behaviour of metals under fatigue conditions and when static stresses are superimposed is discussed and the effects of testing speed are noted. Creep and fatigue data for 6 turbine alloys at various elevated temp. are given, and the construction of diagrams showing the combined effects of creep and fatigue conditions is described. The diagrams show the combined alternating and mean stresses producing failure in a given time or those producing, e.g., 0·1% creep. Tests on a few materials showed the loss of fatigue strength in exhaust gases and when small fillet radii were present, and that welding could be carried out without loss of fatigue strength.—D. M. P.

Torsional Fatigue Failures. J. O. Almen (*Product Eng.*, 1951, **22**, (9), 167–182).—The stresses and fracture characteristics of metal parts under torsional loading are evaluated, and torsional-stress diagrams are developed. It is shown that torsional fatigue failure results from tensile stress.—M. A. H.

Fatigue Loss and Gain by Electroplating. J. O. Almen (*Product Eng.*, 1951, **22**, (6), 109–116).—A. gives data on the effects of various electrodeposits on fatigue strength of the basis metal, which show the deleterious effect of tensile-stressed deposits and the beneficial effect of compressively stressed deposits. 14 ref.—M. A. H.

***On the Wavy Deformation of Metal Surfaces in Contact.** Minoru Tanaka (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (3), 177–181).—[In Japanese, with tables and diagrams in English]. Cylindrical test-pieces, 10 mm. in dia. and 10 mm. long, an Amsler 10-ton universal testing machine, and a needle-trace surface tester were used to obtain measurements of wavy deformation due to compression of the contact plane surfaces between metals of the same and different kinds, e.g. electrolytic Cu, 60 : 40 brass, mild steel, axle steel, and cast Fe. With specimens of the same metal, the contact surfaces become rougher with increasing compressive stress. With specimens of different metals the contact surfaces under compressive stress show greater wavy deformation than that produced by plastic deformation of the harder metals, and the roughness is nearly proportional to the amount of deformation of the harder metals. When grey cast Fe is in contact with other metals the surfaces of the latter deform and show imprints of the distribution of the flaky graphite in the cast Fe.—AUTHOR.

The Zener Effect in Metals. I. Barducci (*Alluminio*, 1950, **19**, (4), 324–333).—The theory of the thermoelastic internal friction in metals is set out in a general article. An indirect method was used for measuring internal friction. 15 ref. —I. S. M.

Formulæ for Calculating the Surface Tension of Metals. L. L. Kunin (*Doklady Akad. Nauk S.S.S.R.*, 1951, **79**, (1), 93–96).—

[In Russian]. Some Russian authors consider that the surface tension is determined by the excess potential energy of the surface atoms; others, by the kinetic energy of the electrons. K. quotes 7 formulæ proposed by these workers and shows that for various metals the calculated values differ from the experimental results, although the equation derived by Samoylovich (*ibid.*, 1945, **46**, 403; *Zhur. Eksper. Teoret. Fiziki*, 1946, **16**, 135; *Zhur. Fiz. Khim.*, 1947, **21**, 161; *M.A.*, **14**, 43, 310) agrees well for monovalent metals. K. transforms this into $\sigma = 0{\cdot}01805\left[21{\cdot}517\left(\frac{1}{K_{em}}\right)^{7/6} - \left(\frac{1}{K_{em}}\right)^{3/2}\right]$, where K_{em} is the Hall coeff. This equation gives values agreeing with experimental data for Ag, Cu, Au, Pt, Li, and Na, but not for Al. K. also gives an empirical formula, $\sigma = 444{\cdot}5\,\frac{\psi}{R^2} - 110$ erg/cm.², in which ψ represents the min. work for the removal of an electron in eV., and R the radius of the atom in Å. Calculated values agree well with experimental, except in the case of Sn and Hg.—G. V. E. T.

On the Theory of the Surface Tension of Metals. I. M. Spitkovsky (*Zhur. Fiz. Khim.*, 1950, **24**, (9), 1090–1093).—[In Russian]. S. extends Glauberman's calculations (*ibid.*, 1949, **23**, 115; *M.A.*, **19**, 634) to the h.c.p. crystal structure. Assuming that the electron gas of the metal does not penetrate within the ions, he obtains for the surface tension σ the relation: $\sigma = \frac{0{\cdot}0021\,e^2}{a^3[1 - (8\pi/3\sqrt{2})(R_i/a)^3]^2}$, where e is the charge on an ion, a the interatomic distance, and R_i the ionic radius. Taking an ionic charge corresponding to the mean chem. valency of the element, he calculates values of 1100 dynes/cm. for Zn and 880 dynes/cm. for Cd, compared with experimental values of 743 and 630 dynes/cm., resp. The following values, all in dynes/cm., are similarly calculated: Mg, 204; Ca, 130; Co, 1530; α-Ni, 1560; Os, 1190.—G. B. H.

Observations on the Thermal Expansion and on Poisson's Ratio of Metals with a Cubic Crystal Structure. M. J. Druyvesteyn (*Alluminio*, 1948, **17**, (7/8), 348–350).—D. has compared the values of Grüneisen's const. γ (which $\propto$ the coeff. of thermal expansion) with Poisson's ratio μ, in metals with f.c.c. and b.c.c. lattices. He found that with the former γ is $> 1{\cdot}8$ and with the latter $< 1{\cdot}8$, and that μ is, resp., $>$ and $< 0{\cdot}31$. For γ there was only one exception to the rule (Ca, belonging to the 1st group), whereas for μ there were several (Nb and the alkali metals in the 1st group and Ir in the 2nd). Such different behaviour is explained qual.—F. E. M.

***The Condensation of Steam on a Metal Surface.** H. Hampson (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 58–61).—The duration for which the ideal dropwise condensation of steam upon an impure Cu surface persists is found to depend upon: (1) the surface finish of the metal; (2) the promoter and its method of appn.; (3) the initial cleanliness of the metal, absence of moisture on it before applying the promoter, and the cleanliness of the apparatus; (4) the heat load or rate of condensation; (5) the pressure and temp. of the steam; (6) the temp. of the surface; (7) the presence of a non-condensable gas; and (8) the length and shape of the condensing surface. A mirror-like polish increased the life of a brass surface promoted with oleic acid to 4 times that for finishes with No. 3–0000 grades of emery paper, but similar polish of a stainless steel surface increased the dropwise condensation life to only 1·5 times that of the other finishes. Fatty acid promoters were effective on all surfaces; benzyl mercaptan doubled the life of a brass surface as compared with oleic acid. The injection of very small proportions of oleic acid into the steam maintained the dropwise condition, but did not completely re-establish it once it had broken down. Excess of promoter on the surface, resulting from injection into the steam, reduced the heat-transfer rate considerably. In the field of filmwise condensation, Nusselt's theory gives a good estimate of the steam-side coeff. with inclined flat and vertical surfaces. With dropwise condensation the steam-side coeff. of heat transfer is substantially const. over a range of heat loads (up to 400×10^3 B.Th.U./ft.²/hr.). A small concentration of non-condensable gas gives a considerable advantage for dropwise over filmwise condensation.—J. S. G. T.

Specific Heat Near the Melting Point. G. M. Bartenev (*Zhur. Fiz. Khim.*, 1950, **24**, (8), 1016–1022).—[In Russian]. Using his previous calculation of the amount of liq. formed by statistical fluctuations in a pure solid substance just below the m.p. (*ibid.*, 1948, **22**, 587), B. calculates approx. the anomaly in sp. heat arising from variation of this quantity with temp. Neglecting the difference in sp. heat between liq. and solid, and taking the first term of a power series for the difference in chem. potential between the two phases, he shows that the anomalous component Cp_{an} of the sp. heat at temp. T is given by $Cp_{an} = (k/\mu).[T_s/(T_s - T)]^2$, where k is Boltzmann's const., T_s is the m.p. of the solid, and μ is the "quantum of fusion", an adjustable parameter. In reviewing experimental data he shows that for yellow P and ice anomalies have been observed, but have not been disentangled from the effects of impurities. His own measurements (*Zhur. Tekhn. Fiziki*, 1947, **17**, 1321) on solid Sn, contg. 0·16% total impurities, including 0·1% Cu, are consistent with this relation, with $\mu = 1400$ atoms; liq. Sn shows anomalies above the m.p. consistent with a "quantum of crystn." equal to 92,000 atoms. Ubbelohde's measurements (*Trans. Faraday Soc.*, 1938, **34**, 292) on pentadecane, hexadecane, and octadecane give for the "quantum of fusion", resp., 2600, 2500, and 2500 atoms, and for the "quantum of crystn.", resp., 5000, 3000, and 6000 atoms. With solid Sn the anomalies were observed as much as 25° C. below the m.p., too low to be explained by known impurities.—G. B. H.

***Thermal Conductance of Metal Surfaces in Contact.** T. N. Cetinkale and Margaret Fishenden (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 271–275).—Measurements were made of the thermal conductance, C, between two parallel metal surfaces, of steel, brass, and Al, ground to various degrees of roughness, with air, spindle oil, or glycerol between them, and pressed together. A theoretical expression for C is derived by the use of Southwell's relaxation theory. Values of ε, ψ, and ζ, "constants" occurring in the theory, were found, by experiment, to be independent of the nature of the metal or fluid, and were const. for a given type of surface roughness. The "contact conductance number" is a function of two dimensionless numbers, the relation between which is given by the contact conductance equation developed in the theory. C for smooth/ground and ground/ground contacts can be estimated, with sufficient accuracy for practical purposes, by using this equation and experimental values of ε, ψ, and ζ.—J. S. G. T.

***†Thermal Conductivities of Metallic Conductors, and Their Estimation.** R. W. Powell (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 290–295).—Means for estimating approx. values of the thermal conductivities (K) of metallic conductors from their resp. elect. resistivities, ρ, are discussed. Electron theory indicates that if K is measured in cal./cm.²/sec. for 1 cm. thickness and 1° C. difference in temp. and ρ in Ω-cm. then the Lorenz ratio $L = k\rho T^{-1} = 0{\cdot}586 \times 10^{-8}$, T being the abs. temp. Deviations from this result are to be expected at low temp., and for the transition metals and Bi. For the principal metals and alloys, experimental values of K and ρ over wide ranges of T are fairly closely related by the equation $K = LT\rho^{-1} + C$, where C is regarded as a measure of the lattice component. Values of L and C for Cu, Ni, Al, Mg, Fe, α-Fe, and γ-Fe are tabulated. Values of K and ρ of molten metals indicate that, just above the m.p., values of L for most metals range between $0{\cdot}6 \times 10^{-8}$ and $0{\cdot}8 \times 10^{-8}$. L decreases with rise of temp. and at ~200° C. above the m.p. may possibly exceed the theoretical value by $\not> 10\%$. Thermal and elect. data for 10 forms of graphite are tabulated. 81 ref.—J. S. G. T.

Coefficients of Thermal Conductivity of Materials Used in the Construction of Heat-Exchangers. Giuseppe delle Canne (*Chim. e Ind.*, 1951, **33**, (1), 11–15).—delle C. briefly examines the factors influencing the coeff. of thermal conductivity with regard to the effect of the type of wall to which the heat-exchanger may be attached. Figures showing the thermal conductivity coeff. at various temp. are then given for some

120 materials which may be used in the construction of heat-exchangers, including Ag, Al, Cu, Mg, Ni, Pb, Sn, Zn, and their alloys.—I. S. M.

***Heat Transfer to Air Passing Through Heated Porous Metals.** P. Grootenhuis, C. A. Mackworth, and O. A. Saunders (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 363–366).—Heat transfer between a porous solid or a bed of packed solids and a gas flowing through it was studied by a steady-heat-flow method in which the heat supply to a porous wall, through which air flows, is maintained by radiation from a high-temp. source heated elect. The porous material used consisted of an atomized Cu–Sn alloy (Porosint) of porosity 0·206–0·395, determined on a weight basis. Previous results for beds of balls, Fe ores, spheres of Pb, steel and glass, granite gravels, and diatomaceous earth are also discussed. All the correlated data lie fairly well along a straight-line graph, of slope 1·0, in which ordinates represent the Nusselt number of the air flow, and abscissæ represent values of $Gd/\mu f$, where G = the mass flow of air/unit frontal area, d = the dia. of the spherical particles, μ = the viscosity of the air, and f = the porosity, except that results obtained in the present work fall a little below this line, probably owing to deformation of the particles and loss of area due to bonding of the particles on sintering.—J. S. G. T.

The Fundamental Physics of Heat Conduction. J. D. Eshelby (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), **1951**, 267–270).—The classical theory of heat conduction by gases, developed by Maxwell and Boltzmann, is very briefly outlined, and then the modern electron theory of heat conduction, more especially by solids, is discussed, attention being directed towards the following matters: (1) the marked difference between metals and non-metals; (2) the temp.-dependence of thermal conductivity K. Approx. K for metals is independent of temp., and varies as $1/T$ for non-metallic crystals; (3) the Wiedemann–Franz law correlating elect. and thermal conductivity in metals; and (4) the effects of impurities, cold working, and alloying upon K.—J. S. G. T.

***Conductivity of Cold-Worked Metals.** D. L. Dexter (*Phys. Rev.*, 1952, [ii], **85**, (5), 936–937).—A letter. The calculations of Landauer (*ibid.*, 1951, [ii], **82**, 520; *M.A.*, **19**, 339) are modified and carried further to give a larger predicted increase of resistivity. The treatment assumes a non-existent singularity of the scattering potential on the dislocation axis, but the fictitious contribution is only about half the total.—P. C. L. P.

***The Formation of a Boundary Between Normal-Conducting and Supraconducting Metal.** M. P. Garfunkel and B. Serin (*Phys. Rev.*, 1952, [ii], **85**, (5), 834–840).—An experiment is described in which a single crystal of 99·998% pure Sn, at a uniform temp., was positioned in a non-uniform magnetic field so that it was partly supraconducting and partly in the normal state. Considerable hysteresis was shown in the critical magnetic field for the destruction of the supraconductivity of part of the crystal, at temp. between 3·68° and 3·37° K. Analysis of the results in terms of the mechanism of the supraconducting $\rightleftharpoons$ normal transition suggests the existence of a surface energy approx. $\propto$ the critical field.—P. C. L. P.

***(I). Magnetostriction in Weak Magnetic Fields. (II). Some Phenomena in Weak Magnetic Fields.** Shigeo Matsumae (*Sci. Rep. Tôhoku Univ.*, 1950, [i], **34**, (3), 129–139; 1951, [i], **35**, (1), 41–43).—[In English]. [I.—] The longitudinal magnetostriction effect, in weak magnetic fields, corresponding to the initial permeability range, was found to exist in Ni but not in Fe. In Ni, the effect $\propto$ the intensity of magnetization; it depends considerably upon the method of demagnetization used, whether by heating or by an alternating magnetic field, and upon the max. amplitude of the demagnetizing field used, and upon internal stress. [II.—] It is now found that the longitudinal magnetostriction effect in Ni decreases with time after rapid cooling from 900° C. rapidly at first and finally attains a const. value. Young's modulus, (E), of Ni demagnetized by heating is smaller by $\sim$1% than E for Ni demagnetized by an alternating magnetic field of $\sim$7 Oe.—J. S. G. T.

***The Periodic Relation Between Magnetic Susceptibility and Field for Metals at Low Temperatures.** B. I. Verkin, B. G. Lazarev, and N. S. Rudenko (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, (1), 93–94).—[In Russian]. Since the de Haas–van Alphen effect had already been observed in Bi (H. and A., *Proc. K. Akad. Wet. Amsterdam*, 1930, **33**, 680, 1106; *Met. Abs.* (*J. Inst. Metals*), 1931, **47**, 72, 264), Zn (Sydoriak and Robinson, *Phys. Rev.*, 1949, [ii], **75**, 118; *M.A.*, **16**, 469), and Sn (V., L., and R., *Uspekhi Fiz. Nauk*, 1949, **39**, 319), V., L., and R., believing it to be a general metallic property, looked for it in single crystals of Be, Mg, In, and Cd. The method of investigation was to measure the couple acting on a crystal suspended by an elastic filament in a uniform magnetic field, with the highest-order axis of symmetry in the plane of the field. The difference between the magnetic susceptibilities $\parallel$ and $\perp$ this axis was measured by the twist in the suspension. Measurements were made at 78°, 20°, and 14° K. and in liq. He down to 2° K., in fields of 4000–14,000 Oe. In Be the effect appeared at 20·4° K. and became stronger at liq.-He temp. In a field of 5000 Oe. the "period of oscillation" of the difference in susceptibilities was 180 Oe., increasing to 1000 Oe. at 12,000 Oe. In Mg the effect appeared at 4·2° K. and increased at lower temp. In a field of 10,000 Oe. the "period of oscillation" was 140 Oe., increasing to 280 Oe. at 14,000 Oe. In In the effect appeared at $\sim$2° K. in fields $>$13,000 Oe. In Cd the effect appeared at temp. $<$2° K. in strong fields, the "period of oscillation" ranging from $\sim$25 Oe. at 12,000 Oe. to 40 Oe. at 14,000 Oe. Since the effect has now been observed in Ga and graphite (Shoenberg, *Nature*, 1949, **164**, 225; *M.A.*, **17**, 638), as well as in all the metals mentioned above, it appears to be a general metallic property.—G. B. H.

The Periodic Relation Between Magnetic Susceptibility and Field at Low Temperatures. Yu. B. Rumer (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, (6), 573).—[In Russian]. Continuing his previous calculations on the basis of a "free electron" theory (*ibid.*, 1948, **18**, 1081), R. deduces that, for practically attainable magnetic fields, the "period of oscillation" ΔH of the magnetic susceptibility of a metal at low temp. in a field H is given approx. by $\Delta H = 2H^2/H_0$, where H_0 is a const. This is in good agreement with Verkin, Lazarev, and Rudenko's measurements on the de Haas–van Alphen effect in Be, Mg, and Cd (*ibid.*, 1950, **20**, (1), 93; preceding abstract).—G. B. H.

***On Magnetic Hysteresis with Respect to Temperature in Ferromagnetics.—II.** Ya. S. Shur and N. A. Baranova (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, (2), 183–191).—[In Russian]. Specimens of Fe–Si alloy (3·7% Si) ribbon, 100 × 2 × 0·35 mm., and of Ni wire 2 mm. in dia., 100–120 mm. long, were heated and cooled between −195° C. and their resp. Curie points in a weak const. magnetic field, while their intensities of magnetization were measured. Before measurement the specimens were annealed *in vacuo* and demagnetized with a 50-c./s. alternating field, slowly decreased from 120 Oe. to zero. For Fe–Si alloy the following effects were observed: Max. occur on the heating curves at 370°–380° C. and at 580°–600° C., in addn. to the Hopkinson max. just below the Curie point. The first max. disappears in fields $>$0·21 Oe. and the second in fields $>$0·12 Oe. On the cooling curves a max. occurs at 430°–450° C., between the two max. on heating, for fields $<$0·9 Oe. The cooling curves show higher intensities of magnetization than the heating curves, but the difference $(I_1 - I_0)/I_0$, where I_0 and I_1 are, resp., the values at −195° C. before and after one cycle of heating and cooling, decreases with increasing field, reaching zero at 30 Oe. On taking the specimen through thermal cycles to different temp., it was found that for max. temp. of 350°–500° C. the cooling curve was initially below the heating curve, but crossed it at a lower temp. to form a "loop". This effect disappeared in fields $>$0·01 Oe. Thermal cycles taken in fields of 0·03 and 0·06 Oe. after measurements in fields up to 30 Oe. showed max. at different temp. from those found at first, with cooling curves forming "loops".

These changes were not removed by heating to 20° C. above the Curie point, but disappeared on annealing for 1 hr. at 1050° C. For Ni, a max. at ~100° C. appears on the heating curve in fields of 0·03–0·045 Oe., in addn. to the Hopkinson max. A max. occurs on the cooling curve at ~200° C. in fields $<1{\cdot}5$ Oe. The hysteresis $(I_1 - I_0)/I_0$ is greater than that of Fe–Si alloy, and decreases with increasing field, reaching zero at 24 Oe. Sh. and B. attribute these effects to the surface energy of domain boundaries, which depends on crystal anisotropy, magnetostriction, and internal stresses. These factors become more important relative to the field energy in weak fields. The observed max. are attributed to the change in sign of the anisotropy const. for Ni at ~130° C. (Kirensky, *Doklady Akad. Nauk S.S.S.R.*, 1949, **64**, 53), the change in sign of the 2nd anisotropy const. for Fe–Si alloy at ~300° C. (Shubina, *Izvest. Akad. Nauk S.S.S.R.*, 1947, [Fiz.], **11**, 527) and the max. value of the magnetostriction in Fe–Si alloy at ~500° C. (Shturkin, *ibid.*, 1947, [Fiz.], **11**, 661).—G. B. H.

***Theory of Magnetic Diffusive Drag.** Louis Néel (*J. Phys. Radium*, 1952, **13**, (5), 249–264).—A theory of magnetic diffusive, or reversible magnetic, drag, according to which foreign interstitial atoms, e.g. C or N, that are the origin of the phenomenon, operate, by magneto-cryst. coupling, between their position and the direction of spontaneous magnetization, is discussed. It is shown that the phenomenon necessitates that energy of stabilization, which can amount to as much as 100 ergs/cm.^{3}, must be added to the terms expressing the ordinary magnetic anisotropy, and a general expression for the pressure at a 90° or 180° boundary, owing to diffusion of foreign atoms, is derived. An approx. method of calculating the effects of diffusion, which are considered as analogous to a small perturbation of the ordinary magnetization process, is developed. Appn. of the theory to magnetic phenomena are discussed. 16 ref.—J. S. G. T.

Quantum Theory of Magnetic Anisotropy. S. V. Tyablikov (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, (7), 661–668).—[In Russian]. T. treats the anisotropy of magnetic properties in magnetically uniaxial crystals as a consequence of directional anisotropy in the interaction of electrons in the unfilled shells. On the basis of the Heitler–London model, the magnetization of a single crystal ∥ and ⊥ the principal axis is calculated, by means of Bogolyubov and T.'s perturbation method (*ibid.*, 1949, 19, 256; *M.A.*, 19, 706). The results, which involve one empirically determined const., reproduce qual. the variation with temp. and magnetic field of the magnetization ∥ and ⊥ the principal axis, e.g. in Co.—G. B. H.

Phase Transitions of the Second Order and Critical Phenomena. V. K. Semenchenko (*Zhur. Fiz. Khim.*, 1947, **21**, 1461–1469; *C. Abs.*, 1948, **42**, 5321).—[In Russian]. An analogy is pointed out between phase transitions of the 2nd order, at which the heat capacity and the coeff. of thermal expansion change abruptly, and critical phenomena, such as the Curie point in ferromagnetism, λ point in liq. He, &c. In two-component systems the transition between solid soln. and randomly distributed mol. or atoms is the corresponding effect. Increase in number and strength of intermolecular bonds is the characteristic feature of all these transitions.

Phase Transitions of the Second Order. G. M. Bartenev (*Zhur. Fiz. Khim.*, 1949, **23**, (11), 1357–1360).—[In Russian]. B. criticizes Epstein's treatment of this subject ("A Textbook of Thermodynamics," New York: **1937**, pp. 128–133), showing that Ehrenfest's relation, $\left(\frac{\partial \Delta V}{\partial T}\right)_p^2 + \frac{\Delta C_p}{T}\left(\frac{\partial \Delta V}{\partial p}\right)_T = 0$, which Epstein did not accept, can be deduced from the latter's assumptions. This simplifies the equations, so that the following relation is deduced: $\frac{dp}{dT} = \frac{\Delta C_p}{T\left(\frac{\partial \Delta V}{\partial T}\right)_p}$, where T is the temp. of the 2nd-order phase change under pressure p, ΔC_p the associated discontinuity in sp. heat, and ΔV the difference in sp. vol. between the two phases at temp. T. Hence there is only one possible curve of transition temp. against pressure, and not two, as Epstein's treatment implies. Landau's treatment (L. and Lifshits, "Statisticheskaya Fizika," Moscow: **1940**, pp. 191–206) which was originally expounded for order–disorder phenomena in crystals but is applicable also to 2nd-order phase changes in general (Semenchenko, *Zhur. Fiz. Khim.*, 1947, **21**, 1461; preceding abstract), shows that a 1st-order phase change in a one-component system must pass over to a 2nd-order transition without change in the slope of the (p, T) curve at the critical temp. S. pointed out that the "liq.–vapour" transition above the critical temp. is a 2nd-order phase change. B. considers that the same phenomenon may in principle occur for solid–liq. transitions above a critical temp., but not for changes of cryst. structure in a solid.—G. B. H.

Status of the Values of the Fundamental Constants for Physical Chemistry as of July 1, 1951. Frederick D. Rossini, Frank T. Gucker, Jr., Herrick L. Johnston, Linus Pauling, and George W. Vinal (*J. Amer. Chem. Soc.*, 1952, **74**, (11), 2699–2701).—J. R.

2—PROPERTIES OF ALLOYS

***The Heat Resistance of Light Alloys.** Jacques Valeur (*Rev. Aluminium*, 1952, (192), 339–346).—See *M.A.*, 20, 239.

***The Effect of Notches on the Strength of Aluminium Alloys Under Static Tensile Loading.** J. H. Palm (*Metalen*, 1952, **7**, (18), 309–316).—[In English]. An investigation was carried out into the U.T.S. of 24S-T and 51S-W alloy notched cylindrical test-bars under static tensile load. An opt. arrangement for observing and measuring changes in the notched section under load is described and illustrated. By varying the length and total size of the test-bar, and the notch angle and the notch depth at const. dia. of the notched section, the following results were obtained: (1) U.T.S. tends somewhat to decrease with increasing length and increasing abs. size of the test-bar; (2) after decreasing with decreasing notch angle (180°–30°), the U.T.S. tends to increase again, reaching original values in some cases; (3) with increasing notch depth the U.T.S. decreases sharply but tends to increase again at the end of the curve; and (4) on prior compression the U.T.S. decreases proportionately. 12 ref. —I. S. M.

***High-Stress-Fatigue of Aluminium and Magnesium Alloys.** T. T. Oberg and W. J. Trapp (*Product Eng.*, 1951, **22**, (2), 159, 161–163).—Stress-cycle curves for Al alloys 75S-T6 extrusion (unnotched) and 14S-T extrusion (notched and unnotched) under reversed bending, are given; similar curves are also given for Mg alloys ZK 60 and A 280 extrusions (both notched and unnotched) under reversed bending and under axial loading.—M. A. H.

***Relief of Residual Stress by a Single Fatigue Cycle [in Light Alloys].** W. P. Wallace and J. P. Frankel (*Weld. J.*, 1949, **28**, (11), 565S).—Effects of the appn. of cyclic stress on residual notch stresses have been studied by Norton, Rosenthal, and Maloof (*ibid.*, 1946, **25**, 269S), R., Sines, and Zizicas (*ibid.*, 1949, **28**, 98S; *M.A.*, **17**, 640), and Hall and Parker (*ibid.*, 1948, **27**, 421S). Their results are compared with those of cyclic loading tests on a light-alloy specimen, in which residual stresses at the base of the notch were relieved during the first cycle of fatigue stress; this contradicts findings of H. and P., but confirms other results.—P. R.

***Equilibrium Diagram of the System Aluminium–Indium.** S. A. Pogodin and I. S. Shumova (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, **17**, 200–203; *C. Abs.*, 1951, **45**, 2384).—[In Russian]. Al and In do not combine to form chem. compounds. In the liq. state they are mutually soluble only to a limited extent, at 640° C. the sepn. into layers extending from 12 to 97% In. There are practically no solid soln. in the system. The system forms a eutectic which crystallizes at 152° C.

***The Effect of Cold Work on the Microstructure and Corrosion-Resistance of Aluminium–5% Magnesium Alloys Containing 0–1% Zinc.** (Brenner and Metcalfe). See col. 416.

***Influence of Vanadium on the Characteristics of Aluminium for Electrical Purposes.** C. Panseri and M. Monticelli (*Alluminio*, 1948, **17**, (7/8), 335–338).—Nine samples of Al for elect. conductors of controlled compn. and with impurities kept as const. as possible (Fe : Si ratio ~1 : 1·5) were prepared with a V content varying from 0 to 0·1%. With a "K" Leeds-Northrup potentiometer it was found that up to a content of 0·06% V the elect. resistance ∝ the % of V present, and within these limits each 0·001% V increased the resistance by 0·0000426 Ω-mm.^{2}/m. The mech. characteristics were practically unaffected. Above 0·06% the effect of increasing V contents became smaller and was no longer directly proportional. This was probably due to the equilibrium condition between the solid soln. and the compound VAl_3.—F. E. M.

Ternäre Legierungen des Aluminiums. (Hanemann and Schrader). See col. 447.

***Methods of Preparing Arsenides of Precise Composition.** M. I. Kochnev (*Zhur. Priklad. Khim.*, 1950, **23**, (6), 561–565).—[In Russian]. By heating stoichiometric mixtures of metal powders and metallic As (obtained by reduction of As_2O_3) in a crucible at the bottom of a vertical tube filled with H, the following arsenides were prepared: Co_5As_2, Co_2As, Co_3As_2, CoAs, Ni_3As_2, Ni_2As, Ni_5As_2, Cu_3As, Cu_5As_2, and Fe_2As. The tube was heated to a temp. near the m.p. of the desired arsenide. Co_2As_3 and $CoAs_2$ were prepared by placing suitable mixtures in a crucible inside a steel bomb (lined with a mixture of Fe_2O_3, Al_2O_3, Fe powder, and water glass) and heating to 950° C. The joint of the bomb was attacked by As vapour.—G. V. E. T.

***Solubility of Nitrogen in Liquid Chromium and Melts of Chromium and Silicon.** (Mozgovoi and Samarin). See col. 385.

***[Preparation and Properties of] Chromium-Base Alloys for Gas-Turbine Applications.** E. A. G. Liddiard and A. H. Sully (*Iron Steel Inst. Special Rep.*, 1952, (43), 243–245; discussion, 345–354).—The difficulties encountered in the purification, melting, and casting of Cr- and Co-rich alloys are discussed. The exclusion of O, N, and C is important, and the most promising method appears to be melting *in vacuo* and casting under A. The creep-resistance in compression of alloys in the solid-soln. ranges of the systems Cr + Fe, Co, Ni, Pd, or Al has been investigated and found to be greatest in Cr–Fe alloys. The oxidation-resistance of the alloys in the range 900°–1000° C. is good, but brittleness at room temp. limits their appn. 7 ref.—D. M. P.

***Creep Properties of Some Cobalt-Rich Alloys for High-Temperature Service.** J. C. Chaston and F. C. Child (*Iron Steel Inst. Special Rep.*, 1952, (43), 246–248; discussion, 345–354).—The creep properties of cast Co–Ta–Cr alloys at 900° C. have been investigated and are shown to improve with the addn. of Ta (up to 15%) to an alloy contg. 10% Cr. The addn. of C up to 0·3% to an 80 : 10 : 10 Co–Ta–Cr alloy produced a further increase in creep-resistance and resulted in a life-to-rupture of 1500 hr. under 4·5 tons/in.^{2} at 900° C. This alloy has been successfully precision-cast. Greater creep-resistance may be obtained by further increases in Ta and Cr contents.—D. M. P.

***On the Rigidity Modulus and Its Temperature Coefficient of Cobalt–Iron–Chromium Alloys.** Hakaru Masumoto and Hideo Saito (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (3), 125–128).—[In Japanese, with tables and diagrams in English]. The mean temp. coeff. of rigidity modulus, *g*, in the temp. range 20°–50° C. of Co–Fe–Cr alloys contg. Co 50–90 and Cr <20% have been measured by the torsion pendulum method, which is described. For most alloys *g* is negative. As the compn. approaches that of stainless Invar (Co 54, Fe 36·5, Cr 9·5%), *g* diminishes, first gradually and then rapidly, becomes zero, and changes its sign. It then reaches a positive max. ($+ 35{\cdot}9 \times 10^{-5}$) at the compn. of stainless Invar. The various alloys having a zero value of *g* are called Co-Elinvar. The results reported here are similar to those of the temp. coeff. of Young's modulus reported previously by M. and S. (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1949, [A], **1**, 17; *M.A.*, **18**, 165), except that *g* shows a range of positive values and a max. positive value in the higher-Co and lower-Cr alloys. The rigidity modulus at 20° C. has been determined also, and results show a max. value of $9{\cdot}34 \times 10^5$ kg./cm.^{2} for an alloy contg. Co 50 and Cr 5% and a min. value of $5{\cdot}92 \times 10^5$ kg./cm.^{2} for an alloy contg. Co 65 and Cr 8%.—AUTHORS.

***On the Quenching and Tempering of β-Tin Bronze.** Yoshikazu Hosoi (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 42–46).—[In Japanese, with tables and diagrams in English]. H. has made more detailed experiments than Matsuda (*J. Inst. Metals*, 1928, **39**, 67) and Imai and Obinata (*Mem. Ryojun Coll. Eng.*, 1930, **3**, (2C), 117) by means of elect. resistance, hardness, and microscopic analysis. The acicular structure appeared in alloys contg. <25% Sn when quenched from 590° C. or above. The decompn. of β or β′ can be arrested at room temp. in alloys belonging to the eutectoid compn. and the elect. resistance and hardness reaches a const. value in these alloys. The elect. resistance of alloys having an acicular structure falls between 200° and 320° C., decreasing to a min. at 320° C. The arrest point appeared at 250° C. In accordance with this the temper hardness increased between 200° and 320° C. in alloys showing an acicular structure and decreased at 250° C. in alloys not having this structure. The max. hardness of both types of alloys was reached at 320° C., and they both decomposed at 400° C. into the (α + δ) phases.
—AUTHOR.

***The Effect of Cold Drawing and Annealing on the Modulus of Rigidity [and Electric Resistance] in Metals [α-Brass and Mild Steel Wires].** Hiroshi Yamamoto and Hideo Takei (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (3), 128–132).—[In Japanese, with graphs in English]. Torsion pendulum and potentiometer methods have been used. It was found that: (1) the modulus of rigidity, *G*, of α-brass wire gradually decreases with increase of cold drawing up to 30% reduction of area, after which it gradually increases; (2) *G*/annealing temp. curves show a small peak at 100°–200° C. and a larger one at 300°–450° C.; (3) the latter phenomenon coincides well with the change of elect. resistance, *R*, in heavily cold-drawn α-brass wire; and (4) in the α-brass wire of lower degree of cold drawing (e.g. 12, 22, and 33% reduction of area), *G* changes considerably with the annealing temp., whereas changes in *R* are small.
—AUTHORS.

***New Values of the Coefficients of Equivalence for Manganese, Iron, Cobalt, and Nickel in Copper–Zinc Alloys.** J. B. Haworth (*J. Inst. Metals*, 1952–53, **81**, (5), 254).—A note. Recent experimental work, described elsewhere (*Phil. Mag.*, 1952, [vii], **43**, 613), has shown that the values ascribed by Guillet to the "coefficients of equivalence" of Mn, Fe, and Ni are in error, and has enabled a hitherto undetermined coeff. to be stated for Co.—AUTHOR.

***Equilibrium Diagram of the System Indium–Antimony.** S. A. Pogodin and S. A. Dubinsky (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, **17**, 204–208; *C. Abs.*, 1951, **45**, 2384).—[In Russian]. In and Sb are completely soluble in the liq. phase. Upon cooling they crystallize into 3 solid phases: In, Sb, and InSb. The last melts at 536° C. and forms with its components 2 eutectics, of which one is located near pure In (m.p. 155° C.), and the other contains. 70·4 at.-% Sb (m.p. 505° C.).

***Phase Diagrams of the Ternary Systems Fe–Cr–W and Fe–Cr–Mo at Low Temperatures.** H. J. Goldschmidt (*Iron Steel Inst. Special Rep.*, 1952, (43), 249–257; discussion, 345–354).—The alloys were prepared by pressing and sintering powder compacts, and the 600° C. isothermal sections of the two systems were investigated by X-rays. In the Fe–Cr–Mo system, the Fe–Cr σ phase occurs over a wide range and inhibits the occurrence of the Fe–Mo ξ(Fe_3Mo_2) phase and a ternary compound around Fe_4CrMo_2 appears. In the Fe–Cr–W system the Fe–W ξ(Fe_3W_2) phase predominates over the α phase and no ternary compound was found. The potential appn. of the alloys are discussed, and the areas of the diagrams pointed out where pptn.-hardening effects may result in useful heat-resisting materials. A theoretical discussion of the systems in terms of size and electronic factors and a note on the σ-structure conclude the paper. 27 ref.—D. M. P.

***Some Proven Gas-Turbine Steels and Related Developments.** D. A. Oliver and G. T. Harris (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 46–59; discussion, 306–330).—An account is given of the development and use of special steels for jet-engines. Results of creep tests up to 30,000 hr. on austenitic steels are presented. The uses and superior creep strength of ferritic steels are discussed. The results of scaling tests in gas atmospheres with and without V_2O_5 are given. Factors governing progress and development trends are noted. An appendix contains extensive data on compn. and high- and low-temp. properties of 6 gas-turbine steels. 21 ref.—D. M. P.

Stresses in Gas-Turbine Discs and Rotors. R. W. Bailey (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 30–35; discussion, 368–378).—A discussion of the centrifugal and thermal stresses which arise in turbine discs. The resistance of ferritic and austenitic materials is compared by hydraulic bursting tests on cylinders. Strain-hardening improves the resistance to bursting of an austenitic alloy. 4 ref.—D. M. P.

Influence of Operating Temperature on the Design and Performance of Gas Turbines. S. L. Bragg (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 23–29; discussion, 368–378).—A discussion of the design of gas-turbines; the properties required of the materials used for blades and discs are noted. —D. M. P.

***The Effect of Minor Additions on the Age-Hardening Properties of a High-Purity Lead-Antimony Alloy.** L. M. T. Hopkin and C. J. Thwaites (*J. Inst. Metals*, 1952–53, **81**, (5), 255–260).—It was found that little or no age-hardening occurred, after suitable treatments, in a Pb-0·85% Sb alloy of extreme purity, although the alloy of commercial purity is known to age-harden extensively. Minor addn. of various elements whose solid solubility in Pb varied with temp. caused marked age-hardening of the alloy and increased the dispersion of the precipitated Sb. As little as 0·001% As promoted marked hardening. Addn. elements either completely or negligibly soluble at all temp. had no effect. A possible mechanism is suggested which is similar to that offered by Hardy (*ibid.*, 1950–51, **78**, 169; *M.A.*, **18**, 163) to explain the effect of minor addn. on the age-hardening characteristics of an Al–4% Cu-0·15% Ti alloy.—AUTHORS.

***Creep at 250° and 300° C. of Some Magnesium Alloys Containing Cerium.** G. A. Mellor and R. W. Ridley (*J. Inst. Metals*, 1952–53, **81**, (5), 245–254).—Following on an investigation (*ibid.*, 1948–49, **75**, 679; *M.A.*, **17**, 807) of the creep strength of Mg–Ce alloys at 200° C., tests have now been made at 250°, 300°, and 316° C., with and without addn. of a third element. At 250° C. slight benefit results from addn. of Si and Zr; Ag, Li, Zn, and Co have little influence on creep behaviour; and Cd and Al are harmful. The best results at 300° and 316° C. were obtained with as-cast alloys contg. ~2% Ce or rare earths, and 1% Mn. Both rolled and cast alloys of low Mn content can be improved by heat-treatment, but cast Mg–Ce–Mn alloys remain the most creep-resistant. Examination with the opt. and electron microscopes showed improvement in creep-resistance to be due mainly to the presence of fine precipitates, Mn apparently restraining the coarsening of the Mg_9Ce particles. —AUTHORS.

Magnesium–Rare-Earth Alloys. L. W. Eastwood (*Product Eng.*, 1951, **22**, (9), 158–161).—The phys. properties of two experimental Mg alloys are described. A casting alloy contg. Ce 6, Mn 0·9, Ni 0·2, and W 0·01% shows improved foundry characteristics. A wrought alloy contains Ce 2, Mn 1·5, and Ni 0·2%. Both alloys show good load-carrying capacity at 600° F. (315° C.). Appn. are indicated. —M. A. H.

New Elektron Magnesium–Zirconium–Zinc–Thorium Casting Alloy [ZT1]. —— (*Metallurgia*, 1952, **46**, (275), 145–146).—The properties of ZT1 alloy contg. nominally Th 3, Zn 2·5, and Zr 0·7% are briefly outlined. The alloy has a better resistance to creep at elevated temp. than earlier Mg alloys, e.g. MCZ and ZRE1.—F. M. L.

***Some Preliminary Comments on the System Molybdenum–Aluminium.** R. L. Wachtell (*Powder Met. Bull.*, 1952, **6**, (3), 99–104).—The Mo–Al system is briefly described, and several microstructures of Al–Mo alloys are reproduced. 6 ref.—W. A. M. P.

***[Creep Properties of] Special Steels [Nickel–Chromium–Iron Alloys] for Gas Turbines.** W. E. Bardgett and G. R. Bolsover (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 135–148; discussion, 306–330).—Creep tests were carried out on a steel contg. C 0·25, Mn 1·63, Si 1·03, Ni 46·52, Cr 20·5, Mo 2·73, Co 3·33, Nb 2·92, W 3·52, and Ti 1·2%, and with one of the last 5 elements omitted. It was shown that the elements which contributed to creep-resistance in order of effect at 650° C. were Ti, W, and Nb and at 800° C. Nb, Ti, and Co. Mech. tests at room and elevated temp. and creep tests at 650° C. on the above alloy showed that direction and amount of hot work had little effect on the essential properties. Details are given of the mech. and creep properties of 2 steels (contg. Cr 25, Ni 15% and Cr 20, Ni 30, Ti 1–2%) suitable for combustion chambers. The latter alloy showed good creep-resistance at 650° C., and the available data on dimensional stability and consequent " negative creep " are presented. —D. M. P.

***Development of a High-Temperature [Nickel–Chromium–Cobalt–Iron] Alloy for Gas-Turbine Rotor Blades.** G. T. Harris and H. C. Child (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 67–80; discussion, 306–330).—H. and C. report an investigation into the effects on the creep properties of: (1) the carbide-forming elements W, Mo, Nb, and V on a 20 : 20 : 20 Ni–Cr–Co austenitic alloy contg. 0·5% C, singly and in threes in such a way as to build up an approx. " quaternary diagram " of creep-resistance; (2) the C content on the same base +3% each of V, Mo, and Nb; and (3) the base compn. on alloys contg. 3% each of V, Mo, and Nb + 0·5% C. A further study based on the results of the above investigations resulted in the development of alloys of the Jessop G.32 type (C 0·3, Mn 0·8, Si 0·3, Ni 12, Co 45, Cr 19, V 2·8, Mo 2, Nb 1·2, Fe ~16%). The heat-treatment of the alloys was examined and the optimum found to be soaking for 10 min. at 1280° C. and oil-quenching followed by ageing for 46 hr. at 750° C. The high-temp. creep properties of G.32 are tabulated. The nature of the carbides formed in these alloys was examined by chem. and X-ray analysis after extraction from the alloy. 5 ref.—D. M. P.

***Nickel–Chromium–Titanium Alloys of the Nimonic 80 Type.** L. B. Pfeil, N. P. Allen, and C. G. Conway (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 37–45; discussion, 306–330).—The potential suitability of Ni–Cr-base alloys for use in gas turbines was indicated by the survey of the information available in 1939 which is given in this paper. A preliminary survey of the effect of addn. elements on mech. properties showed the most promising group to be the 80 : 20 Ni–Cr alloy with Ti and Al (the latter being present in the Ti master alloys). A more detailed study of the solid solubility of (Ti and Al) in 80 : 20 Ni–Cr alloy, its effect on the response to heat-treatment and on the forgeability at the predetermined optimum temp. of ~1130° C., and of the effects of heat-treatment on the creep properties at 650° and 800° C. led to the development of Nimonic 80 (Cr 18–21, Ti 2–2½, Al ~0·5%, balance Ni). The microstructure, scaling-resistance, and creep properties of Nimonic 80 and 80A (in which the Al content is increased to ~1·25% resulting in better high-temp. properties) are then discussed in more detail, and creep curves are given for Nimonic 80A at 650°, 700°, and 750° C. 24 ref.—D. M. P.

***Study in Static and Alternating Fields of a Ferro-Nickel with a Very Small Potential.** Israël Epelboin and Guy Gilarding (*Compt. rend.*, 1952, **234**, (19), 1860–1862).—Two of the methods of studying magnetization in static and alternating fields of a specimen in a closed magnetic field previously described are used to compare ribbons of high-permeability Mumetal and to explain the origin of the often very pronounced magnetic structure.—J. H. W.

***The Origin of High Permeability in Permalloy.** Sōshin Chikazumi (*Phys. Rev.*, 1952, [ii], **85**, (5), 918–919).—A letter. Although the directional character of the high permeability developed by slowly cooling alloys of compn. Ni_3Fe in a magnetic field can be semi-quant. attributed to the strains of directional ordering, the origin of the high initial permeability on quenching from 600° C. remains doubtful. Observations have been made of the changes in the domain pattern observed on the (110) face of an Ni_3Fe crystal on subsequent disordering, and a very fine domain pattern was found to develop. It is possible that this structure gives a high permeability.—P. C. L. P.

***Investigation of the System Palladium–Gold–Copper.** V. A. Nemilov, A. A. Rudnitsky, and R. S. Polyakova (*Izvest. Sekt. Platiny*, 1949, (**24**), 35–51; *C. Abs.*, 1951, **45**, 522).—[In Russian]. The alloys studied were annealed *in vacuo* at 900°–1000° C., depending on the m.p. of the alloy, for 10 days and were then maintained for 5 days at 600° C., 3 days at 500° C., and 2 days at 300–400° C., before being finally cooled in the furnace. The transformation/temp. curves for sections at 70, 75, and 80 at.-% Cu had flat max. The transformation in these sections consisted of chem. reactions of Cu_3Au and Cu_5Pd, forming solid soln. Both of these compounds have the same cryst. structure. The transformation curve for 60 at.-% Cu rose sharply upon addn. of Pd to Au–Cu alloy and then proceeded almost horizontally toward the Pd–Cu side. The transformations in this section led to the formation of CuAu and Pd_3Cu_5. These compounds, having different crystal structures, were not likely to form solid soln. At 50 at.-% Cu, the curve consisted of 2 parts which joined at 30 at.-% Pd. One part of the curve is due to the formation of CuAu and the other to Cu_5Pd_3, which combine to form ternary solid soln. At 40 at.-% Cu a transformation was observed only on the Au–Cu side, caused by the formation of CuAu. At 30, 20, and 10 at.-% Cu no transformation in the solid state was observed. The hardness curves of hardened specimens had the smooth form characteristic of solid soln. The hardness curves of annealed specimens indicated the presence of Cu_3Au, CuAu, Cu_5Pd, and Cu_5Pd_3. The elect. resistance curves obtained from measurements made at 25° and 100° C. on hardened and annealed specimens were analogous to the hardness curves. No ternary daltonides were observed.

***Alloys of Palladium and Tungsten.** V. A. Nemilov, A. A. Rudnitsky, and R. S. Polyakova (*Izvest. Sekt. Platiny*, 1949, (**23**), 101–103; *C. Abs.*, 1951, **45**, 1939).—[In Russian]. Pd–W alloys with up to 22·6 wt.-% (14·5 at.-%) W were investigated. All formed solid soln. The B.H.N. increased from 31·5 kg./mm.^{2} for Pd to 70·5 kg./mm.^{2} at 6·58 wt.-% W and to 88·5 kg./mm.^{2} at 22·60 wt.-% W. The sp. resistance at 25° and 100° C. increased with the W content. The thermo-e.m.f., negative for Pd, was approx. zero at 2·06 wt.-% W, became positive with increasing W content, and reached 23 mV. at 1000° C. at 12·21 wt.-% W.

***X-Ray Investigation of Alloys of the System Platinum–Copper–Nickel.** V. G. Kuznetsov (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1948, **16**, (4), 150–167; *C. Abs.*, 1951, **45**, 522).—[In Russian]. Ternary Pt–Cu–Ni alloys with 50, 60, 80, and 90 at.-% Pt and binary Pt–Cu alloys with 24·97, 50·22, 60·73, and 73·99 at.-% Pt, were studied. Ternary alloys heated at 950° C. or higher formed a continuous series of solid soln. with a f.c.c. structure. In the binary Pt–Cu system at 850° C., PtCu, having a rhombohedral lattice, was formed. The same compound separated out upon slowly cooling the ternary alloy contg. Pt 50, Cu 45, and Ni 5 at.-%. Careful measurements of the lattice parameter of the ternary alloys showed deviations from Vegard's law; in this case the lattice expanded. It was further observed that the combined action of Cu and Ni in this respect affected the expansion of the lattice more than the same number of atoms of one of these elements alone.

***Investigation of the System Platinum–Silver–Palladium.** V. A. Nemilov, A. A. Rudnitsky, and R. S. Polyakova (*Izvest. Sekt. Platiny*, 1949, (**23**), 104–115; *C. Abs.*, 1951, **45**, 1940).—[In Russian]. Alloys contg. 10–80% of each of the components were investigated. For these alloys the initial and final temp. of solidification, B.H.N., sp. resistivity at 25° and 100° C., the temp. coeff. of resistivity, and microstructure were determined. The results are tabulated and presented diagrammatically.

***On the Question of the Electrical Resistance of Single-Phase Univalent Metals [Potassium–Rubidium Alloys].** N. V. Grum-Grzhimaylo (*Doklady Akad. Nauk S.S.S.R.*, 1951, **79**, (3), 461–462).—[In Russian]. G.-G. has found that the behaviour of K–Rb alloys is not in agreement with Nordheim's theory (*Ann. Physik*, 1931, [v], **9**, 641). The resistance is given by $\rho = [(T/b) - a] \times 10^{-6}$, where T is the abs. temp., and a and b are const. which ∝ the compn. of the alloy.—G. V. E. T.

***Experimental Study of the Surface Tension of Potassium Amalgam.** P. P. Pugachevich and O. A. Timofeevicheva (*Doklady Akad. Nauk S.S.S.R.*, 1951, **79**, (5), 831–832).—[In Russian]. The surface tension (σ) of amalgams contg. up to 0·6 at.-% K was determined at 20° C. in a special enclosed apparatus which could be tilted until the amalgam fell from a capillary under a measurable hydrostatic head. The apparatus contained a reservoir of conc. amalgam; by tilting about another axis, measured quantities of this could be added to the specimen and the experiment repeated for another concentration. σ fell rapidly from 470 for pure Hg to ~410 dynes/cm. for Hg-0·01 at.-% K, then gradually to ~375 dynes/cm. for Hg-0·68 at.-% K.—G. V. E. T.

***The Viscosity of Liquid Tin-Bismuth and Tin–Zinc Eutectic Alloys.** V. A. Konstantinov (*Zhur. Fiz. Khim.*, 1950, **24**, (8), 953–954).—[In Russian]. In order to check Sergeev and Polyak's measurements (*Zavod. Lab.*, 1947, **13**, 336; *M.A.*, **20**, 404), which showed a large increase in the viscosity of these alloys just above the crystn. temp., K. made measurements *in vacuo*, at temp. of 300°–135° C. for Sn–Bi and 300°–200° C. for Sn–Zn, by observing the damping of torsional oscillations of a steel sphere suspended by a steel strip in the molten alloy. The apparatus was arranged so that the alloy, previously melted *in vacuo*, could be poured under vacuum into the glass measuring vessel, which had previously been heated to the required temp. *in vacuo*; after measurements had been made, the alloy was displaced by gas into a lower evacuated reservoir. Results (as shown in graphs) ranged from 0·02 poise at 300° C. to 0·055 poise at 135° C. for the Sn–Bi eutectic and from 0·025 poise at 300° C. to 0·05 poise at 200° C. for the Sn–Zn eutectic. K. explains the difference between these and the much higher values of S. and P. by (i) the formation of an elastic oxide skin when the latter's measurements were made in air, with consequent high apparent viscosity, and (ii) the excess damping introduced by crystn. of the alloy on the steel sphere at temp. a few °C. above the liquidus.—G. B. H.

***A Realistic Approach to the Use of Titanium [Alloys].** Herman H. Hanink (*Product Eng.*, 1951, **22**, (11), 164–171).—The mech. properties of certain hot-rolled and annealed Ti alloys in various conditions of heat-treatment have been determined. It is concluded that notch sensitivity is a factor which may have to be taken into account in designing moving parts.—M. A. H.

Bearing Metals. M. E. Hargreaves (*Australasian Eng.*, **1952**, (Aug.), 46–56).—See *M.A.*, **20**, 245.—T. A. H.

Effect of Low Temperatures on Aircraft Metals. J. B. Johnson and D. A. Shinn (*Product Eng.*, 1951, **22**, (10), 187–192).—Data on the properties (tensile, fatigue, and impact) of two Al alloys, a Mg alloy, commercial Ti, and five steels at temp. between 75° and −320° F. (25° and −196° C.) are given.—M. A. H.

†A Survey of the Development of Creep-Resisting Alloys. N. P. Allen (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 1–10; discussion, 305; and *Engineering*, 1951, **171**, (4439), 235–238).—A comprehensive review of the many types of high-temp. creep-resistant alloys which have been developed in Great Britain, the U.S.A., and Germany since 1918. The alloys were required initially for exhaust valves of reciprocating engines, chem. plant, steam turbines, &c., and from 1939 researches were intensified to find materials for use in gas turbines. Full analyses are given of 40 of the most successful alloys which were available in the three countries during the pre- and post-war periods. The properties of these materials are compared diagrammatically by considering the stress to give plastic deformation of the order of 0·1% in 1000 hr. at temp. ranging from 500° to 800° C., and it is shown that the most satisfactory alloys in the high-temp–high-stress ranges are of the austenitic type, contg. Ni, Cr, and Co as major alloying elements, but that ferritic alloys may prove useful for lower temp. 17 ref.—D. M. P.

***Fatigue Tests [on Heat-Resisting Alloys] at Elevated Temperatures.** P. H. Frith (*Iron Steel Inst. Special Rep.*, 1952, (43), 175–181; discussion, 355–367).—Tests were carried out on Nimonic 80 (Cr 20, Fe 3·7, Ti 2·4%, remainder Ni), G32 (Cr 18·8, Fe 15·0, Ni 13, V 2·6, Mo 1·8, Nb 1·7%, remainder Co), and G18B (Cr 14, Ni 14, Nb 3·8, W 2, Mo 1·7%, remainder Fe) hollow test-bars with reversed bending with and without a superimposed static tension or bending stress at 20°, 500°, 600°, 700°, and 800° C. Static tensile tests were carried out at room temp. The results of a few fatigue tests on turbine blades manufactured from Nimonic 80 in various ways are also given.—D. M. P.

Materials and Performance [High-Alloy Steels for Gas Turbines]. A. T. Bowden and W. Hryniszak (*Iron Steel Inst. Special Rep.*, 1952, (43), 11–17; discussion, 368–378).—The advantages of heat-resisting high-alloy steels in the design of gas-turbines are considered. The effect of the allowable inlet temp. of the compressor on the performance of the turbine is also dealt with. 7 ref.—D. M. P.

Gas Turbine Performance and Materials. J. B. Bucher (*Iron Steel Inst. Special Rep.*, 1952, (43), 17–23; discussion, 368–378).—The design of heater-tubing calls for more information on the creep properties of heat-resisting alloys for periods exceeding 10,000 hr. at stresses of 1–6 tons/in.^{2} and temp. 600°–800° C. and of the effect of cold work on these properties. More investigation is needed into the welding of these alloys and the prevention of attack by V_2O_5 in the fuels.—D. M. P.

Future Needs in Materials for Land and Marine Gas Turbines. J. M. Robertson (*Iron Steel Inst. Special Rep.*, 1952, (43), 292–303; discussion, 368–378).—The range of uses of turbines is noted and the factors influencing the choice of materials for components are discussed. The materials at present in use are listed, and future requirements and possible fields for their fulfilment are suggested.—D. M. P.

Metal–Refractory Alloys. W. G. Lidman and H. J. Hamjian (*Product Eng.*, 1951, 22, (10), 147–150).—The properties of various metal–ceramic combinations are given, with details of the methods of fabrication.—M. A. H.

***Studies on the Solid Retainer Materials for Ball Bearings. I.—Wearing Properties of Various Metals and Alloys in Contact with Quenched Chromium-Bearing Steels.** Toranosuke Kawaguchi (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, 16, (1), 30–34).—[In Japanese, with tables and diagrams in English]. K. has used Suzuki's wear-testing machine to examine 13 Cu, 8 Al, and 5 ferrous alloys. The following are recommended as excellent retainer materials in the order of superiority: stainless steel (annealed and forged, Cr 13·1, C 0·2%); bronze (as cast, Sn 9·8, Zn 5%); malleable cast Fe; free-cutting steel; phosphor-bronze ((1) as cast, Sn 9·33, P 0·82%; (2) as forged, Sn 6·93, P 0·34%; and (3) as cast, Sn 15·95, P 0·94%); Silzin bronze (as forged, Zn 1·55, Si 3·32%); Al–9·63% Sn–0·89% Cu alloy (as cast); mild steel (annealed and forged); and Pb bronze (as cast, Sn 9·36, Pb 10·88%). Although the stainless steel is the best, it is difficult to use owing to its low thermal conductivity.—Author.

***Studies on the Solid Retainer Materials for Ball Bearings. II.—Running Test Results.** Toranosuke Kawaguchi and Tetsurô Yamaguchi (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, 16, (4), 228–231).—[In Japanese, with tables and diagrams in English]. Brass (Zn 28·39, Sn 1·86, Al 1·39, Fe 0·89, Si 0·12%), mild steel (0·4% C), malleable cast Fe, free cutting steel, Al–9·65% Sn–0·96% Cu alloy, Duralumin, and phenol-resin have been used as solid retainer materials in tests with bearings under no load for 150 hr. The wear of various bearing components constructed with the materials has been measured and is recorded in a series of graphs. Tests indicate that the most desirable materials are the malleable metals, e.g. cast Fe, brass, mild steel, and Al–Sn–Cu alloy.—Authors.

Elastic Constants and Hardness of Some Binary Alloys. M. J. Druyvesteyn and M. A. Dutoit Meyer (*Ingenieur*, 1950, 62, MK99–104; *C. Abs.*, 1951, 45, 3784).—The most simple cases are those in which complete miscibility occurs. The elastic modulus is a linear function of the compn. and hardness a quadratic function. In two-phase alloys both properties are linear functions, though the elastic modulus is somewhat lower and the hardness somewhat higher than calculated. In a series of examples (Hume-Rothery alloys) explanations are tried qual. from the crystal structures.

***Attempts at Using the Method of Microhardness in the Physico-Chemical Analysis of Metallic Systems.** S. A. Pogodin, L. M. Kefeli, and E. S. Berkovich (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, 17, 193–199; *C. Abs.*, 1951, 45, 2381).—[In Russian]. Six basic types of compn./microhardness relationships are derived for 2-component systems: (1) the components A and B are mutually soluble in all proportions; (2) A and B crystallize in their pure states from a liq. phase; (3) A and B form a compound A_mB_m; (4) A and B form a limited solid soln. with a eutectic; (5) the components form a daltonide, A_mB_n, which combines with its components to yield a limited solid soln., and in addn. 2 solid soln. are formed based on A and B, resp.; and (6) in addn. to 2 solid soln., based on A and B, resp., a berthollide is formed. The above was tested experimentally on the systems Sn–Mg and Pb–Mg. No solid soln. based on Mg_2Sn were found. A small region occupied by a solid soln. based on Mg_2Pb was found.

***Determination of the Viscosity of Molten Alloys and Its Applications.** S. V. Sergeev and E. V. Polyak (*Zavod. Lab.*, 1947, 13, 336–344; *C. Abs.*, 1948, 42, 1545).—[In Russian]. Abs. viscosities η were determined from the photographically recorded logarithmic decrements of torsional oscillations of a steel sphere (20–25 mm.) suspended on a 50–60-cm. steel wire 0·20–0·25 mm. in dia., calibrating the apparatus with water and benzene and calculating η by the formulæ of Verschaffelt (*Proc. K. Akad. Wetensch.*, 1916, 18, 840). It was found that: (1) In pure Al, Bi, Sn, and in eutectic Silumin (Al 88, Si 12%), η increases sharply with falling temp. near the f.p.; the curve of log η against $1/T$ is linear over a wide temp. interval, but deviates from linearity close to freezing; this indicates the beginning of structure formation in the liq. state at the precrystn. stage, or the existence of superheated structure above the m.p. (2) The binary Al–Si system (up to 19% Si) at 700° C. shows, besides the min. value of η at the eutectic compn., an initial increase with increasing Si up to a sharp max. at ~0·8% Si. Similarly, Al–Cu alloys (up to 54% Cu) at 670°–680° C. show an initial increase of η with Cu content up to a sharp max. at ~2% Cu, then a fall to a min. at the eutectic compn. In Bi–Sn alloys (complete system), the η curves for 270°, 320°, 350°, and 400° C. all have a max. at ~4% Sn; the 200° and 235° C. curves have a max. at the eutectic; the 235°, 270°, 320°, and 350° C. curves have a 2nd max. at ~96% Sn. The two max. at low Bi and at low Sn are sharp at lower temp., but become increasingly flatter with rising temp.; the max. at the eutectic, sharp at 200°–235° C., disappears between 235° and 270° C. and becomes a shallow min. from 270° C. upwards; the 600° and 1000° C. curves are practically horizontal straight lines. The temp. dependence of the max. of η is in keeping with the gradual disappearance of the structure in molten alloys with rising temp., as revealed by X-rays. (3) Modification of Silumin by addn. of 0·1% Na raised η to such an extent that oscillation was at first suppressed altogether; after 5–7 min., η was still 15 times higher than for unmodified Silumin and continued to decrease over a period of 40 min. Addn. of Na, Ca, Mg, Ti, Sn, Si, and Cu, in amounts of the order of 0·1%, to pure Al, had an analogous effect. Since the "modifiers" are surface-active towards Al and Silumin, their action can be interpreted by adsorption at the interface of the crystn. nuclei and the melt, resulting in inhibited crystal growth and a finer grain; the effect on η indicated that this takes place even in the liq. phase long before crystn.; thus capillary activity of the modifiers and sharp increase in η were already pronounced at $\sim$100° C. above the f.p.

The Structure and Properties of Magnetic Alloys. D. W. Davison (*Australasian Eng.*, 1952, (Sept.), 44–49).—" Hard " alloys for permanent magnets and " soft " alloys for coil cores are discussed, with particular ref. to atomic arrangement. The magnetic properties of various alloys of these two types are tabulated.—T. A. H.

***Intrinsic Magnetization in Alloys.** W. J. Carr, Jr. (*Phys. Rev.*, 1952, [ii], **85**, (4), 590–594).—A discussion is given of the saturation magnetic moments of alloys of the Fe group of transitional metals, on the assumption that the *d* electrons are effectively tightly bound. It is suggested that where the overlap of the 3*d* shells is small, the spins of the *d* electrons will tend to be aligned parallel, but with larger overlaps an anti-parallel arrangement will be favoured. The demarcation is drawn between Ni–Mn and Co–Mn on an empirical basis. The negative spin interaction between Ni–Cr and Ni–V is used to explain why Cr and V decrease the saturation magnetization of Ni whereas Co increases it. The theory also explains the peculiar effect of Mn, and attributes the greater saturation magnetization of ordered than of disordered Ni_3Mn to the elimination of Mn–Mn interactions. The behaviour of Co-rich alloys is similarly explained. C. agrees with Zener (*ibid.*, (2), 324; *M.A.*, **20**, 134) that in Fe-rich alloys the spin interaction is negative, so that on alloying Fe behaves differently.—P. C. L. P.

---

3 — STRUCTURE

(Metallography; Macrography; Crystal Structure.)

[For all abstracts on the constitution of alloy systems, including X-ray studies, see 2—Properties of Alloys.]

***A Method of Determining Orientations in Aluminium Single Crystals and Polycrystalline Aggregates.** G. E. G. Tucker and P. C. Murphy (*J. Inst. Metals*, 1952–53, **81**, (5), 235–244).—A method is described for constructing (100) pole figures from angular measurements of the etch-pits that are developed in micrographically prepared commercial and super-purity Al. Suitable etching reagents are indicated, and the technique of angular measurement using a met. microscope is discussed in detail. A table of angular values which reduces the calculation necessary in plotting the stereographic projections from the etch-pit data has been constructed, and is reproduced. It has been found that the pole figures prepared by this method agree well with those produced from data obtained by X-ray-diffraction techniques. —AUTHORS.

***On the Effect of Scratched Lines on the Deformation of Aluminium Single Crystals.** Seiichi Karashima (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (3), 140–142).—[In Japanese, with tables and diagrams in English]. K. reports that, in an investigation of plastic deformation of coarse-grained Al specimens on which index lines were scratched for measuring strains, it was found that in some grains lines ‖ the index lines appeared on the other side of the specimens, after certain extensions. The phenomenon was studied in detail by back-reflection Laue X-ray photography, microphotography, &c. K. attributes the cause to a state somewhat similar to a polycryst. one produced near the index lines by the scratching which inhibited slip.—AUTHOR.

Practical Control of Recrystallization in Light Alloy Sheets. —— (*Alluminio*, 1948, **17**, (7/8), 357–368).—A report from the Istituto Sperimentale dei Metalli Leggeri. In order to forecast the grain-size of recrystallized light-alloy sheets after working with intermittent annealing, the following method is used by the I.S.M.L. A number of tapered strips 200 cm. long and 25 and 15 cm. wide, resp., at the ends are taken from the alloy sheet under observation. The samples are subjected to permanent deformations followed by annealing repeated as many times as these operations are carried out in practice. The deformations are produced in a tensile testing machine until the samples break. Failure occurs in the narrower part, giving a progressive degree of deformation along the length of the samples. Annealing is carried out at 380°–500° C. (depending on the alloy) in order to produce a recrystn. structure which will vary with the degree of deformation. It was found that a final reduction of 25–30% gave the right crystal-size to avoid the formation of large crystals in subsequent plastic deformations and heat-treatments. Annealing before soln. treatment of Duralumin sheets produces a finer grain after repeated deformation and recrystn. Mg as an addn. to alloys of the Duralumin type was found to be favourable to the formation of finer grains.—F. E. M.

***A Study on the Recrystallization Temperature of Refined High-Purity Aluminium. III.—The Effect of Small Amounts of Impurities. IV.—The Relation Between the Purity of Aluminium and the Recrystallization Temperature. V.—The Effect of Heating Time. VI.—Effect of Initial Grain-Size.** Namio Kawashima and Yuzo Nakamura (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 26–29; (2), 73–76; (4), 183–188; (5), 246–249).—[In English]. [III.—] The effects of Fe 0·009–0·412, Cu 0·008–0·439, Si 0·022–0·282, Mg 0·009–0·463, Zn 0·023–0·357, Mn 0·02–0·140, and Ti 0·02–0·12% on the recrystn. temp. of high-purity Al (99·99%) was investigated with the aid of X-ray analysis. Results show that Cu, Mg, and Zn raise the temp. of initial recrystn., whereas Fe and Si ($>$0·05%) lower it. Mn and Ti within the solubility limits also raise it, but the temp. range of recrystn. is narrowed. [IV.—] Six Al samples with purity between 99·5 and 99·998% have been examined. Results show that: (i) there is a max. temp. of initial recrystn. which can be explained by the results of the previous investigation; (ii) the max. initial temp. is at 99·92% purity and the final temp. is highest at 99·992 %purity; and (iii) the recrystn. temp. of the highest-purity Al (99·998%) is $\sim$230° C., but recrystn. seems to begin at room temp. after ageing for a few months. [V.—] The results of X-ray analysis and hardness detn. of Al with purity of 99·998, 99·992, 99·92, and 99·5% after various heat-treatments are reported and discussed. It has been found that the recrystn. temp. of high-purity Al is lowered by increasing heating time and a linear relation exists between $1/T$ and $\log t$ for all specimens of Al, where T = abs. heating temp. and t = heating time. It has been found also that the softening rate is highest in 99·998% Al and lowest in 99·992% Al. [VI.—] A report is given of an investigation, by hardness, microscopic, and X-ray methods, of the effect of grain-size on recrystn. temp. of 99·9% and 99·3% Al. The results show that the larger the grain-size, the higher the recrystn. temp.

Al Purity, %	Initial Grain-Size, grains/mm.	Recrystn. Temp., °C.
99·9	6	360
	400	300
99·3	420	320
	600	280

—S. R. W.

***Intercrystalline Fracture in Aluminium–Zinc Alloys.** Eric Charles Perryman (*Compt. rend.*, 1952, **235**, (16), 884–886).—Intercryst. brittleness in the Al–Zn alloys, especially when the Zn is increased from 8 to 13%, has already been observed (P. and Blade, *J. Inst. Metals*, 1950, **77**, 263; *M.A.*, **17**, 911) as being connected with the discontinuous pptn. occurring after ageing at room temp. Ageing experiments were carried out on the 13·4% Zn alloy for 10 days at room temp. and for 284 days at 125° C. By this latter treatment equilibrium in the alloy was obtained, the structure consisting of a precipitate rich in Zn in a 5% Zn solid soln. The specimens were corroded under a stress of 7·9 kg./mm.^{2} in a 3% salt soln.; the specimen aged for 10 days cracked after 2 days; the specimen aged for 284 days had not failed after 57 days. These results and a micrographic examination confirmed the hypothesis that intercryst. fracture of the Al-Zn alloys is due to the formation of a new solid soln. at the grain boundaries during ageing and that fracture is due to a relative movement between the grains.—J. H. W.

***Pseudo-Kossel Pattern Produced by the Capillary X-Ray Tube and Its Application [to Aluminium].** Tohoru Imura (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 10–15).—[In Japanese, with diagrams in English]. Characteristic diffraction patterns due to divergent radiation were obtained by both transmission and reflection methods. Single-crystal plates of Al ~0·1–0·5 mm. thick were used as specimens, and the radiation was CuK_α. A study of the characteristic patterns relating to the perfectness of deformed crystals showed that: (1) the diffraction patterns were composed of black and white lines and were like those of Kossel; they were thus called "pseudo-Kossel patterns"; (2) the accuracy of the measurements of lattice dimensions and orientations was much better than is obtained by the routine method; and (3) the results of the perfectness of the crystals were little different from those obtained by Heidenreich and Shockley (*J. Appl. Physics*, 1947, **18**, 1029; *M.A.*, **15**, 503) using the electron-diffraction method. The diffusion or disappearance of pseudo-Kossel lines was indefinite, and some of them remained even after considerable deformation.
—AUTHOR.

***The Study of Work-Stress and Deformation of Metals in Single Crystals. III.—Study [of Aluminium] by the Capillary X-Ray Tube.** S. Yoshioka and T. Imura (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (2), 77–81).—[In Japanese, with tables and diagrams in English]. The structure of an Al single crystal was examined by a capillary X-ray tube designed by Y. and I. The change of lattice structure with varying tensions, the recovery, and recrystn. were progressively observed. The diffraction patterns obtained were composed of ellipses of the back-reflection type and in the transmission type they were analogous to Kossel patterns. It was thought that the patterns indicated the state of the atomic arrangement in the direction nearly ⊥ the surface of the specimen. Results showed that: (1) by the use of the tube quant. measurement of the lattice distortion, d, which is thought to be impossible by the usual Laue method, became possible; (2) d varied with the deformation, but the order was indefinite; and (3) the effect of recovery by heating was clearly indicated by the reappearance of the sharpness of the reflection lines, though complete restoration of their initial sharpness could not be expected without recrystn. by full annealing.—AUTHORS.

***†The X-Ray *K* Absorption Edges of Covalently-Bonded Chromium, Manganese, Iron, and Nickel.** G. Mitchell and W. W. Beeman (*Proc. Conf. Applications of X-Ray Spectroscopy to Solid State Problems* (*Univ. Wisconsin*), **1950**, 63–68).—Recently measured absorption edges of covalent Ni complexes, and edges of Cr, Mn, and Fe in similar complexes, found in the literature of the subject, are briefly discussed. An empirical correlation is established between the low-energy K edge absorption and empty $4p$ orbitals in 17 covalent complexes of elements of the first transition series. This correlation depends on the assumption that tetrahedral bonding orbitals in Cr and Mn are largely sd^3, and in Ni almost completely sp^3.
—J. S. G. T.

***Structure and Oxidation Kinetics of Cathode Copper.** (Marchenko and Sysoev). See col. 418.

***On the Superlattices of the Cu–Au System.** Makoto Hirabayashi (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (2), 67–72).—[In English]. See *J. Phys. Soc. Japan*, 1951, **6**, 129; *M.A.*, **19**, 773.—S. R. W.

***Theory of Order–Disorder Transformations in AuCu$_3$.** Gérard Fournet (*Compt. rend.*, 1952, **234**, (21), 2049–2051).—The hypothesis laid down for the order–disorder transformations in the Cu–Zn alloys (*ibid.*, 1951, **232**, 155; *M.A.*, **18**, 601) are applied to AuCu$_3$. According to Yvon (*Cahiers Phys.*, 1945, (28), 1; *M.A.*, **18**, 601), the energy, $W(r)$, is defined by: $W(r) = W_{\text{CuCu}}(r) + W_{\text{AuAu}}(r) - 2W_{\text{CuAu}}(r)$, where $W_{\text{AB}}(r)$ is the mutual potential energy of an atomic couple, A + B, at a distance r from each other. This theory gives the possible soln. for large distances. Theoretical estimates of the order at a small distance were found to be in good agreement with the observed facts.—J. H. W.

***An Electron-Diffraction Investigation of the Structure of Electrodeposited Coatings on Iron Single Crystals.** D. J. Evans and M. R. Hopkins (*J. Electrodepositors' Tech. Soc.*, *Advance Copy*, 1952, **28**, (8), 10 pp.).—Results are described which have been obtained when Zn (h.c.p.), Cu (f.c.c.), and Ni (in the usual f.c.c. and the more unusual h.c.p. form) are deposited under different plating conditions on single crystals of b.c.c. Fe having planes of high Miller indices ∥ the surface. Full details are included of soln. compn. and plating conditions. In the case of Zn, the (0001) plane of the deposit was ∥ the (110) plane of the Fe. When Ni is deposited in the h.c.p. form, the relative crystallographic dispositions of the deposit and substrate are the same as for Zn. Cu and Ni in the f.c.c. forms were found to deposit with their (111) planes ∥ the (110) plane of the Fe. The orientation of the deposit with respect to that of the substrate is independent of the crystallographic disposition of the surface of the substrate. The results obtained emphasize the importance of the compn. of the plating soln. on the structure of the deposit, since with some soln. only a polycryst. deposit was observed under a variety of plating conditions. 5 ref.—G. T. C.

***Metallographic Determination of the Temperatures to Which [80 : 20 Nickel–Chromium Alloy] Parts of Jet Engines Are Subjected in Service.** Jean Poulinger, (Mme) Charlotte Bückle, and Pierre Jacquet (*Compt. rend.*, 1952, **235**, (14), 718–720).—The method using coloured films previously described (B. and P., *ibid.*, 1951, **233**, 869; *M.A.*, **19**, 446, and P. and J., *Rev. Mét.*, 1952, **49**, 541; *M.A.*, **20**, 255) has been applied to the detn. and location with great precision of the mean temp. attained on the faces and in the interior of a jet-engine blade during service. The results were compared with those of hardness measurements. Two possible extensions of this method are: (1) in the range 600°–850° C. it can be used to determine the mean local temp. reached by 80 : 20 Ni–Cr parts in actual conditions of service; and (2) in other temp. ranges the structural changes can be followed.—J. H. W.

***On the Order–Disorder Transformation in Ternary [Nickel–Iron–Copper and Nickel–Iron–Molybdenum] Alloys.** Emilio Josso (*Rev. Mét.*, 1952, **49**, (10), 727–732).—J. investigated the action of addn. of Cu and Mo on the order–disorder transformation of ferro-nickels of compn. about that of Ni$_3$Fe. For pure Fe–Ni and for the Cu-contg. alloy, sufficient variation in the sp. vol. occurs to be measured dilatometrically. For the Mo-contg. alloy thermomagnetic measurements were more sensitive. The transformation temp. for the sintered and for the cast and forged alloys were as follows: Fe–Ni (Ni 78·45%), 490° and 505° C.; for the alloy contg. 5·6% Cu, 468° and 475° C.; for the alloy contg. 3·6% Mo, 472° and 495° C., resp. These differences are too great to be due solely to experimental error, and further work is being undertaken to find the cause. Resistivity experiments showed that Cu and Mo lower the critical temp. of transformation (especially Cu) and spread the transformation over a range of temp. with an equilibrium between the ordered and disordered states. The critical transformation temp. in the Fe–Co alloys is max. at 725° C. for the 44·4% Co alloy, but was observed down to 35% Co at 655° C. The addn. of 1% Cr or V lowers the temp. by some 12° C., possibly owing partly to the increase of the thermal hysteresis. Other elements affect this property to different degrees. With even <1% Cr or V an intermetallic phase appears after prolonged annealing at 850°–900° C., tending to redissolve above 950° C. The formation of this precipitate, by removing the Cr or V from the solid soln., has the direct consequence of diminishing the effect of these elements on the structure: it should be inhibited if the max. efficacity of the addn. is to be preserved. 15 ref.—J. H. W.

Relationship Between Lattices of Types NiAs and Ni$_2$In and Certain Rhombic Ones. P. I. Kripyakevich (*Doklady Akad. Nauk S.S.S.R.*, 1951, **79**, (3), 439–442).—[In Russian]. The relationship between the lattice structures of Ni$_2$In, NiAs, MnP, Co$_2$Si, Co$_2$P, FeB, and Cr$_3$C$_2$ is discussed.—G. V. E. T.

***The Crystal Structure of Electrodeposited Silver.** D. M. Layton (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (9), 6 pp.).—Ag was deposited on polished brass cathodes to a thickness of 25,000 Å. from various Ag-plating soln. and at various temp. and c.d. At low c.d. the crystal orientation, determined by electron-diffraction examination, was (111), characteristic of simple lateral growth. At slightly higher c.d. the crystals were orientated so as to have a (112) plane || the substrate, though others in (111) orientation also occurred. Higher c.d. resulted in some of the deposit crystals growing with the atoms in a h.c.p. lattice in $(10\bar{1}0)$ orientation. The general appearance of deposits from a simple K–Ag cyanide bath was smooth and glossy-white, but those from baths contg. added carbonate or chloride had a satin sheen. In all the baths there was little variation in crystal size with c.d., but in deposits from carbonate and chloride baths, the crystals were larger at 60° than at 20° C. 17 ref.—G. T. C.

The Physical and Chemical Changes Which Accompany the Polishing of Metals. A. J. W. Moore (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (1), 8 pp.).—The following factors are briefly considered: area of contact between the sliding surfaces during polishing, the temp. produced during polishing, and the nature of and reason for surfaces damaging each other during sliding. The properties of a polished metal, the mechanism of polishing, the role of the abrasive, and the structure of the deformed layers are all discussed. 21 ref.—G. T. C.

Cathodic Vacuum Etching of Metals. Don M. McCutcheon and William Pahl (*Steel Processing*, 1949, **35**, (11), 590–591).—A short description, indicating the advantages of the process.—S. R. W.

***An Experimental Study of Electropolishing.** J. Edwards (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (2), 16 pp.).—The smoothing action during electropolishing is discussed, and the theoretical smoothing efficiencies obtainable in certain specific cases are described. A method for the experimental detn. of smoothing efficiency is given. This employed a positive copy of a plain-cut "Microgroove" gramophone record, electroformed in Cu. The experiment was arranged in a manner which ensured that current distribution on the specimens was as uniform as possible. Microscopic examination after polishing enabled the smoothing action to be assessed. A few experiments were also carried out, using chem. polishing. It is shown that in the case of both electrolytic and chem. polishing, the process is diffusion-controlled and that smoothing results from the fact that a diffusion layer formed over a rough surface has steeper concentration gradients on convex than on concave portions so that the prominences are preferentially dissolved. The max. rate of smoothing was obtained when the dissoln. was entirely diffusion-controlled. The limiting-current conditions prevailing in the electropolishing range correspond to the steepest possible concentration gradients, and thus to the max. diffusion force. Various practical recommendations for electropolishing are made. 8 ref.—G. T. C.

The Electrolytic Polishing of Micro-Specimens. J. W. Holleman and W. A. Schultze (*Metalen*, 1951, **6**, (16), 306–311).—[In Dutch]. The theory of electrolytic polishing is surveyed, and the anodic polishing of microscope samples is then discussed in detail. Apart from steel, Cu, Zn, and Cd have been polished successfully with an electrolyte contg. 2 parts of HNO$_3$ (sp. gr. = 1·4) and 1 part of C$_4$H$_6$O$_3$. Bath temp. <25° C. and anodic c.d. 10 amp./cm.^{2}, etching time being 5–10 sec.—I. S. M.

***On Electrolytic Polishing in Sulphamic Acid Solutions.** Sakae Tajima and Takemi Mori (*Compt. rend.*, 1952, **234**, (20), 1976–1978).—Among the metals tried (Al, Mg, Fe, Ni, Sn, Pb, Cu, Zn, Cd, Mo, W, and Ag) only Cd, Zn, and Mo could be electropolished in aq. sulphamic acid (NH$_2$HSO$_3$), although Cu, Sn, and Pb took on locally a certain brilliance. As in other cases of electropolishing, the current/voltage curve shows periodic oscillations that result in alternate passivation and activation, suggesting that the mechanism in sulphamic acid is essentially the same as that in other electrolytes. The conditions for polishing in a 5% soln. of the acid are: Zn, 10–14 V., c.d. 140–240 amp./dm.^{2} and voltage corresponding to the oscillations 9·5 V.; Cd, 8–12 V., 140–220 amp./dm.^{2}, and 6–9 V.; Mo, 17–20 V., 100–140 amp./dm.^{2}, and 6–20 V., resp. For the same normality (1·1 *N*) and for the same *d* (1·060), sulphamic acid gives a better polish than H$_3$PO$_4$. The experiments prove that from the point of view of polishing, the properties of an anolyte are more important than those of the mass of the bath. In general, however, the more conc. or more viscous soln. is to be preferred. Various non-ionizing addn. (such as glycerine) improve the quality of the polish on Zn and Cd, but have no effect on Mo, for which the addn. of chromic acid gives excellent results. The periodic oscillations are observed down to −2° C., Zn and Cd being polished just as well as at room temp., but Mo less so.—J. H. W.

Abnormal Structure in Alloys of Eutectic or Eutectoid Type. IV.—Abnormal Structure in Ternary Eutectic. Kyuya Nagasaki (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (1), 23–26).—[In Japanese, with diagrams in English]. The eutectic and eutectoid structures of ternary alloys are considered. An abnormal structure of the eutectic which is theoretically described with the aid of a ternary diagram is exemplified by the Cd–Bi–Sn system. Al–Cu–Si and Al–Cu–Mg systems are also used as examples.—AUTHOR.

***Evidence of Heterogeneities in a Solid Solution in Equilibrium.** Christopher B. Walker, Jean Blin, and André Guinier (*Compt. rend.*, 1952, **235**, (3), 254–255).—The distribution of atoms in a homogeneous and disordered binary solid soln. is generally perfectly random, but this is not so in solid soln. above the order–disorder transformation temp. The distribution of atoms was studied by X-rays in the case of a solid soln. where the two sorts of atoms have a tendency, not to be ordered, but to become separated, i.e. a solid soln. at a temp. above that at which it decomposes into a phase poor in the dissolved element and a precipitate richer in that element, as, for example, the Al–20% Ag alloy. It was found that the atoms are not distributed perfectly at random. The curves obtained can be explained by the presence of very small labile aggregates of Ag atoms more frequent than required by the laws of simple chance. These experiments constitute a first direct proof of the existence in solid soln. of the fluctuations introduced under various names ("nuclei", "embryos") by different theorists to explain the nucleation of a new phase.—J. H. W.

***Crystallization and Heat-Transfer in a Stream of Liquid Metal Flowing Along the Surface of a Mould.** E. W. Fell (*Proc. General Discussion on Heat Transfer* (Inst. Mech. Eng. and Amer. Soc. Mech. Eng.), 1951, 276–278).—Flow of liq. metal in a mould affects the solidification process in casting, very considerably. A theoretical discussion of heat transfer in molten metals flowing in moulds is developed with a view to elucidating the solidification process. The inclination of "columnar" crystals (separating out from the melt as it moves along the mould) to the liquid/mould interface varies from 0° to 30°, depending apparently upon the velocity of the melt along the surface. The inclination (θ) is towards the direction from which the stream flows, and is correctly attributed by Phragmén, in the case of normal steel ingots, to natural convection currents in the liq. metal. Experimental values of θ were found as follows: (1) Pb poured near the mould surface, with forced circulation, 15°, and (2) Al–10% Mg alloy, poured near a water-cooled mould surface with forced circulation, 20°. Motion of the melt promotes the appearance of "banded" crystal structures probably owing to inter-

ference with crystal growth by the washing action of the stream. The temp. distribution in a thin plate immersed in streams of H_2O, Hg, Pb, Fe, and stearic acid was determined, when the temp of the main stream was 50° C. above that of the plate. The temp. distribution in streams of various melts is calculated for: (i) laminar flow in a circular pipe with uniform wall-temp. gradient, and (ii) for boundary layer flow along a flat surface. Turbulent flow is not discussed. The rate of approach of a melt to a growing crystal is greatest for that part of the crystal which faces the oncoming stream. The motion, by causing the approach of somewhat similarly oriented atoms towards those of the crystal, therefore promotes crystal growth where the crystal faces the stream.
—J. S. G. T.

Factors Affecting the Choice Between Cubical and Hexagonal Close Packing. J. A. Prins, J. M. Dumoré and Lie Tiam Tjoan (*Physica*, 1952, 18, (5), 307–314).—[In English]. By summing the London interaction energy terms for cubic and h.c.p. lattices, it is shown that hexagonal packing is more stable by ~1 part in 1000. Changes in the exponent in the energy formula give similar results and the difference remains of the same sign even when the lattice in compressed or extended 10% in the trigonal direction. The fact that inert gases and many metals form cubic packing is discussed, and it is suggested that lack of spherical symmetry in the electron shell is the reason.—E. O. H.

***On the Theory of Kinking.** F. C. Frank and A. N. Stroh (*Proc. Phys. Soc.*, 1952, [B], 65, (10), 811–821).—The stability of a kink band is investigated math. The model chosen is a two-dimensional one, consisting of a thin plate of material, sheared out of alignment with the matrix in a direction transverse to the slip directions. The edges of the plate are made of dislocation walls. It is shown that if the angle of shear $\gg$ ~3½°, stress concentrations are present at the ends of the plate, which will, if the applied stress is sufficiently large, cause the kink to grow into the macroscopic kink bands or deformation bands observed in practice. The appn. of this theory to twinning, martensitic transformations, and other deformation phenomena is also discussed.—E. O. H.

The Production of Large Tensile Stresses by Dislocations. J. S. Koehler (*Phys. Rev.*, 1952, [ii], 85, (3), 480–481).—A letter. K. shows that an array of edge dislocations moving on the same glide plane and held up by an obstacle, causes large tensile stresses and lattice dilatation over fairly extensive distances.—P. C. L. P.

***X-Ray Diffraction by Face-Centred Cubic Crystals with Deformation Faults.** M. S. Paterson (*J. Appl. Physics*, 1952, 23, (8), 805–811).—Theoretical. The effects of stacking faults || one {111} family due to plastic deformation on the powder pattern are calculated as a function of fault frequency. The treatment resembles Wilson's treatment of the Co transformation (*Proc. Roy. Soc.*, 1942, [A], 180, 277; *M.A.*, 9, 244). The result is to broaden the (200) and (400) lines while moving their peaks in opposite directions; (331) is split into two broadened components, while the remaining lines retain some sharp unchanged components together with broadened slipped ones. If more than one family of {111} planes in each grain is faulted, the sharp unchanged components disappear. Particle-size broadening can be distinguished from the fact that it affects all spectra equally; the best practical way of disentangling the effect is to use a Fourier analysis of line shape. Differences in the diffraction by growth-faulted crystals are gone into.—R. W. C.

†Notes on X-Ray Spectra and the Theory of Solids. N. F. Mott (*Proc. Conf. Applications of X-Ray Spectroscopy to Solid State Problems* (*Univ. Wisconsin*), 1950, 1–7).—The following matters are very briefly discussed: X-ray absorption by free atoms, absorption and emission by insulators, the electronic structure of metals, e.g. Zn or Al, dissolved in Cu, absorption by pure metals, and metallic structure. Discussing Pauling's recent contribution to the theory of the metallic state, M. considers that P.'s use, more especially in the case of the transition metals, of various atomic states with different numbers of *d* electrons does not contradict the band picture of metallic structure, but does not agree with P.'s concept of resonating bonds; if the wave-function of a metal could be described mainly in terms of bonds travelling through the lattice, then the electronic sp. heat would vary as $T^{3/2}$ instead of as T, as observed. M. has suggested that all electrons in an alloy, e.g. dil. Zn in Cu should be described by Bloch-type wave functions; such an approach is consistent with the Jones–Hume-Rothery model, according to which the Zn electrons "belong to the lattice". Recent calculations by Friedel suggest that this approach to the subject is wrong; and an alternative, probably not inconsistent with the Jones–Hume-Rothery model, is suggested and discussed. 14 ref.
—J. S. G. T.

†The Fine Structure Near X-Ray Absorption Edges of Crystalline Matter. S. Kiestra (*Proc. Conf. Applications of X-Ray Spectroscopy to Solid State Problems* (*Univ. Wisconsin*), 1950, 26–38).—A critical review of the literature, published since 1920, relating to the fine structure of X-ray adsorption line edges, leads to the following conclusions: (1) close to the edge, the behaviour of the absorption coeff. is mainly a property of the atom concerned; (2) in the adjacent region, the fine structure is determined by the immediate surroundings of the atom; (3) in the region of highest energy, the fine structure depends essentially on the whole crystal lattice; and (4) for the transition metals and their compounds (1) comprises the region up to ~40 eV. from the edge, (2) the region ~40–150 eV., and (3) the region above 150 eV. 27 ref.—J. S. G. T.

A Review of Past Work in X-Ray Spectroscopy at Uppsala. A. E. Sandström (*Proc. Conf. Applications of X-Ray Spectroscopy to Solid State Problems* (*Univ. Wisconsin*), 1950, 39–44).—Work relating to X-ray spectroscopy carried out at Uppsala since 1925, and concerned with the precision measurement of diagram lines, the fine structure of absorption lines, and atomic structure is briefly and authoritatively reviewed. 15 ref.
—J. S. G. T.

The Recent X-Ray Spectroscopic Work at the University of Paris. Y. Cauchois (*Proc. Conf. Applications of X-Ray Spectroscopy to Solid State Problems* (*Univ. Wisconsin*), 1950, 8–25).—The following matters studied within recent years at the Laboratoire de Chimie Physique of the University of Paris are briefly reviewed: (1) X-ray spectroscopy, employing the bent-crystal technique; (2) precautions necessary when recording and measuring absorption edges; (3) comparison of *K* emission "band" and absorption "edge" from Ni, As, and Cu compounds; (4) *K* absorption spectra of Ni diluted in Cu; (5) possible influence of crystal imperfections on the X-ray spectra of metals; (6) soft X-ray spectroscopy; (7) X-ray absorption spectra of solids and crystal structure; and (8) *L* spectra of pure Cu and pure Sn. 122 ref.
—J. S. G. T.

***Production of Monochromatic X-Radiation for Microradiography by Excitation of Fluorescent Radiation.** T. H. Rogers (*J. Appl. Physics*, 1952, 23, (8), 881–887).—The method described was originally developed for medical appn. Monochromatic X-rays are required for quant. detn. of variation of compn. across a thin specimen. X-rays from various anticathodes were allowed to fall on secondary targets ranging from Ti to Mo. If the wave-length of the incident radiation is only slightly smaller than the absorption edge of the secondary target and if the tube voltage is not too high, the secondary radiation consists almost exclusively of monochromatic fluorescent X-rays with little Compton scattering. The intensity is reduced by a factor of ~10 compared with the primary beam.—R. W. C.

The Examination of Metal Surfaces by Means of the Electron Microscope. I.—Technique of Preparation. V. Ch. Dalitz and J. A. Schuchmann (*Metalen*, 1952, 7, (9), 153–161).—[In Dutch]. A detailed description of a method developed by D. and S. An etched metal is placed in a vacuum bell-jar and a thick Ag layer evaporated upon it. The negative Ag replica is then vacuum-coated with a thin film of SiO, and the Ag side brought into contact with a 40% HNO_3 soln. After the Ag has been dissolved, the positive replica is shadow-treated. Another method is the polymerization of a hydrocarbon film on the Ag replica, instead of SiO. The Ag/SiO or Ag/hydrocarbon method may be used for such metals and alloys as steel, Cu, Pt, Al, brass, bronze, and bearing metal. 7 ref.—I. S. M.

***Splitting of Bands in Solids.** E. Katz (*Phys. Rev.*, 1952, [ii], **85**, (3), 495–496).—A letter. Slater (*ibid.*, 1951, [ii], **84**, 179; *M.A.*, **20**, 23) has suggested that if the true unit cell allowing for superlattice formation and electron spin is an integral multiple of the basic unit cell, then discontinuities will arise within the Brillouin zone. K. considers that in most cases, this type of discontinuity is likely to be important only if the zone boundary introduced by the perturbation coincides approx. with an equi-energy contour within the original zone.—P. C. L. P.

†The Band Structure of Metals. G. V. Raynor (*Rep. Progress Physics*, 1952, **15**, 173–248).—After a brief outline of the historical development of the electron theory of metals, detailed consideration is given to the calculations which have been made on various real metals. For the univalent metals (alkali metals and metals of Group 1B) reasonably good agreement between experimental and theoretical values of the binding energy, lattice const., and elastic const. has been obtained, but it is emphasized that the difficulties increase considerably with metals of higher valency. The more recent work covered by the report includes that of Herring and Hill on Be, of Raimes and of Jones on Mg, of Matyas and of Leigh on Al, and of Fletcher and Wohlfarth on Ni. In the final section alloys are considered. The difficulties implicit in the usual assumption that for atoms of similar sizes the valency electrons are shared equally between all the atoms, are discussed with ref. especially to soft X-ray spectra. The possibility of bound states existing round a solute atom is briefly considered. 78 ref.—J. W. C.

***Collective Description of Electron Interactions. II.—Collective *vs.* Individual Particle Aspects of the Interactions.** David Pines and David Bohm (*Phys. Rev.*, 1952, [ii], **85**, (2), 338–353).—Cf. *ibid.*, 1951, [ii], **82**, 625; *M.A.*, **19**, 355. The long range of Coulomb interactions in dense electron gases makes each electron interact simultaneously with every other so that they behave collectively, in a manner conveniently analysed in terms of electron d fluctuations. These fluctuations are of two types: organized long-distance oscillations and local fluctuations by random motion. For phys. phenomena involving distances greater than the Debye length, the long-distance oscillations aspect is dominant, but phenomena involving shorter distances may be treated satisfactorily as though the electrons were a collection of effectively free particles. Consideration of the motion of an electron with higher-than-average speed indicates that it gives rise to a wake of collective oscillations. Experimental evidence on the energy loss of fast electrons in thin films of Al and Be is in agreement with the theory.—P. C. L. P.

The Stopping Power of a Metal for Charged Particles. David Pines (*Phys. Rev.*, 1952, [ii], **85**, (5), 931).—A letter. P. and Bohm's treatment (*ibid.*, (2), 338; preceding abstract) is used to calculate the total energy loss of a fast-moving particle to the conduction electrons of a metal as a sum of that expended in large-scale collective oscillations and in short-range collisions. The expression obtained is only slightly different from that given by other workers. —P. C. L. P.

***Motion of an Electron in a Perturbed Periodic Potential.** E. N. Adams, II (*Phys. Rev.*, 1952, [ii], **85**, (1), 41–50).—Slater's (*ibid.*, 1949, [ii], **76**, 1592; *M.A.*, **17**, 908) adaptation and appn. of Wannier's treatment of the motion of an electron in a perturbed periodic potential, (*ibid.*, 1937, [ii], **52**, 191) is made more general by taking transitions between energy bands into account. Detailed consideration is given to the case arising from slowly varying combined elect. and magnetic fields. The treatment is developed to the stage where it appears likely to give a useful discussion of the results of perturbation of a periodic lattice by lattice vibrations. —P. C. L. P.

***Periodic Potential [Fields] in Metals.** Shin Chiba and Yasuo Shiota (*Sci. Rep. Tôhoku Univ.*, 1949, [i], **33**, (2), 64–69).—[In English]. Bethe's method of analysis (*Ann. Physik*, 1928, [iv], **87**, 55) is used to derive expressions for the Fourier coeff. of the periodic potential fields in crystals of Li, Na, Fe, and Cu. Appn. of the results for purposes of numerical calculation are illustrated in the case of the b.c.c. lattice of Fe, and numerical values of the coeff. for the five metals Li, Na, Fe, Ni, and Cu are tabulated.—J. S. G. T.

***[Electric] Potential Near a Crystal Surface.** Yasuo Siota (*Sci. Rep. Tôhoku Univ.*, 1951, [i], **35**, (1), 11–14).—[In English]. An expression is derived for the elect. potential (V) near the crystal surface of a f.c.c. crystal, on the assumption that any difference between the atomic arrangement and electron distribution at the surface layer and inside the crystal can be ignored. Numerical results for Cu and KCl are discussed. The mean value of V in the unit cell near the surface is smaller than that of unit cells sufficiently deep in the crystal interior.—J. S. G. T.

***Absolute Values of Bloch-Type Solutions in a Periodic Potential Field.** Shôzô Okada (*Sci. Rep. Tôhoku Univ.*, 1951, [i], **35**, (1), 15–20).—[In English]. Bloch's treatment of electronic states in crystal lattices considers the *eigen*-values of the electronic states as comprising permitted and forbidden bands. Expressions for the wave functions and statistical electron d in a one-dimensional lattice periodic potential field of rectangular type are derived and evaluated numerically.—J. S. G. T.

Electronic Structure of Diamond-Lattice-Type Crystals. Akira Morita (*Sci. Rep. Tôhoku Univ.*, 1949, [i], **33**, (2), 92–98).—[In English]. The electronic structure of diamond-lattice-type crystals, e.g. diamond and Si, is derived by appn. of the " tight-binding " approximation method. Electronic band structures in these substances calculated from the results are in better agreement with experimental values, derived from elect. conductivity and Hall-effect data, than are structures calculated by the Wigner–Seitz method. —J. S. G. T.

***The Spontaneous Magnetization of a Two-Dimensional Ising Model.** C. N. Yang (*Phys. Rev.*, 1952, [ii], **85**, (5), 808–815).—The variation of the spontaneous magnetization of a two-dimensional Ising model of a ferromagnet has been calculated as a function of temp. right up to the critical temp., by math. methods similar to those used by Kaufman and Onsager (*ibid.*, 1949, [ii], **76**, 1244; *M.A.*, **17**, 718) to calculate short-range order.—P. C. L. P.

Theory of Co-Operative Phenomena. Gérard Fournet (*Phys. Rev.*, 1952, [ii], **85**, (4), 692).—A letter. Comparison of the theory developed by Kikuchi (*ibid.*, 1951, [ii], **81**, 988; *M.A.*, **19**, 193) with that of Yvon (*Cahiers Phys.*, 1945, (28), 1; *M.A.*, **18**, 601), to which attention is drawn, reveals that the former gives the same results by a more complicated route.—P. C. L. P.

Imperfections in Nearly Perfect Crystals. (Shockley, *et al.*). See col. 445.

---

5 — POWDER METALLURGY

Sintered Aluminium with High Strength at Elevated Temperatures. R. Irmann (*Metallurgia*, 1952, **46**, (275), 125–133).—See *M.A.*, **20**, 107, 173.—F. M. L.

***Sintered [Cobalt–Chromium–Tungsten] Alloys for High-Temperature Service in Gas Turbines.** R. W. A. Buswell, W. R. Pitkin, and I. Jenkins (*Iron Steel Inst. Special Rep.*, 1952, (43), 258–268; discussion, 345–354; and *G.E.C. Journal*, 1951, **18**, (3), 139–156).—The development of a satisfactory sintered alloy of the Vitallium type (Co 64, Cr 30, W 6%), having low porosity and properties comparable to those of the cast alloy, is described. The influence of particle size, compacting pressure, time and temp. of sintering, and the sintering atmosphere on the d of the compact and its properties was investigated. The scaling-resistance of the best sintered products is high, and the room-temp. mech. properties are similar to those of the cast alloys, but at elevated temp. the creep-resistance is low. The addn. of small quantities (~0·1%) of ThO_2 results in an improvement in creep properties, and the effects of C addn. are being investigated. Sensitivity to ageing treatments was noted, and the significance of the transformation from a f.c.c. to a h.c.p. structure is discussed. 11 ref.—D. M. P.

***Bonding Investigation of Titanium Carbide with Various Elements.** Walter J. Engel ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2187), 15 pp.; *C. Abs.*, 1951, 45, 1931).—Cupped TiC bodies were fabricated from material having a particle size of approx. 5 μ. Al, Be, Cr, Co, Nb, Au, Fe, Pb, Mg, Mn, Ni, Pt, Si, Ti, and V were selected for use as trial binders. A graphite elect.-resistance, muffle-type furnace, employing a He atmosphere, was used for fusing the powdered elements in the carbide cups. Metallographic studies of the fused-cup structure indicated that Ni and Co appeared to produce extensive interlocking bonds, with Ni showing the greater penetration into the TiC. The observed interfaces provided information concerning some of the reactions involved in bonding the elements to solid TiC. Cr and Si adhered to TiC after cooling below the m.p. The amount of penetration, however, suggests that neither element surrounds the TiC particles so thoroughly as Ni and Co.

***Research on the Sintering of Doped Tungsten Powder.** Chikao Ushioda, Gen-ichi Nakazawa, and Makoto Terui (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, 16, (1), 63–66).—[In Japanese, with tables and diagrams in English]. Two samples of W powder with different particle-size distribution were prepared from WO_3 contg. SiO_2 0·4, K_2O 0·2, and Al_2O_3 0·03%, and the compacts were sintered using currents equivalent to 65, 75, 80, 85, 88, 90, 92, and 95% of the fusion current. The changes in shrinkage, hardness, apparent *d*, elect. resistivity, grain-size, and decrease of non-volatile residue and FeO during sintering were studied. Grain-growth measurements showed that (1) the grain-size of pure W compacts sintered at 90% current was ~100 times as large as that of doped compacts and (2) the grain-growth was rapid between 65 and 80% current, but it could be controlled by doping. The impurity content (FeO) was not affected to any extent by doping. The residue contained ~50% SiO_2 and 40% Al_2O_3.—Authors.

***The Sintering Mechanism Between Zirconium Carbide and Niobium.** H. J. Hamjian and W. G. Lidman ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2198), 29 pp.; *C. Abs.*, 1951, 45, 2381).—ZrC and Nb specimens were prepared by hot pressing at 2000 lb./in.^{2}. Sintering temp. of 3700°, 3900°, and 4050° F. (2040°, 2150°, and 2230° C.) for 5 min. and 3900° F. (2150° C.) for 15, 30, 45, and 90 min. were used to determine the effects on structure. A bonding study was also made of these specimens by X-ray and microscopic methods. Density and modulus-of-rupture tests were carried out to determine the phys. properties of the compacts. The results indicated that the sintering mechanism is one in which the Nb atoms diffuse into the ZrC lattice, displace Zr atoms, and form NbC and Zr metal. The NbC is completely soluble in the matrix of ZrC, forming a homogeneous solid soln. of the carbides. At a sintering temp. of 3900° F., the Zr metal forms in the grain corners of the carbide structure. Modulus-of-rupture evaluation showed that the specimen sintered at 3900° F. was stronger than the one sintered at 4050° F., indicating that lower strength corresponds with larger coalesced metal inclusions.

Some Observations on the Role of the Binder in Cemented Refractory Alloys. J. T. Norton (*Powder Met. Bull.*, 1951, 6, (2), 75–78).—The criteria for a satisfactory binder material to be used with refractory substances are discussed.—W. A. M. P.

Sweat Cooling: A Review of Present Knowledge and Its Application to the Gas Turbine. P. Grootenhuis and N. P. W. Moore (*Iron Steel Inst. Special Rep.*, 1952, (43), 281–288; discussion, 345–354).—The mechanism of sweat-cooling is discussed, and the suitability of sintered powder products is noted. The known mech. properties of porous materials are reviewed, and designs for such components as turbine blades are suggested. 47 ref.—D. M. P.

Report on Powder Metallurgy. Thomas A. Dickinson (*Steel Processing*, 1950, 36, (12), 619–621).—A brief review of the present appn. of powder metallurgy and the continued growth of the powder-met. industry.—S. R. W.

Sintered Carbides: New Tool of Ceramics. John W. Graham and W. L. Kennicott (*Ceram. Ind.*, 1950, 55, (6), 93, 94, 96).—A review of the sintered carbides WC and TiC and their appn. in the ceramic industry.—N. B. V.

6 — CORROSION AND RELATED PHENOMENA

***Corrosion of Aluminium by Carbon Tetrachloride.** Milton Stern and Herbert H. Uhlig (*J. Electrochem. Soc.*, 1952, 99, (10), 381–388).—When Al (99·99%) is exposed to boiling CCl_4 an induction period exists before any reaction occurs. This period is lengthened when there is present in the CCl_4 dissolved H_2O, CS_2, or O, and is shortened by the presence of corrosion products. After the induction period, the corrosion rate is very high, ~20 in./year in relatively dry CCl_4 and twice as fast in H_2O-saturated CCl_4. The presence of O, N, or corrosion products has no effect on this corrosion rate, but HCl increases it 35%. The induction period for Al alloys may be ~30 times as long as for pure Al and the corrosion rate is usually lower for alloys than for the pure metal. It is shown that the main reaction is $6CCl_4 + 2Al = 2AlCl_3 + 3C_2Cl_6$, although in addn. to $AlCl_3$ there is produced a tarry residue contg. highly chlorinated hydrocarbons and probably Al complexes. The electrochem. nature of the corrosion is suggested by galvanic-couple tests, applied potential tests, and the variable effects of Al impurities on the induction period and corrosion rate. 14 ref.—G. T. C.

***Effect of Oxide Films on the Reaction of Aluminium with Carbon Tetrachloride.** Milton Stern and Herbert H. Uhlig (*J. Electrochem. Soc.*, 1952, 99, (10), 389–392).—The fact that the induction period for the reaction of Al with CCl_4 is associated with the protective film on the metal (see preceding abstract) led to attempts to improve this film by anodizing. This resulted in some extension of the induction period, but not to an extent which was of practical significance. This was the case whether the oxide coatings were produced in NH_4 tartrate or in H_2SO_4 with oxide thicknesses of several thousand Å. Commercially produced coatings also broke down at a few spots from which rapid corrosion progressed radially. Heating the Al for several hr. at 235° C. only doubled the induction period. On the other hand, rather better results were obtained with the natural film formed on evaporated Al, but even these films could be damaged by scratching either before or during exposure to CCl_4. It is suggested that the superior resistance to CCl_4 of certain alloys is not related to the oxide film but to a protective film of another kind.—G. T. C.

***The Effect of Cold Work on the Microstructure and Corrosion-Resistance of Aluminium–5% Magnesium Alloys Containing 0–1% Zinc.** P. Brenner and G. J. Metcalfe (*J. Inst. Metals*, 1952–53, 81, (5), 261–268).—The effects of cold work, of prolonged ageing at 70° C., and of addn. of up to 1% Zn on the microstructure, corrosion-resistance, and stress-corrosion-resistance of an Al alloy contg. 5% Mg have been investigated. The presence of Zn has little effect on the mech. properties, but the alloy contg. 1% Zn shows pronounced pptn. after ageing at 70° C., and corrosion attack is greater than in the case of the 0–0·5% Zn alloys, which show an incomplete grain-boundary network after ageing. Small amounts of cold work reduce the corrosion-resistance of the alloys appreciably, but heavy cold work both of aged and unaged materials results in a corrosion-resistance as high as that of unworked material. The max. corrosion attack, at ~30% cold work, is associated with the presence of continuous grain-boundary films formed on ageing. Comparable behaviour has been found in stress-corrosion tests, the max. susceptibility to stress-corrosion occurring in material that has been given 30% cold work. Heavy cold work (50% reduction in thickness) produces material of high stress-corrosion resistance. Explanations of the corrosion behaviour are given.—Authors.

***Stress-Corrosion: Results of Researches on Some Aluminium Alloys.** E. Hugony (*Alluminio*, 1948, **17**, (5/6), 225–247).—Experiments carried out with the Brenner apparatus consisted of immersing the samples in a 3% NaCl soln. with the addn. of 1% HCl and of subjecting them to a stress equal to $\frac{2}{3}$ of the U.T.S. It was found that Duralumin correctly soln.-treated and age-hardened at room temp. did not show the intercryst. corrosion which was found after age-hardening at elevated temp. Zn as an addn. to alloys of the type Al–3·5% Mg produced intercryst. corrosion and failure of the samples in only 5 days (normal duration of the test without addn. of Zn was 263 days). The addn. of Cu to alloys of the type Al–Si–Mg–Cr (61S) has been found to be harmful. Stress-corrosion in wires of the same type of alloys was such as to prevent their use for sp. purposes. Corrosion in wires of Al–5% Mg alloy was found to be not intercryst. but uniformly distributed without even the presence of pits.—F. E. M.

***[Corrosion] Studies on Some Light [Aluminium] Alloys.** Mario Raffo (*Ricerca Sci.*, 1951, **21**, (8), 1375–1383).—Marine corrosion tests were conducted on six Italian Al alloys contg. varying percentages of Cu, Mn, Mg, Si, and Fe. Continuous and intermittent immersion as well as exposure to saline atmosphere were employed. Mech. tests and X-ray studies were then made of the alloys tested.—I. S. M.

Aluminium Alloy Dock Pilings Prove Immune to Salt-Water. Gilbert S. Schaller (*Western Metals*, 1952, **10**, (6), 53).—61S-T6 Al-alloy beams subjected to the const. action of sea-water for more than two years were found to show no evidence of corrosion.—T. G.

***New Method for the Coloriscopic Detection of Corrosion on Aluminium Surfaces.** Gyula Bácskai and Klára Kovacs (*Magyar Kém. Lapja*, 1949, **4**, 532–535; *C. Abs.*, 1951, **45**, 2840).—The colour reaction of Al with hematoxylin serves as a basis for the method described. A 10% aq. gelatin soln. is prepared, 1% NH_4Cl is added as an electrolyte and stabilizer, and then a small portion of the soln. is treated with H_2O_2. After oxidation, this portion is mixed with the unoxidized gelatin soln., and 1 c.c. of a 10% soln. of hematoxylin in ethyl alcohol is added. This reagent turns brown when treated on a water bath. The Al surface to be examined is carefully freed from traces of grease and covered with a thin film of the gelatin soln. with a brush or a spray. The corroded spots become deep violet in colour in 0·5–3·0 min. This is clearly visible through the brownish gelatin film. The final result can be observed 10 min. after applying the gelatin film. Experiments proved that this method is more sensitive than that of Goldowski (*Korrosion u. Metallschutz*, 1937, **13**, 128; *M.A.*, **4**, 637). Some anti-corrosion agents may form coloured compounds with hematoxylin. Al surfaces previously treated with such agents should be tested before applying this method.

***A Preliminary Study of the Oxidation and Vapour Pressure of Chromium.** E. A. Gulbransen and K. F. Andrew (*J. Electrochem. Soc.*, 1952, **99**, (10), 402–406).—The high-vacuum behaviour of Cr at 600°–1015° C. was studied. Three reactions were noted. First, Cr acts as a getter to O, forming an oxide film. Second, the surface oxide and internal oxides can react with C in the metal to form CO at temp. >800° C. Third, Cr vaporizes at an appreciable rate at temp. >825° C. The rate of vaporization is not affected by the presence of oxide or nitride films in the temp. and thickness ranges investigated. Cr reacts with O at an appreciable rate above 700° C., and the parabolic-rate law gives a reasonable correlation. The heat of activation is calculated as 66,300 cal./mole and the entropy of activation as 10·7–13·5 entropy units. The vapour pressure was determined between 885° and 1015° C. and a value of $93·9 \pm 0·2$ kg.cal./mole was found for ΔH_0^0. A comparison of the rate of oxidation was made with other metals. Although Cr is superior to many metals, it is inferior to Be and to alloys of the Ni–Cr series. 22 ref.—G. T. C.

***Structure and Oxidation Kinetics of Cathode Copper.** N. A. Marchenko and A. N. Sysoev (*Zhur. Priklad. Khim.*, 1950, **23**, (5), 493–495).—[In Russian]. 0·8 cm. of Cu was deposited from a bath contg. $CuSO_4.5H_2O$ 200, H_2SO_4 75 g./l., at 25° C., and at c.d. of 1, 2·5, 5, and 8 amp./dm.^{2} on cathodes of rolled Cu which had been mech. polished, then amalgamated in 0·75% $HgCl_2$ soln. X-ray examination of the various deposits showed that those obtained at 1 amp./dm.^{2} had a distinct [100] orientation (that of the basis Cu). The degree of orientation was less for deposits at 2·5 amp./dm.^{2}, but a max. at 5 amp./dm.^{2}. Deposits at 8 amp./dm.^{2} were almost completely disorientated. The deposits were removed from the cathode, degreased with French chalk, and cathodically pickled in 0·1*N*-KCl at 1 amp./dm.^{2}. They were then oxidized at 1000° C. in an elect. tube furnace, while hanging by a Pt wire from a balance. Parabolic gain-in-wt./time curves were obtained, but the rates of reaction varied with the c.d. at which the deposits were prepared, as follows: $5 > 2·5 > 1 > 8$ amp./dm.^{2} > basis Cu, i.e. the rate of oxidation increases with increasing degree of orientation.—G. V. E. T.

***Thermo-Gravimetric Study of Metallic Oxidation (Copper).** F. de Carli and N. Collari (*Chim. e Ind.*, 1951, **33**, (2), 77–80).—A thermo-grav. re-examination of the laws of the oxidation of Cu, using a continuous photographic recording apparatus, has enabled some controversial points to be clarified. Temp. range of the examination was 400°–900° C. At 400° C. the oxidation increase follows a law of the type $1/P = A - B \ln t$, where P = the weight in g. of O/unit surface (= the thickness of the oxide layer), t = the time in min., and A and B = const.; between 500° and 900° C. it follows the parabolic rule $P^2 = Kt$. The discontinuity, noted in oxidation in dry-air streams at atm. pressure, is attributed to the disappearance of CuO in the oxidized zone by the reaction $CuO + Cu \rightleftharpoons Cu_2O + 3·5$ kg.cal. rather than by $2CuO \rightleftharpoons Cu_2O + \frac{1}{2}O_2 - 34$ kg.cal. 9 ref.—I. S. M.

***Study of Some Electrochemical Phenomena by Using an Electro-Gravimetric Apparatus. I.—Copper.** Paolo Spinedi (*Chim. e Ind.*, 1951, **33**, (12), 777–782).—An apparatus is described which allows the photographic recording of weight variations occurring in a metallic test-piece while undergoing anodic dissoln. by the action of const. or variable e.m.f., according to the potentiometric system. Some experimental results are examined which were obtained with this apparatus on Cu test-pieces subjected to anodic dissoln. by the action of increasing e.m.f. and with varying c.d. 10 ref.—I. S. M.

***Electron-Diffraction Studies on the Oxidation of Pure Copper and Pure Zinc Between 200° and 500° C.** E. A. Gulbransen and W. R. McMillan (*J. Electrochem. Soc.*, 1952, **99**, (10), 393–401).—Studies on pure Cu, electrodeposited Cu, and evaporated Cu show that the extra reflections ascribed by some workers to an oxide CuO′, in addn. to CuO and Cu_2O, are in fact due to an oxide of Pb or of Zn. Pb and Zn, although present only as traces, concentrate at the surface layer during oxidation or else are picked up from the atmosphere. It is shown that Cu_2O is the main oxide in the surface layer, although CuO is formed at 400° and 500° C. for longer periods of oxidation. A study of the oxidation of super-pure Zn and of evaporated Zn over the temp. range 200°–375° C. shows normal ZnO to be the only oxide formed. No evidence was found of an intermediate pseudomorphic oxide previously reported in the literature. 37 ref.
—G. T. C.

***High-Temperature Oxidation Characteristics of a Group of Oxidation-Resistant Copper-Base Alloys.** J. P. Dennison and A. Preece (*J. Inst. Metals*, 1952–53, **81**, (5), 229–234).—The influence of small separate addn. of Al, Be, Cr, Mg, and Si on the high-temp. oxidation of Cu has been examined, these alloying elements having been chosen because their oxides have a high elect. resistivity. The compn. and microstructure of the scales formed were studied and correlated with the rates of oxidation. The effectiveness of the addn. in conferring oxidation-resistance was in the decreasing order: Be, Al, Mg, Si, Cr. Alloys having rates of oxidation less than that of Cu showed divergences from the established parabolic and exponential laws. These divergences may be accounted for by a consideration of the relative rates of diffusion of Cu and of the alloying element through the oxide layers.—AUTHORS.

Newly Developed Bronze Alloys Feature Anti-Galling and Corrosion-Resistance. John P. Meyer (*Western Metals*, 1952, **10**, (3), 35–36).—M. describes in general terms the development and properties of a high-Pb bronze contg. Ni, Zn, and Sn. No compn. is given.—T. G.

***A Polarographic Study of the Influence of Temperature on the Rate of Oxygen Consumption by Iron, Lead, and Zinc.** Paul Delahay, Clarence F. Pillon, Jr., and Douglas Perry (*J. Electrochem. Soc.*, 1952, **99**, (10), 414–416).—Rates of O consumption by Fe, Pb, and Zn in an acetate buffer of $pH = 5{\cdot}0$ were determined at various temp. On the assumption that the overall rate of O reduction is controlled by the rate of the chem. reaction and by diffusion of O towards the metal surface, experimental data were analysed, and it was found that diffusion rate is the predominant factor controlling the overall rate. The influence of the rate of the chem. process was, however, by no means negligible, as indicated by the fact that the relative increase in the overall rate with temp. varies with the metal. This is to be expected because the rate of O consumption is partially controlled by two factors which vary with the metal, namely, the overvoltage of O reduction and the overvoltage, if any, for the anodic oxidation of the metal. 11 ref.—G. T. C.

***Scaling of Heat-Resisting Steels: Influence of Combustible Sulphur and Oil-Fuel Ash Constituents.** C. Sykes and H. T. Shirley (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 153–169; discussion, 355–367).—A comprehensive investigation has been made of the scaling behaviour of 15 heat- and creep-resisting steels under varying conditions. The samples were normally subjected to seven 6-hr. heating periods at the max. working temp. of the material concerned under an atmosphere produced mainly by the combustion of town's gas. The samples were cooled to room temp. between the heating periods and brushed to remove loose scale. The gain in weight at each stage due to attack was noted. The influence of the following factors on the rate of scaling and their interplay was examined: (1) air/gas ratio; (2) combustible S introduced as H_2S into the town's gas; (3) ash constituents present in the fuels normally used in service, including Na_2SO_4, NaCl, and V_2O_5; these tests were carried out by part immersion of the samples in the ash and by subjecting them either to the normal cycling under combustion products or to heating in static air for periods up to 640 hr.; and (4) MoO_3 which may be formed in service from Mo in the steels and which accelerates attack. 9 ref.—D. M. P.

***The Corrosion-Resistance of Electrodeposited Tin-Nickel Alloy.** S. C. Britton and R. M. Angles (*J. Electrodepositors' Tech. Soc.*, 1951, **27**, 293–319; discussion, 320–322; and (abridged) *Electroplating*, 1951, **4**, (6), 201–203; *Metal Ind.*, 1951, **79**, (3), 46–50).—Corrosion tests are described on electrodeposits of Sn–Ni alloy (65% Sn) using exposure to natural and simulated atmospheres for deposits up to 0·001 in. thick and using a variety of agents for thick, pore-free coatings. At 30° C., it is shown that the alloy is unaffected by alkalies, neutral soln., HNO_3, and soln. of other acids with a $pH > 1{\cdot}2$. It is not tarnished in atmospheres grossly polluted with SO_2 and H_2S, but on prolonged exposure to certain aq. soln. tarnishing may occur. There is evidence that a protective film is formed on the alloy and that this is easily replaced if damaged. Corrosion in crevices is not serious. The alloy is cathodic to steel and brass. A coating 0·0005 in. thick keeps corrosion at pores to within reasonable limits on brass for moderately severe exposure. On steel, an undercoating of 0·0005 in. of Cu is desirable for outdoor service. 10 ref.—G. T. C.

***High-Temperature Corrosion Rates of Several Metals with Hydrogen Sulphide and Sulphur Dioxide.** Milton Farber and Donald M. Ehrenberg (*J. Electrochem. Soc.*, 1952, **99**, (10), 427–434).—The corrosion rates of Cu, Ag, Inconel, Ni, 18 : 8 stainless steel, Fe, W, Mo, and Ta were determined at high temp. in H_2S, SO_2, and CO. This was done by measuring the increase in elect. resistance of the metals in the form of filaments of 0·010 in. dia. The metals most easily corroded in H_2S were Inconel, Fe, Cu, and stainless steel, whereas W, Mo, and Ta had the best corrosion-resistance. Inconel was very resistant to SO_2, but W, Ta, and Mo were easily corroded. CO increased the corrosion rate of Ni in H_2S, but that of Fe was decreased by the presence of CO. 22 ref.—G. T. C.

***Measurement of the Corrosion Rate of a Metal from Its Polarizing Characteristics.** W. J. Schwerdtfeger and O. N. McDorman (*J. Electrochem. Soc.*, 1952, **99**, (10), 407–413).—The theoretical elect. relationships between the polarizing characteristics of a galvanic couple and the polarizing characteristics of the couple itself are described and demonstrated experimentally. An expression for the measurement of corrosion rate is shown to apply generally to the corrosion of steel in soils, and presumably in other aq. media. This expression has been confirmed only for the special case of single couples. From the relationship between the potential of a corroding metal and the applied external current, the type of control of the corrosion rate was deduced. The significance of the slope of polarization curves with respect to corrosion rate is indicated. 15 ref.—G. T. C.

***The Electrochemical Nature of the Corrosion of Metals in Liquid Dielectrics.** L. G. Gindin (*Doklady Akad. Nauk S.S.S.R.*, 1950, **73**, (3), 515–518; *C. Abs.*, 1951, **45**, 524).—[In Russian]. In a $1{\cdot}5N$-soln. of CH_3COOH in iso-octane, of sp. elect. conductivity $\sigma = 3{\cdot}3 \times 10^{-14}$ ohm^{-1} cm.$^{-1}$ when subjected to 30 kV./cm., a Mg/Fe couple showed acetate spots on the Fe spreading from its middle, but stopping at a distance of ~0·2 mm. from the Fe/Mg boundary. The Mg was corroded particularly along the boundary with Fe. This corrosion pattern, wherein the periphery of the Fe was protected by the contact with Mg, indicated clearly its electrochem. nature. This was confirmed by the potential distribution, mapped under a microscope, which showed a steep fall from the Mg to the Fe potential within a narrow boundary strip ~0·5 mm. wide. Consequently, in the dielect. soln., Fe and Mg form a cell in which Mg is anodic to Fe. That these electrochem. boundary effects, observed with macrocouples, must be much more intense with microcouples, follows from the observation that the zone of marked corrosion of Mg spreads and increases with decreasing dimensions of the electrodes. In a Fe/Mg microcouple of a dia. of 10^4 Å., with a gap of ~10 Å. between the 2 metals, the elect. field E in the gap is ~10^7 V./cm., and, on the surface of the microcouple, 10^{-3} to 10^{-4} V./cm. On account of the proportionality between log σ and $\sqrt{E}$, this field gives rise to a sufficiently high elect. conductivity around the microcouple to set up a galvanic cell.

***Electrochemical Analysis of the Corrosion-Inhibiting Action of Emulsive Oils.** Antonio Ferri (*Chim. e Ind.*, 1951, **33**, (8), 477–479).—F. has studied the nature and the mechanism of the corrosion-inhibiting action of emulsive oils used in the working of metals, and found that their action, being only anodic, is liable to cause localized corrosion. A possibility of combining the anodic action of the oils with the cathodic one of other inhibiting agents is found in $K_2Cr_2O_7$, which, together with oil, forms an "ideal" inhibitor, combining the efficiency of anodic with the security of cathodic inhibitors. 7 ref.—I. S. M.

***Inorganic Corrosion Inhibitors in Acid Solutions.** Cecil V. King, Emil Goldschmidt, and Natalie Mayer (*J. Electrochem. Soc.*, 1952, **99**, (10), 423–426).—The rate of dissoln. of Fe, Zn, and Cu in a dil. HCl soln. with excess of KNO_3 as a depolarizer, is at, or near, a max. controlled by the rate of convection and the speed of diffusion of the H ion. Addn. of dichromate, molybdate, or tungstate reduces the rate of dissoln. considerably. In the case of very pure Fe and Zn, the rate in the presence of dichromate can be reduced still further by addn. of a soluble fluoride. The experiments described are an aid to interpreting the mechanism of inhibition, though none of the soln. described are in fact highly protective. 14 ref.—G. T. C.

Scaling of Gas-Turbine Alloys. A. Preece (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 149–152; discussion, 355–367).—

A discussion of the factors controlling the rate of formation and the breakdown of oxide films. The alloy and fuel compn. and the effects of free O are considered. 24 ref.—D. M. P.

***Theoretical Analysis of the Diffusion Processes Determining the Oxidation Rate of Alloys.** Carl Wagner (*J. Electrochem. Soc.*, 1952, 99, (10), 369–380).—A theoretical analysis is made for specified ideal conditions of the interplay of diffusion processes in metallic and oxide phases during the oxidation of alloys. The oxidation rate of alloys contg. a noble metal (Au or Pt) and an oxidizable metal (Ni, Cu, or Zn) is calculated as a function of the alloy compn. It is shown that the oxidation rate of Ni–Pt alloys at 850° and 1100° C. is essentially determined by the diffusion of Ni to the alloy/NiO interface if the mole fraction of Ni is $<0{\cdot}5$. Oxidation rates observed are in accordance with values calculated from diffusion data. If both constituents are oxidizable, the alloy may form a one-phase or a two-phase scale. A necessary condition for the formation of a scale consisting of one oxide only is stated, and it is shown that the results for Cu–Ni and Cu–Zn alloys are essentially in accordance with theory. It is illustrated by observations on the Cu–Al system that two oxides can nucleate initially and continue to grow even under conditions under which exclusive formation of oxide of the less noble metal is a possible process. 47 ref.—G. T. C.

Corrosion Factors in Design. F. M. Reinhart (*Product Eng.*, 1951, 22, (7), 101–107, 158–159).—Factors which influence the rate of corrosion are outlined, and methods of minimizing attack by atmospheric, under-water, galvanic, concentration-cell, stress, and fretting corrosion are described.—M. A. H.

Corrosion of Metals: An Historical Treatment of the Subject. U. R. Evans (*Chem. and Ind.*, 1952, (41), 986–993).—Corrosion of ferrous and non-ferrous metals is discussed with special ref. to early and comparatively early knowledge and studies of its mechanism and prevention. 26 ref.—J. R.

Developments in Corrosion During the Past Fifty Years: A Brief Review. Herbert H. Uhlig (*J. Electrochem. Soc.*, 1952, 99, (10), 275C–276C).—Outstanding contributions to the study of corrosion in the past fifty years are described.—G. T. C.

---

7 — PROTECTION

(Other than by Electrodeposition)

***Protection of Aluminium and Its Alloys Against Corrosion by [Chemical] Treatment with Various Aqueous Acid Solutions.** Hikozô Endô and Hideo Sugawara (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, 16, (3), 153–157).—[In Japanese]. The corrosion-resistance of protective films formed on specimens of Al, Duralumin, Super-Duralumin, Silumin, Lautal, and KS-Seewasser by treatment at room temp. for 10–15 min. with various aq. salt soln. (chloride, sulphate, nitrate, phosphate, &c.) contg. Na_2SiO_3, $ZnCO_3$, Na_2SiF_6, &c., were compared with films formed by 3% NaCl soln. Experiments showed that aq. $Zn(H_2PO_4)_2$ soln. contg. $ZnCO_3$ and Na_2SiF_6 gave the best protective film.—Authors.

***Amorphous and Crystalline Layers of Aluminium Oxide on Aluminium.** A. J. Dekker and W. Ch. Van Geel (*Alluminio*, 1948, 17, (7/8), 351–355).—99·99% pure Al was electrolytically coated with an amorphous Al_2O_3 layer in an oxalic acid bath. After heating to 450° C. to drive off the remaining acid, the sample was immersed in a boric acid bath which formed cryst. Al_2O_3. It was found that the presence of the amorphous layer increased the effect of the c.d. fourfold compared with an electrolytic treatment of the specimen in boric acid alone. This showed that the original theory that the cryst. layer is formed underneath the amorphous layer is incorrect and that instead the cryst. layer is formed in the pores of the amorphous layer. These pores constituted ~25% of the amorphous layer. In the experiments a relation between the c.d. and the porosity of the amorphous layer formed in the oxalic acid bath was found, i.e. the greater the c.d., the more compact the oxide layer.—F. E. M.

***On the Time of Sealing and the Possibility of Improving the Corrosion-Resistance of Anodic Oxide Films on Aluminium.** G. Bolognesi (*Alluminio*, 1948, 17, (11/12), 572–575).—B. has studied the corrosion-resistance of anodic oxide films on Al as a function of the time of oxidation and of the time of sealing in boiling water. Pure Al samples (99·5%) of 9 cm.^{2} effective area were anodized in a 15% H_2SO_4 bath under a potential of 12 V. for times varying from 0 to 60 min. They were then immersed in a soln. of *N*-HCl for 8 hr. at 25° ± 1° C. The H evolved in this bath was a measure of the attack by HCl and consequently of the pores in the oxide film. The min. time for effective anodizing was found to be 20 min. B. carried out experiments by immersing samples anodized for 20 min. in a 5% soln. of hexamethylenetetramine for 24 hr. Sealing was effected by boiling in this soln. for 10 min. In weak HCl soln. this last treatment gave better results than sealing with boiling water.—F. E. M.

***The Protective Properties of Chromate-Base Paints on Light Alloys.** Corrado Bighi and Giorgio Mantovani (*Chim. e Ind.*, 1951, 33, (6), 355–359).—B. and M. have used the A.R.D. sea-water spray test described by Hudson and Banfield, (*J. Iron Steel Inst.*, 1946, 154, 229P; *M.A.*, 16, 287), as well as their own electrochem. immersion method, to determine the protective power of paints based on various types of chromate. The use of such paints on Al is, on the whole, shown to be beneficial, especially in neutral atmospheres. 11 ref.—I. S. M.

Aluminium Paints. G. Salomone (*Ind. Vernice*, 1948, 2, 125–127; *C. Abs.*, 1950, 44, 9159).—The loss in brightness and decrease in protective properties which occur in Al paint films on ageing appear to be due to a re-orientation of the Al flakes so that a continuous metallic surface is no longer present. Pretreatment of the Al powder with aq. Na_2CrO_4 before its incorporation in the medium increases the period during which the protective structure is maintained, but does not prevent the loss in brightness.

How to Make Brass Liners Work. William C. Uhl (*Petroleum Processing*, 1952, 7, (4), 475–477).—Practical hints are given on the use of brass liners for the protection of pipe-lines and vessels carrying gasoline liquid and vapour.—G. T. C.

***On the Protection of Thin Deposits of Molybdenum by a Layer of SiO Obtained by Vacuum Evaporation.** Charles Feldman (*Compt. rend.*, 1952, 235, (14), 706—707).—Elect.-conductivity measurements were made to determine to what extent a dielect. film (in this case SiO) afforded protection to a metal (Mo) against the action of the atmosphere (cf. *ibid.*, 1952, 234, 1858; *M.A.*, 20, 388). A deposit $>1\ \mu$ thick afforded reasonable protection. Protected areas of the metal were exposed to a vacuum, to air at room temp., and to air at the temp. of liq. N. *In vacuo* ageing is small—smaller than without a protecting film—probably owing to a modification in the structure of the film. At liq.-N temp. there is at first a somewhat rapid variation of the elect. resistance, then a slow diminution; but when the room temp. is reached, the resistance becomes exactly what it was before the specimen was cooled. Hence it appears that at low temp. there is no oxidation of the film. An advantage of this method of study is that the rate of oxidation can be controlled by altering the thickness of the protecting film. —J. H. W.

Testing the Quality of Tinplate for Canning. W. Czarnek (*Przemysl Rolny i Spozywczy*, 1950, 4, 55–59; *C. Abs.*, 1951, 45, 3786).—Methods for the detn. of the porosity and thickness of Sn coating, and of the corrosion-resistance of the steel base are reviewed. Several tinplates were tested.

Automatically Controlled Galvanizing Kettles. —— (*Metal Ind.*, 1952, 81, (18), 349–350).—A pneumatic and an elect. automatic control system for galvanizing are described, the type used depending on the working conditions and the accuracy of control desired. For small and medium-sized galvanizing plants, the pneumatic system may be preferable for batch processing, but an elect. system can be used for both batch and const. galvanizing methods.—J. H. W.

Zinc Protects Trailer Parts from Corrosion. W. G. Patton (*Iron Age*, 1952, **169**, (22), 82–84).—Zincilate 300, applied by brush, spray, or dip, as a protection of steel parts from corrosion is more expensive than prime paint but cheaper than Cd. —J. H. W.

Applications of Corrosion-Resistant Coatings. C. M. Jelcot (*Organic Finishing*, 1950, **11**, (6), 14–17; *C. Abs.*, 1950, **44**, 9163).—A review is given of the various corrosion-resistant organic finishes and their methods of appn.

Cathodic Protection of Buried Pipelines.—I.–II. K. A. Spencer (*Petroleum*, 1952, **15**, (5), 120–123; (6), 149–153).—An elementary introduction to the theory and practice of cathodic protection. The use of Mg, Al, and Zn anodes is mentioned. 13 ref.—G. T. C.

Checking the Cathodic Protective System. William E. Huddleston (*Petroleum Eng.*, 1952, **24**, (5), D67–D68).—The need for making regular examinations of cathodic-protection systems is emphasized.—G. T. C.

Current Output of Galvo-Pak Magnesium Anodes as a Function of Anode Size and Soil Resistivity. T. H. Gilbert (*Gas*, 1950, **26**, (8), 103, 106; *C. Abs.*, 1950, **44**, 8798).—The standard Galvo-Pak Mg electrode is approx. 8 in. in dia. and 26 in. long. Formulæ are given which make it possible to estimate the length of bare or poorly insulated steel pipe of various sizes that can be protected by a single Mg anode; these lengths are also given in graphic charts for various soil resistivities.

Metallizing for Corrosion Prevention. Gilbert C. Close (*Products Finishing*, 1952, **16**, (7), 58–65).—A general review of the appn. of sprayed metal coatings and of their properties. —G. T. C.

Metal Spraying. E. A. Williams (*Min. J.*, 1949, **232**, (5934), 352–354).—Brief descriptions of metal spraying by the Schori metallizing process (powder technique) and the wire process are given. Appn. to building-up and decorative purposes, and particularly the use of Zn and Al coatings, are mentioned.—S. R. W.

Metal Surfacing for Original Parts. —— (*Product Eng.*, 1951, **22**, (7), 122—124).—A brief outline of the characteristics of welded, sprayed, and evaporated metallic coatings, with indications of their uses, is given.—M. A. H.

An Analysis of Metal Coatings. R. A. Schaus (*Western Metals*, 1952, **10**, (6), 56–59).—A description of the various processes available.—T. G.

Protective Coatings. G. E. Seidel (*Steel Processing*, 1949, **35**, (7), 368–370).—Successful coatings involve the choice of the right coating, proper surface prepn., and correct appn. These are briefly discussed.—S. R. W.

High-Temperature Ceramic Coatings. —— (*Product Eng.*, 1951, **22**, (11), 177–179).—The value of various ceramic coatings in prolonging the life of components operating at high temp. is assessed.—M. A. H.

Ceramic-Coating Saves Scarce Metals and Lengthens Life of High-Temperature Parts. Hugh Penton (*Western Metals*, 1952, **10**, (4), 37–39).—T. G.

Nylon-Coated Wire. John S. Trevor (*Textile Recorder*, 1952, **70**, (832), 88).—Nylon (dielect. strength 1500 V./mil) is a very suitable jacketing material for elect. cables exposed to highly abrasive conditions or to attack by certain chem. solvents; it is also resistant to relatively high temp. The nylon covering can be applied from a cold dispersion giving coatings ~½ mil thick, or by a hot-extrusion process (mainly used for large wires and cables) which gives coating thicknesses of 0·010–0·060 in.—J. R.

8 — ELECTRODEPOSITION

The Plating of Aluminium Articles as a Production Process. [The Vogt Process.] A. W. Wallbank (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (7), 9 pp.; and *Plating*, 1952, **39**, (7), 755–758, 764).—Methods of plating Al are briefly reviewed, particular attention being paid to the Vogt process, of which full details are included. The sequence used is: electrolytic cleaning, acid etch, second electrolytic cleaning, Zn plating, brass plating, and Ni plating. It is stated that most Al alloys can be plated by this process, the most difficult being those with $>3\%$ Mg; these can, however, be plated if care is exercised in their cleaning. Possible future developments in the plating of Al are discussed. 11 ref.—G. T. C.

Fitting Hard-Chromium Plating to the Job. Gilbert C. Close (*Steel Processing*, 1950, **36**, (6), 281–284).—The phys. and chem. properties of electrodeposited Cr and their value in industry are described.—S. R. W.

Copper Striking on Zinc-Base Die-Castings. Irvin K. Hauseman (*Plating*, 1952, **39**, (10), 1125–1126, 1131; and (abridged) *Metal Ind.*, 1952, **81**, (17), 330–333).—Difficulties encountered in the plating of Zn-base die-castings are described briefly, together with methods developed for overcoming them. Among the defects described are dullness on Cu striking, brassy Cu deposit, rough Cu deposit, pitting caused by striking, blistering caused by striking, and casting pattern streaking shown up by striking.—G. T. C.

Increasing Copper Content in White Brass and Copper Cyanide Solutions by Electrolytic Regeneration. Walter R. Binai (*Plating*, 1952, **39**, (10), 1120–1121, 1131).—An electrolytic regeneration cell is described which enables the Cu content of Cu cyanide and white brass plating baths to be increased intermittently or continuously. In alloy plating, one or more metals may be introduced by means of electrolytic regeneration. Such addn. may be controlled to equal the rate of deposition. Replenishment of Cu by this method, as distinct from Cu addn., prevents cyanogen accumulation, cyanogen polymerization, carbonate increase, organic contamination, and addn.-agent breakdown. —G. T. C.

***The Crystal Structure of Electrodeposited Silver.** (Layton). See col. 409.

***The Electrodeposition of Thallium: Preliminary Note.** E. Bertorelle and L. Giuffrè (*Chim. e Ind.*, 1949, **31**, (11), 387; and *Chem. Age*, 1951, **64**, (1647), 211–212).—B. and G. report on experiments relating to the electrodeposition of thin Tl films. Following the actual electrolytic deposition, a potentiometric method of analysis was used to determine Tl^+. This was based on the reaction of Tl salts with $KMnO_4$. The best results were obtained with $TlAl(SO_4).12H_2O$ and with $TlBF_4$.—I. S. M.

***The Electrodeposition of Thallium and Thallium–Lead Alloys.** E. Bertorelle, L. Giuffrè, and A. Tunesi (*Chim. e Ind.*, 1950, **32**, (11), 517–522).—The results of electrolysing fluoboric Tl soln. contg. a quantity of free acid and some organic substances are given. Using these soln. it is possible to deposit Tl in compact form and in very thin films. After giving the electrochem. properties of Tl and Pb films electrodeposited from fluoboric soln., the optimum conditions for electrodepositing Tl–Pb alloys are indicated. 16 ref.—I. S. M.

***The Electrodeposition of Tin–Antimony Alloys from Chloride–Fluoride Electrolytes.** J. W. Cuthbertson and N. Parkinson (*J. Electrodepositors' Tech. Soc., Advance Copy*, 1952, **28**, (6), 7 pp.).—It is shown that alloys of Sn and Sb contg. up to ~45% Sb can be satisfactorily deposited from a soln. contg. chlorides of the metals with a fluoride as an addn. agent. The work described is of only a preliminary nature, but it is considered to be of significance that bright plate can be deposited. The conditions for deposition are fully described, and the effect of the Sb content of the electrolyte on the compn. of the deposit has been studied, as well as the effect of the fluoride content of the electrolyte. Hull-cell tests have been carried out on the effect of electrolyte compn. and temp. on the c.d., plating range, and the appearance of the deposit. 5 ref.—G. T. C.

The Development of Tin–Zinc Alloy Plating in the U.S.A. Frederick A. Lowenheim and Robert M. MacIntosh (*J. Electrodepositors' Tech. Soc.*, 1951, **27**, 115–128).—American interest in Sn–Zn alloy plating developed only in 1950 and was stimulated by the acute shortage of Cd and the concurrent development of an all-K bath which made the process directly competitive with Cd plating in speed of deposition. L. and MacI. present full details of the compn. and operating conditions of the K-type bath. Its compn. is: K stannate 120, $Zn(CN)_2$ 9, KCN 19, and free KOH 6·8 g./l. The operating conditions are: temp. 150° ± 5° F. (66° ± 3° C.); cathode c.d. 10–75 amp./ft.^{2}; anode c.d. 15–25 amp./ft.^{2}; anode : cathode ratio 1·5 : 1 to 2 : 1; anodes of cast 80 : 20 Sn–Zn alloy; and voltage 3·5–6·0 V. Information is also included on a similar bath suitable for barrel plating. Details are included on analytical control methods. 12 ref.—G. T. C.

***An Electron-Diffraction Investigation of the Structure of Electrodeposited Coatings on Iron Single Crystals.** (Evans and Hopkins). See col. 408.

Finishes for Communications Equipment, with Special Reference to Electroplate Coatings. E. C. J. Marsh (*J. Electrodepositors' Tech. Soc.*, 1952, **28**, 69–88).—The different types of finish are briefly reviewed, and an indication is given as to how a finish is chosen and tested. Various types of electrodeposited finish on steel and on non-ferrous metals are described, and there is information on plant used in the finishing shop. The influence of design on finishing is also discussed.—G. T. C.

Fatigue Loss and Gain by Electroplating. (Almen). See col. 390.

Electroforming Solves Difficult Design and Fabrication Problems on New Military Items. M. H. Orbaugh (*Western Metals*, 1952, **10**, (4), 46–48).—Advantages and appn. of electroforming, using Ni, Cu, Fe, Ag, Au, brass, and Al, are described briefly.—T. G.

Your Dentist's a Plater, Too. Walter J. Krumbeck (*Plating*, 1952, **39**, (10), 1122–1124).—The various electrochem. techniques employed in dental laboratories and in dentistry are briefly described. These include electroforming and electropolishing.—G. T. C.

Factors Influencing the Design of Automatic Plating Plant. A. Smart (*J. Electrodepositors' Tech. Soc.*, 1951, **27**, 277–288; discussion, 289–291).—The development of automatic plant is traced from the stage of the preliminary survey, through laboratory-scale testing, pilot-scale testing, plant design, and final installation. Brief information is included on elect. equipment, testing, and operation of the plant and its performance under prodn. conditions.—G. T. C.

9—ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

***The Electrode Potential of Manganese in Aqueous Solutions.** R. I. Agladze and A. E. Legran (*Zhur. Fiz. Khim.*, 1950, **24**, (9), 1122–1127).—[In Russian]. Measurements were made on electrolytic Mn, previously cleaned on a carborundum wheel, in soln. of $MnSO_4$, $(NH_4)_2SO_4$, K_2SO_4, H_2SO_4, $MnSO_4 + H_2SO_4$, and $MnSO_4 + (NH_4)_2SO_4$. The concentrations of the salts were 0·01, 0·1, and 1·0*N*, and those of H_2SO_4 0·01, 0·1, and 0·5*N*. All measurements were made by the null method against a saturated calomel electrode at 24°–25° C. A. and L. deduce that: (*a*) in K_2SO_4 soln. the electrode potential of Mn is practically independent of the concentration of K_2SO_4; (*b*) in $MnSO_4$ soln. and in H_2SO_4 soln., the potential becomes more negative in more dil. soln.; (*c*) the change of potential with time of immersion in the soln. is caused by oxide formation on the electrode surface, which is favoured by weakly acid and neutral soln.; (*d*) in 0·5*N*-H_2SO_4 soln. contg. 0·01, 0·1, or 1·0*N*-$MnSO_4$, or in 0·1*N*-H_2SO_4 soln. contg. 0·01 or 0·1*N*-$MnSO_4$, the potential varies only with the *p*H, but for the other conditions investigated it depends also on the concentration of Mn ions; (*e*) in $(NH_4)_2SO_4$ soln. the potential becomes less negative with increasing dilution, in contrast to soln. of $MnSO_4$ and H_2SO_4; this can be explained by the formation of the complex ion $[Mn(NH_3)_x]^{++}$; and (*f*) addn. of $(NH_4)_2SO_4$ to $MnSO_4$ soln. makes the potential more negative; this also can be explained by the formation of complex ions.—G. B. H.

***Study of the Sintering Process in Platinized Platinum by Electrochemical Methods. II.—Heat-Treatment in Air.** M. N. Nikolaeva and A. I. Shlygin (*Zhur. Fiz. Khim.*, 1950, **24**, (5), 534–538).—[In Russian]. N. and Sh. repeated their experiments (*ibid.*, (4), 427; *M.A.*, **20**, 360) on the changes in sp. adsorption power for H and catalytic effect on the decompn. of H_2O_2, with platinized Pt heated for 1 hr. at 50°–400° C. in air. The adsorption power for H falls almost linearly with increasing temp., reaching 50% of the initial value after treatment at 400° C. The adsorption power for O, determined after measurement of the adsorption power for H, shows similar behaviour, but the amount of O adsorbed at the temp. of heat-treatment, as measured from cathodic polarization curves, rises to a max. of double the initial value at 200° C. and then falls to ~60% of the initial value at 400° C. N. and Sh. explain these facts by assuming that the electrode is not saturated with adsorbed O at room temp., and that the max. is caused by the combined effect of decreasing adsorption power at saturation and increasing rate of approach to saturation. The catalytic effect on the decompn. of H_2O_2 varied with temp. of heat-treatment in a manner qual. similar to that of specimens heated in H, but the whole curve was shifted to higher temp. The velocity const. of the catalysed reaction fell to half the initial value after treatment of the platinized Pt at 100° C. and then rose to a max. after treatment at 300° C.; but even after treatment at 400° C. it was higher than the initial value. These results differ markedly from those of Maxted and Moon (*J. Chem. Soc.*, **1935**, 393), who heat-treated platinized Pt *in vacuo*.—G. B. H.

***Passivity of Titanium in Hydrochloric Acid Solutions.** David Schlain and Joseph S. Smatko (*J. Electrochem. Soc.*, 1952, **99**, (10), 417–422).—Electrode-potential measurements and wt.-loss tests indicate that air or low-concentrations of certain dissolved metals passivate Ti in HCl and that passivity is accompanied by more noble electrode potentials. Air passivates Ti in acid up to 3*N*. 200 mg./l. Cu^{++} is effective in 10*N*-HCl and 2 mg./l. Cu^{++} in 5*N*-HCl. The more noble electrode potentials indicate that a reaction involving O, metal ions, or ions contg. O takes place at the passive surface, but there is no visible deposit on, or change in, the surface, apart from an occasional slight tarnishing. After a time in contact with 10*N*-HCl contg. 200 mg./l. Cu^{++}, Ti loses its passivity, the time interval being apparently directly proportional to the ratio of soln. vol. to specimen area. The loss of passivity is attributed to the presence of atomic H. 14 ref. —G. T. C.

***Investigation of the Potentials of Alloys Formed in Galvanic Cells as a Method of Physico-Chemical Analysis.** D. P. Zosimovich (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, **17**, 241–248; *C. Abs.*, 1951, **45**, 1900).—[In Russian]. Closing the circuit formed by 2 metal electrodes immersed in a soln. of a salt of the more electronegative of the metals causes the electronegative metal in soln. to plate out on the electropositive one and alloy with it. As the alloy forms, it goes through all the phases that exist in the particular binary system, and the formation of the phases is reflected in the e.m.f. of the cell. A specially designed cell was used to study the Zn–Ag, Cu–Au, Ni–Cu, and Zn–Cd systems. The results were plotted as current/e.m.f. curves. The breaks in the curves corresponded well with the formation of phases. This method also facilitates the detn. of certain critical points which are hard to determine otherwise. The method is therefore considered as a suitable procedure in physicochem. analysis.

***On the Nature of " Secondary Potentials " During the Cathodic Deposition of Metals.** M. A. Loshkarev and A. M. Ozerov (*Zhur. Fiz. Khim.*, 1950, **24**, (6), 731–744).—[In Russian]. L. and O. examine Kudra's hypothesis (*ibid.*, 1937, **11**, 286; 1938, **12**, 148) that the powdery deposits formed in the electrodeposition of metals at high c.d. are caused by the discharge of complex ions; in support of this view, K. produced evidence of a " second decompn. potential " on the polarization curves of electrolytes at high c.d. L. and O. studied this phenomenon in the electrolysis of the following soln. at 30° C.: 0·05N-$CdSO_4$, 0·05N-$CuSO_4$, 0·05N-$ZnSO_4$, 0·05N-Ag_2SO_4, and 0·05N-$HgSO_4$. To all these soln. were added, as a supporting electrolyte, 1N-Na_2SO_4 and 0·025N-H_2SO_4. The first four soln. were also examined without H_2SO_4. The experiments were repeated with soln. of the chlorides of the same metals and a supporting electrolyte of 1N-NaCl, with and without 0·01N-HCl, and also with soln. of Br, I, and $K_3Fe(CN)_6$. In all measurements a dropping Hg cathode was used, the area of the cathode surface being 0·2 cm.^{2} and the max. current 2 m.amp.; the anode was a pool of Hg at the bottom of the cell. The cathode potential was measured potentiometrically with respect to a saturated calomel electrode. To prevent irregularities caused by surface mobility on the cathode, it was usually necessary to add ~0·1 g./l. of saponin, gelatin, or other surface-active agents to the electrolyte. For all the electrolytes studied, the polarization curves showed a second decompn. potential at −1·5 V. in acid soln. and a third decompn. potential at −2·0 V. in both acid and neutral soln., irrespective of the metal ion. The second potential is attributed to the discharge of H^+ ions, and the third potential to the dissociation of H_2O mol. with increase of the pH in the cathode double layer to ~10. By interrupting the polarizing current at 40 c./s. and examining the variation in cathode potential with a cathode-ray oscillograph, it was shown that there is a considerable " ohmic " p.d. across the cathode double layer when soln. without the supporting electrolyte were electrolysed at c.d. above the third decompn. potential. Addn. of supporting electrolyte (1N-Na_2SO_4 or 1N-NaCl) removed this effect.—G. B. H.

On the Dependence of Hydrogen Overvoltage on the Nature of the Cathode Metal. A. K. Lorents (*Zhur. Fiz. Khim.*, 1950, **24**, (7), 853–855).—[In Russian]. On the assumption that the H overvoltage (η) on a metal varies with the c.d. (i) according to the relation $\eta = a + b \ln i$, b is a const. equal to 0·05 for many metals, and a is another const. which varies considerably from metal to metal. On the assumption, on which slow-discharge theory is based, that the rate-determining reaction in the evolution of H is $H_3O^+ + \ominus \rightarrow H + H_2O$, L. shows that the const. a is a function of the chem. potential of the electron gas in the metal, which, according to Frenkel (" Statisticheskava Fizika ", **1948**), $\propto 1/\sqrt{\kappa}$, κ being the compressibility of the metal. From a survey of the literature, L. shows that for many metals a is given by the relation: $a = 1{\cdot}92 - 0{\cdot}94/\sqrt{\kappa \times 10^6}$.—G. B. H.

On Some Regularities in Hydrogen and Oxygen Overvoltage Phenomena. N. E. Khomutov (*Zhur. Fiz. Khim.*, 1950, **24**, (10), 1201–1203).—[In Russian]. Kh. plots the values of Bockris (*Trans. Faraday Soc.*, 1947, **43**, 417; *M.A.*, **15**, 242) and of Hickling and Salt (*ibid.*, 1940, **36**, 1226; *M.A.*, **8**, 116) for the H overvoltage of various metals in acid soln. at a c.d. of 10^{-3} amp./cm.^{2} against the interatomic distances in the metals. The results lie on a curve with a min. overvoltage at 2·74–2·76 Å. Similar treatment of H. and Hill's values (*Discussions Faraday Soc.*, 1947, (1), 236; *M.A.*, **17**, 589) for O overvoltage in alkaline soln. at a c.d. of 10^{-3} amp./cm.^{2} shows a max. O overvoltage at 2·74–2·76 Å. Kh. suggests a connection between these facts and the intermolecular distance of 2·76 Å. in solid H_2O. The best values of the const. b in Tafel's equation for H overvoltage give $b = 2{\cdot}3RT/\alpha F$, where $\alpha = 0{\cdot}49$–$0{\cdot}74$; the mean value, 0·62, is close to the cosine of half the angle between the O–H bonds in the H_2O mol. Hence Kh. suggests that in the electrolysis of H_2O the mol. take up a definite orientation on the electrode, which, in the most favourable case, places the O–H bonds at an angle of $\cos^{-1}$ (0·62) to the elect. field.
—G. B. H.

***Potentials for the Evolution of Hydrogen in Acid Solutions.** A. G. Pecherskaya and V. V. Stender (*Zhur. Fiz. Khim.*, 1950, **24**, (7), 856–859).—[In Russian]. In order to provide useful information on indust. electrolytic processes, P. and S. measured the H overvoltages of several metals in 2N-H_2SO_4 soln. at 25° ± 0·5° C. and c.d. of 10–2000 amp./m.^{2}. The metals, in the form of plates, were mounted in plastic with an area 1 cm. × 1 cm. exposed. Before measurement, the specimens were ground on 000 emery paper, polished with dry wool, and cleaned by a preliminary cathodic polarization, first in 5N-NaOH soln. for 1 hr. at 25° C. and a c.d. of 200 amp./m.^{2}, and then, after washing in distilled water, in 2N-H_2SO_4 for 1 hr. at 25° C. and a c.d. of 200 amp./m.^{2}. For Ge, which was obtained as a bead, the exposed electrode surface was a polished flat of area 22 mm.^{2}, but the rest of the procedure was as described. The measurements were made in an open cell with a Pt anode somewhat smaller than the cathode, the distance between the electrodes being 70 mm. An HgO comparison electrode was used. The metals studied were as follows: electrolytic Al, remelted and rolled down to 0·5 mm.; W, Mo, and Ta in 0·3–0·5-mm. sheet, made by powder-met. methods from powder prepared by H reduction of very pure oxides; a melted Ge bead; commercial 0·5-mm. soft Fe sheet; electrolytic Co; electrolytic Cd, remelted and cast in a plate; electrolytic Cu, deposited from soln. in H_2SO_4; electrolytic Ni; electrolytic Sn; Re electrolytically deposited from $KReO_4$ soln.; Pb prepared by electrolysis of a soln. in H_2SiF_6, remelted and cast; Sb deposited from soln., remelted and cast; commercial 99·95% Ag, made from electrolytic Ag; electrolytic Zn, coated with a fresh deposit of Zn. H overvoltages were found to obey the relation $\eta = a - b \log_{10} i$, where η is the overvoltage at c.d. i, and a and b are const. Values of η in V. (negative with respect to the HgO electrode) at 10 and 2000 amp./m.^{2}, resp., and values of b are as follows: Pb, 1·05–1·26, $b = 0{\cdot}095$; Zn, 0·83–1·22, $b = 0{\cdot}17$; Sb, 0·63–0·86, $b = 0{\cdot}10$; Al, 0·58–0·79, $b = 0{\cdot}09$; Sn, 0·63–0·86, $b = 0{\cdot}10$; Cd, 0·51–1·00, $b = 0{\cdot}21$; Cu, 0·48–0·78, $b = 0{\cdot}13$; Ta, 0·46–0·61 (at 1000 amp./m.^{2}), $b = 0{\cdot}075$; Ge, 0·39–0·92, $b = 0{\cdot}23$; Fe, 0·36–0·64, $b = 0{\cdot}12$; Mo, 0·35–0·59, $b = 0{\cdot}105$; Ag, 0·34–0·53, $b = 0{\cdot}08$; Co, 0·32–0·56, $b = 0{\cdot}10$; Ni, 0·30–0·51, $b = 0{\cdot}095$; W, 0·26–0·49, $b = 0{\cdot}10$; Re, 0·09 (at 50 amp./m.^{2})–0·32, $b = 0{\cdot}14$.
—G. B. H.

Electrolytic Processes in Chemical and Metallurgical Industry. H. J. T. Ellingham (*Chem. and Ind.*, 1952, (46), 1115–1118).—Paper read at a Symposium organized jointly by the Society of Chemical Industry and the Royal Institute of Chemistry. Fundamental processes underlying indust. electrolytic processes are reviewed.—J. R.

11 — ANALYSIS

***Industrial Method for Determining Oxide in Cast Aluminium, in Remelted Aluminium, and in the Refined Metal.** P. Urech, R. Sulzberger, and E. Schaad (*Chimia*, 1950, **4**, (10), 233–235; *C. Abs.*, 1951, **45**, 1914).—Two procedures have been used for determining Al_2O_3 in Al metal: (1) The sample is heated in a stream of HCl + H at a temp. such that the $AlCl_3$ is sublimed, and the residue of oxide is weighed; (2) the vol. of H evolved on treatment with acid is measured and calculated to the corresponding weight of Al, which is deducted from the original weight taken for analysis and the difference in the weights is called Al_2O_3. Suitable apparatus is illustrated, and the procedure for carrying out the first method is given. It is recommended, however, that the residue be filtered off, ashed in a crucible, heated with a 10% soln. of a mixture of 3 parts Na_2CO_3 + 1 part $Na_2B_4O_7$, evaporated to dryness, and fused. From the melt, Fe_2O_3 cannot be extracted with water, but Al_2O_3 can. To the water extract add 5% HCl soln. to *p*H 2, make up to 100 ml., and use a 25-ml. aliquot. To this add 5 ml. of buffer soln. (contg. 80 g. Na acetate + 2 ml. acetic acid in 200 ml. water) and add 10 ml. of reagent contg. 2 g. eriochromcyanin and 2 ml. acetic acid in 2 l. After 5 min., measure the colour in a Lange colorimeter with red filter RG 2 or in a Lumetron with filter No. 610. Numerous samples examined by this method gave values ranging from 0·004 to 0·081% Al_2O_3. The values are much lower than those given by the Br–methanol method of Werner (*Z. anal. Chem.*, 1941, **121**, 385; *M.A.*, **9**, 362).

***The Bromate Titration of Tervalent Arsenic.** D. J. Kew, M. D. Amos, and M. C. Greaves (*Analyst*, 1952, **77**, (918), 488–489).—Titrations of As^{III} at 80° C. with $KBrO_3$ and methyl orange as indicator are accurate, provided that the acidity at the end-point is maintained between 1·2 and 3·5*N* with respect to HCl.—F. M. L.

***Precipitation and Determination of Beryllium as Oxide.** Thérèse Dupuis (*Mikrochemie ver. Mikrochim. Acta*, 1950, **35**, 477–487; *C. Abs.* 1951, **45**, 493).—The pptn. of Be^{++} by cold and by hot NH_4OH, a current of NH_3, hexamethylenetetramine, guanidinium carbonate, KCN, NH_4 benzoate, H_2SeO_3, NH_4NO_2, tannin, mixture of KI and KIO_3, $(NH_4)_2CO_3$, α-picoline, $C_2H_5NH_2$, hydrazinium carbonate, $(NH_4)_2S$, Na_2SO_3, NaOH, and oxine was studied with the Chevenard thermobalance. The min. temp. to which the precipitates must be ignited to convert the $Be(OH)_2$ into weighable BeO were, in the above order, 851°, 595°, 922°, 885°, 1000°, 900°, 1000°, 571°, 555°, 925°, 817°, 1000°, 817°, 780°, 1000°, 592°, 970°, 576°–866°, and 824° C.

Determination of Boron in Certain Metal Borides. Herman Blumenthal (*Powder Met. Bull.*, 1951, **6**, (2), 80–82).—The detn. of B in the borides of Mg, Mn, Mo, and Co is briefly described.—W. A. M. P.

***Alkalimetric Method for the Estimation of Cadmium and Zinc.** Mahadeo M. Tillu (*Analyt. Chem.*, 1952, **24**, (9), 1495–1496).—$Cd(OH)_2$ and basic Zn carbonate can be dissolved in cold thiosulphate soln. and the liberated alkali titrated with standard acid.—F. M. L.

The Polarography of Cadmium. Paolo Silvestroni (*Ricerca Sci.*, 1950, **20**, (11), 1699–1705).—A review. The position occupied by Cd in the electromotive series under various conditions is considered, and the various procedures by which Cd may be determined in Cd-base materials are described.
—I. S. M.

***The Colour Reactions of Chloranilic Acid, with Particular Reference to the Estimation of Calcium and Zirconium.** R. E. U. Frost-Jones and J. T. Yardley (*Analyst*, 1952, **77**, (918), 468–472).—The colour reactions of chloranilic acid have been studied in detail, with particular ref. to interference in the detn. of small amounts of Ca and Zr.—F. M. L.

***Iodometric Determination of Copper.** Louis Meites (*Analyt. Chem.*, 1952, **24**, (10), 1618–1620).—In the conventional iodometric detn. of Cu accuracy is increased if sufficient KI is added to dissolve all the Cu_2I_2 formed, leaving a clear soln. for titration.—F. M. L.

***Spectrographic Determination of Gallium in Aluminium.** F. Farhan (*Mikrochemie ver. Mikrochim. Acta*, 1950, **35**, 565–568; *C. Abs.*, 1951, **45**, 494).—A series of standard alloys of pure Al with known Ga content were prepared and analysed by the intermittent-arc method. As little as 0·01% Ga can be determined in this way.

***Determination of Gold and Platinum Metals in Gold Bullion.** D. I. Ryabchikov and G. B. Knyazheva (*Izvest. Sekt. Platiny*, 1948, (**22**), 121–128; *C. Abs.*, 1951, **45**, 62).—[In Russian]. Dissolve the sample in *aqua regia* and add twice the theoretical quantity of NaCl to the soln. The Au forms $Na[AuCl_4]$, in which form the trivalent state is stable and suitable for titration. The titration is carried out potentiometrically with hydroquinone or $(NH_4)_2Fe(SO_4)_2$ soln., the end-point being marked by a jump in potential. To determine the Pt metals in the sample, dissolve 5 g. of metal in *aqua regia*, and boil off N oxides, NOCl, and excess HCl. Filter to remove AgCl and an insoluble residue, and to the soln. add twice the theoretical quantity of NaCl. The soln. now contains Au, Pt, Ir, and Pd. Transfer to a 250-ml. flask and dilute to vol. To a 25-ml. aliquot add 20 ml. of a saturated NH_4 oxalate soln., and boil for 15–20 min. Au is reduced to metal, which is filtered and weighed. Evaporate the combined filtrate and wash water to half its vol. and add 20–30 ml. of 25% H_2SO_4. Heat to boiling and add 0·3 *N*-$KMnO_4$ slowly until a stable colour appears. This destroys the oxalate and oxidizes Pt, Ir, and Pd to their highest valency. In this soln. Pt is determined potentiometrically, as outlined by R. (*ibid.*, p. 28; *M.A.*, **20**, 365). If the soln. is coloured brown, add to the titrated soln. a little Mohr's salt soln. If the colour fades, it indicates the absence of Pd and the presence of Ir. If the colour remains, it indicates the presence of Pd. In either case, titrate the reduced soln. with 0·1*N*-$KMnO_4$. The 1st jump in potential is due to $Fe^{2+} \rightarrow Fe^{3+}$ and the 2nd to $Ir^{3+} \rightarrow Ir^{4+}$ if present. In the latter case Pt and Ir are calculated accordingly. To the titrated soln. add 20–30 ml. of a saturated aq. soln. of dimethylglyoxime, allow to stand overnight, filter, and weigh the dried Pd precipitate.

Absorption Characteristics of the Dithizone Mixed Colour System [for the Determination of Trace Amounts of Lead]. Robert G. Milkey (*Analyt. Chem.*, 1952, **24**, (10), 1675–1676).—A theoretical treatment.—F. M. L.

***Determination of Traces of Magnesium in Aluminium Metal by (*p*-Nitrophenylazo)-1-Naphthol (Magneson II).** F. Farhan (*Mikrochemie ver. Mikrochim. Acta*, 1950, **35**, 560–564; *C. Abs.*, 1951, **45**, 494).—Weisselberg (Dissertation, Vienna, **1930**) recommended this reagent for detecting Mg, and F. describes the spectrophotometric appn. of the test. Treat 10 g. of sample with 200 ml. 6*N*-HCl, heating gently. Cool the soln. to −15° C. and saturate it with HCl gas. Filter through sintered glass. Add to the filtrate an equal vol. of twice distilled water and evaporate almost to dryness. Add water, heat to boiling, and add NH_4OH. Saturate with H_2S, filter, wash, and examine the precipitate for traces of Mg. Evaporate the filtrate in Pyrex glass until no more NH_4Cl fumes are evolved. Take up in a little 6*N*-HCl and dilute to 50 ml. Add 5 ml. of a 0·005% soln. of the reagent + 5 ml. of *N*-NaOH, stir, and measure the opt. *d* at 650 mμ.

The Polarography of Mercury. Paolo Silvestroni (*Ricerca Sci.*, 1950, **20**, (12), 1861–1866).—A summary of published literature on the behaviour of the dropping Hg cathode in soln. of its salts. 30 ref.—I. S. M.

***The Direct Determination of Chromatographized Elements on Paper with a Microgramme Scale: Molybdenum and Vanadium.** (Mlle) Alice Lacourt, (Mlle) Ghislaine Sommereyns, (Mlle) Jeannine Hoffmann, (Mme) Anne Stadler, and Georges Wantier (*Compt. rend.*, 1952, **234**, (24), 2365–2367).—The direct estn. of Mo and V on the paper after chromatography without elution or incineration is described. The method is sensitive, the reproducibility good, and the time is reduced from 2–3 hr. for chromatographic sepn. to 15–20 min.—J. H. W.

***The Determination of Nickel and Manganese in Uranium.** J. Haslam, F. R. Russell, and N. T. Wilkinson (*Analyst*, 1952, **77**, (918), 464–467).—After dissoln. of the U and removal of SiO_2 and acid-insoluble sulphides from the soln., Ni and Mn are co-precipitated with added Fe by adding an excess of Na_2CO_3 soln. to the cold sample soln., boiling, and then diluting and re-boiling. The mixed precipitate is dissolved in acid and Ni determined by the hypobromite–dimethylglyoxime method, and Mn, after oxidation to permanganate, by titration with standard ferrous ammonium sulphate.—F. M. L.

The Analysis of Commercial Platinum. Ivo Ubaldini and Luisa Nebbia (*Chim. e Ind.*, 1951, **33**, (6), 360-361).—A method is described by which the metals that are not in the Pt group are precipitated with $NaBrO_3$, after which Pd is separated, followed, finally, by the sepn. of Rh from Ir with 2-mercaptobenzothiazole. 7 ref.—I. S. M.

***Spectrographic Determination of Small Quantities of Platinum and Palladium in Refined Silver.** A. V. Babaeva and E. S. Lapir (*Izvest. Sekt. Platiny*, 1949, (23), 97–100; *C. Abs.*, 1951, **45**, 1908).—[In Russian]. For determining Pt and Pd in Ag the 3-standard method was used. A total of 15 standards was used, covering the range 0·00005–0·200% Pt and Pd. The standard alloys were forged and drawn into 3–4-cm. lengths, having a sq. cross-section of 2·5 mm.^{2} and one end wedge-shaped. The wires were used as electrodes. For determining concentrations of 0·02–0·2%, the following lines were used: Pt 2702·40, Ag 2721·77; Pt 2733·96, Ag 2721·77; Pd 3489·77, Ag 3501·94; Pd 3460·77, Ag 3501·94 Å.; and for concentrations of 0·001–0·02% the following: Pt 2997·97, Ag 3099·12; Pt 2997·97, Ag 2721·77; Pd 3489·77, Ag 3542·61; Pd 3242·70, Ag 3130·01 Å.

***Colorimetric Determination of Silver with *p*-Dimethylaminobenzalrhodanine.** G. C. B. Cave and David N. Hume (*Analyt. Chem.*, 1952, **24**, (9), 1503–1505).—By carefully controlling acidity, concentration, and amount of rhodanine, time of standing, agitation, exposure to light, and presence of inert salts, organic solvents, and foreign anions, Ag can be determined accurately and reproducibly at a concentration of $10^{-6}M$.—F. M. L.

***Determination of Sodium and Lithium in Welding Flux for Aluminium Alloys.** Tsunenobu Shigematsu, Hirozo Kimura, and Masao Yasumaru (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (3), 173–176).—[In Japanese]. A new difference method for analysis of Na and Li in welding flux for Al alloys has been developed. When Na was determined grav. by pptn. as a triple salt with Ni uranyl acetate or Zn uranyl acetate, Li and K were co-precipitated, and the results were high. The Mg uranyl acetate reagent was less sensitive to Li and K, and under suitable conditions Na was precipitated alone. The Zn uranyl acetate (Li) was less sensitive to K than Ni uranyl acetate and precipitated Na and Li as a triple acetate without K interference. The Li content was given by difference. —AUTHORS.

***The Polarographic Determination of Titanium in Aluminium Alloys.** R. P. Graham and A. Hitchen (*Analyst*, 1952, **77**, (919), 533-537).—A sample of Al alloy is first leached with NaOH, and then Ti is extracted from the residue with H_2SO_4. This soln. is neutralized with NH_4OH, and H_2SO_4 and tartaric acid are then added so that it is $1M$ with respect to tartaric acid, $0{\cdot}5M$ to H_2SO_4, and $1{\cdot}2M$ to $(NH_4)_2SO_4$. Ti is then determined polarographically.—F. M. L.

***The Absorptiometric Determination of Zirconium by Means of Alizarin-S, with Special Reference to Magnesium Alloys.** A. Mayer and G. Bradshaw (*Analyst*, 1952, **77**, (918), 476–483).—Alizarin sulphonate gives an almost sp. coloured reaction product with zirconyl ions which is of definite compn. and stable in acid soln. Using this reagent Zr can be determined rapidly and accurately in Mg alloys, Fe and steel, and minerals. A method of determining the approx. compn. of HfO_2–ZrO_2 mixtures is briefly indicated. —F. M. L.

***Photometric Determination of Zirconium in Magnesium Alloys.** Glenn B. Wengert (*Analyt. Chem.*, 1952, **24**, (9), 1449–1451).—The photometric alizarin red S method is used for the detn. of Zr in Mg alloys.—F. M. L.

***The Separation and Detection of Zirconium and Hafnium on Paper Strips.** N. F. Kember and R. A. Wells (*Chem. and Ind.*, 1952, (46), 1129).—A letter. Good chromatographic sepn. of Zr and Hf from prepared mixtures of the nitrates has been obtained by the use of dichlorotriethyleneglycol in the solvent, and developing with a sprayed saturated soln. of alizarin in ethyl alcohol contg. 5 vol.-% of $2N$-HCl, followed by gentle heating. Chromatograms obtained are reproduced.—J. R.

***Application of Powdered Samples to Graphite Electrodes for Spectrochemical Analysis.** Ray C. Hughes (*Analyt. Chem.*, 1952, **24**, (9), 1406–1409).—Powdered, solid, insoluble, non-conducting substances are dispersed in glycerol, and the flat end of a graphite electrode dipped into the dispersion. The glycerol is then driven off by heating to leave a thin adherent coating of sample which will burn in a smooth and reproducible manner in an A.C. arc.—F. M. L.

---

13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

Compression Test Extensometer for Cylindrical Specimens. A. M. Stott and J. M. McCaughey (*Product Eng.*, 1951, **22**, (3), 157–160).—The development of an instrument for measuring strain in cylindrical specimens in compression, by measuring changes in dia. recorded by resistance strain-gauges, is described.—M. A. H.

***Hot Fatigue Testing.** H. E. Gresham and B. Hall (*Iron Steel Inst. Special Rep.*, 1952, (**43**), 181–185; discussion, 355–367).—The design and construction of a single-point-loading fatigue machine in which tests can be carried out at elevated temp. in air, artificial combustion products, or solid corrosive media is described. The furnace can be replaced by a salt-spray chamber. The effects of temp. and corrosive media on fatigue are illustrated by the results of a few experiments.—D. M. P.

***The Use of the Electron Microscope for Measuring Very Small Indentations Produced in Microhardness Testing.** M. M. Khrushchev and E. S. Berkovich (*Izvest. Akad. Nauk S.S.S.R.*, **1950**, [Tekhn.], (11), 1645–1647).—[In Russian]. The error in measurement of indentations having a diagonal <5 μ has hitherto rendered such measurement insufficiently reliable. Kh. and B. show that with the aid of electron micrographs, measurements can be taken, in hardness tests with a diamond pyramid, of indentations having linear dimensions <1 μ. 1 μ impressions made with a triangular diamond pyramid produce clearer indentations than those made with a sq. pyramid, thus permitting more accurate measurement. Photographs and relevant data of hardness tests of a ruby and tempered steel are included.—H. W.

Dip and Etch Simplify Aluminium Forging Inspection. Alfred H. Pope (*Iron Age*, 1952, **170**, (14), 102–103).—Al forgings are dipped in 10–12% HNO_3 for 15 sec., washed, alkaline-etched in Pennsalt AE-18 at 140° F. (60° C.), washed, re-immersed in acid to remove any smut remaining after etching, washed, and dried. A discoloration remains in cracks and in discontinuities caused by metal overlap.—J. H. W.

***Photoelastic Study of the Stresses in the Matthaes Stress-Corrosion Test-Piece.** W. Ruff (*Alluminio*, 1950, **1**, (5), 411-415).—R. determined stress distribution by photoelastic methods and confirmed results previously obtained by tensile tests on a compressed test-piece of Superduralumin D.—I. S. M.

***Ultrasonic Examination of Alloys.** Italo Barducci (*Ricerca Sci.*, 1951, **21**, (6), 897–903; also *Nuovo Cimento*, 1950, **7**, (Suppt. No. 2), 347–358).—A summary is given of the present state of knowledge regarding the effects of alloying elements on the damping and internal friction of metals subjected to ultrasonic vibration. The importance of data obtained in this way is emphasized, and the principal causes of internal friction, whether common to alloys and pure metals, or peculiar to alloys only, are discussed. 22 ref.—I. S. M.

The Ultrasonic Examination of Metals. A. E. Williams (*Min. J.*, 1949, **233**, (5961), 1098–1100).—A description and appn. of the ultrasonic flaw detector.—S. R. W.

***On a Method of Suppressing Secondary Waves in Ultrasonic Testing of Metallic Specimens.** Louis Beaujard (*Compt. rend.*, 1952, **235**, (15), 804–806).—It is shown that secondary echoes are due to surface waves that pass directly from the transmitter to the receiver. These echoes can be suppressed by photographic integration, whereas the echoes due to defects are unaffected.—J. H. W.

Some Simple Methods for the Non-Destructive Examination of Defects in Work-Pieces. J. van Nieuwkoop (*Metalen*, 1952, **7**, (15), 247–254).—[In Dutch]. After a review of methods for the detn. of defects extending to the surface of magnetic materials, van N. gives methods for non-magnetic materials, including those methods using fluorescent oil, electronic and elect. resistance apparatus. 12 ref.—I. S. M.

***Inspection of Light Alloys by Radiography: On Some Difficulties in the Interpretation of the Negatives.** Gaston Gauthier and Martial Renouard (*Rev. Mét.*, 1952, **49**, (10), 720–725; discussion, 726).—G. and R. carried out experiments on the radiographic examination of forged Al alloys, their results being complementary to those of Glaisher, Betteridge, and Eborall on cast Al alloys (*J. Inst. Metals*, 1944, **70**, 81; *M.A.*, **11**, 419). The object of the examination was to detect certain discontinuities, such as inclusions, segregates, cracks, and heterogeneity. The results showed that radiographic examination alone may lead to erroneous conclusions, since the diffusion or diffraction spots may cause abnormal contrasts in zones of oriented or nearly oriented crystn., and these spots may mask the real defects or may exaggerate slight local inhomogeneities. The appearance of defects, such as segregation or cracks, requires close examination before the defect can be attributed to a particular cause, and radiographic examination must always be supplemented by other means of inspection, such as macrography, micrography, and semi-microradiography.—J. H. W.

***Radiography of Spot Welds in Various Sheet Gauges and Dissimilar Gauge Combinations.** R. C. McMaster, F. C. Lindvall, and L. P. Gaard (*Weld. J.*, 1947, **26**, (1), 19–29).—The influence of gauge and of combinations of gauges on the interpretation of radiographs of spot welds in Alclad sheet was investigated by radiographic examination of welds in two-layer combinations of sheets having (*a*) equal, and (*b*) dissimilar thickness, the gauges ranging from the min. to max. practicable. Commercially available equipment and film were found adequate for the radiography of welds over the entire range of sheet thicknesses normally spot-welded in aircraft construction. Increased thickness demands a longer exposure and higher exposure voltage, but radiographs show more detail than with thin sheet. Combination of different thicknesses did not influence the quality of the radiograph. 6 ref.—P. R.

Some Applications of Gamma-Radiography Techniques. M. Falk (*Atomics*, 1952, **3**, (10), 249–255).—The technique of γ-ray stereography, employing simultaneously two sources of γ-rays, normally of different strengths and mounted at a known distance apart on a graduated horizontal bar which can be raised or lowered, as desired, for the detection and localization of flaws in castings, &c., is described and illustrated. A re-arrangement of the apparatus makes it possible to examine a number of castings simultaneously. Apparatus for examining circumferential welds joining sections of penstocks, of dia. 3–20 ft., in hydroelect. plant, and easily transportable over rough country is described. A large exposure container has been designed to allow use to be made of the more powerful γ-ray sources now available. Its use is illustrated by the examination of the welds of a very large H-filled generator casing, necessitating 25 exposures, each of 1 hr., the examination of compressed-air forged-steel bottles with welded-in bottoms, and the study of possible porosity in a Mn-steel built-up weld in a Pelton turbine.—J. S. G. T.

Gamma Radiography in the United Kingdom. W. S. Eastwood (*Nucleonics*, 1951, **9**, (1), 50–59).—The use of artificially made γ-emitting isotopes for radiography is comprehensively reviewed and discussed. 21 ref.
—W. A. M. P.

Fluoroscopy with Gamma Rays. C. Garrett and A. Morrison (*Nucleonics*, 1951, **9**, (3), 44–51).—The use of a 14-mm.^{3} Ir^{192} γ-ray source for fluoroscopy is described.
—W. A. M. P.

Photographic Fundamentals of Autoradiography. R. H. Herz (*Nucleonics*, 1951, **9**, (3), 24–39).—The choice of suitable photographic material and the characteristics of various photographic emulsions for autoradiographic techniques are discussed in detail. 49 ref.—W. A. M. P.

X-Ray in Industry. R. J. Wright (*Australasian Eng.*, **1952**, (Sept.), 68–71).—The use of radiology for the following purposes is discussed: establishing the best method for making a casting or other part, checking quality during prodn., examining plant and equipment while in service. 9 ref.—T. A. H.

15 — FOUNDRY PRACTICE AND APPLIANCES

Hunter Douglas' Continuous-Casting Method Converts Molten Aluminium to Strip in a Matter of Minutes. J. L. Hunter and R. A. Quadt (*Western Metals*, 1952, **10**, (5), 33–35).—The prodn. of Al strip for Venetian blinds is described. The casting machine produces bars that are subsequently hot rolled to strip having a good surface and edge quality without scalping or edge trimming. The daily output is 70,000–80,000 lb. of 2S, 3S, 4S, 52S, 61S, and 24S.
—T. G.

Porosity Due to Dross [in Aluminium–Magnesium Alloy]. —— (*Metal Ind.*, 1952, **81**, (20), 391).—Porosity in castings is not necessarily due to gas pick-up. In a casting made to D.T.D. 165 specification (Al with 3–6% Mg) and exhibiting sub-surface porosity, it was found that the Si content was too high. Lowering the Si to the correct amount resulted in a more easily controlled slag and in sound, clean castings.
—J. H. W.

Modern Technique of Die-Casting Aluminium and Magnesium Alloys. B. Guastalla (*Alluminio*, 1948, **17**, (1/2), 7–33).—Uniform section thickness whenever possible, easy access to the thicker parts with runners and risers, and consideration of the casting properties of the alloys are recommended in the design of light-metal castings. The best systems of feeding light-metal die-castings are shown, with numerous photographs. Precise rules about the construction of the mould and the practical die-casting process are given to ensure reduction in the number of rejects.
—F. E. M.

Beryllium Copper Parts Designed for Investment Casting. John T. Richards (*Product Eng.*, 1951, **22**, (2), 114–117).—After outlining the limitations and advantages of investment casting, R. lists the properties of Cu-Be alloys and describes casting procedure, heat-treatment, and finishing.—M. A. H.

Use of Cast Metal Offers Important Savings for Limited Manufacture. Frank Charity (*Western Metals*, 1952, **10**, (3), 41–43).—The casting of moulds and dies for use as tools for various operations involving light and heavy non-ferrous metals is briefly described.—T. G.

***Crystallization and Heat-Transfer in a Stream of Liquid Metal Flowing Along the Surface of a Mould.** (Fell). See col. 410.

On the Fracture in Service of Cast Parts Submitted to Cyclic Stresses. Henri de Leiris (*Fonderie*, 1952, (81), 3125–3129).—From a general consideration of fractures due to cyclic stresses, it is shown that three essential conditions must be observed: (1) the dimensions laid down must be adhered to; (2) metal of good general soundness must be used; and (3) there must not be any local defects in the regions most strongly stressed.—J. H. W.

Fully-Automatic Die-Casting Developments. H. K. Barton (*Metal Ind.*, 1952, **81**, (16), 301–304).—All fully automatic die-casting machines require the const. services of one or more operators. Efforts are being made to obviate this, but it is doubtful whether much will be gained thereby. B. discusses the problems of producing small die-castings at fast rates and of the rapid prodn. of small but complex die-castings in Zn-base alloys. For these purposes the principle of ejecting the castings from one of the die members has been abandoned in favour of holding the castings stationary while both members are retracted from it.—J. H. W.

Die-Design for Automatic Die-Casting. H. K. and L. C. Barton (*Metal Ind.*, 1952, **81**, (17), 321–323).—The problems encountered in designing dies for a top operating speed of 1000 shots/hr. are discussed.—J. H. W.

Establishing Tolerances for Die-Castings. H. K. Barton (*Product Eng.*, 1951, **22**, (3), 118–123; (4), 118–123, 169–171).—Reasons for variations in dimensional accuracy and recommended tolerances for various materials are given.
—M. A. H.

The Future of Pressure Die-Casting. Frank G. Woollard (*Metal Ind.*, 1952, **81**, (20), 381).—A brief description of the metals available for pressure die-casting and of the appn. and advantages of the process.—J. H. W.

Precision-Casting of Turbine Blades. E. R. Gadd (*Iron Steel Inst. Special Rep.*, 1952, (43), 212–216; discussion, 331–336).—The prodn. of turbine blades, &c., by the lost-wax process is described, and the defects which may occur are discussed. An outline is given of the detn. of resistance to thermal shock of a few cast alloys.—D. M. P.

Investment-Casting of Nozzle Guide Vanes. H. E. Gresham and A. Dunlop (*Iron Steel Inst. Special Rep.*, 1952, (43), 209–212; discussion, 331–336).—A description of the casting of gas-turbine alloys by a development of the lost-wax process. Factors affecting dimensional accuracy are noted. The creep properties at high temp. of a few alloys produced in this way are tabulated.—D. M. P.

Investment Casting: Alloy Selection and Design. D. V. Ludwig and R. L. Wood (*Product Eng.*, 1951, **22**, (11), 185–190, 203–207).—See *M.A.*, **19**, 682.—M. A. H.

Investment-Casting Method for Superalloy Parts Lowers Costs, Improves Design, and Saves Material. Clarence E. Levoe (*Western Metals*, 1952, **10**, (7), 44–47).—A brief description is given of the process as applied to Ni-base Hastelloy and Co-base Stellite alloys.—T. G.

Special Techniques Used in Producing Extra-Large Investment Castings. —— (*Western Metals*, 1952, **10**, (9), 61–63).—Castings up to 15 lb. are produced by investment casting by a Californian firm. The mould with the investment is vibrated, so that air is removed and the mould is closely packed. After setting, the mould is fired at temp. up to 1750° F. (950° C.). Any residues from the wax are burnt off at this temp. The metal is forced into the mould either by vacuum, centrifugal force, or by air pressure or a combination of vacuum and air pressure.—T. G.

Mould Method Cuts Metal Casting Costs. Robert Reinhardt (*Western Metals*, 1952, **10**, (10), 43–45).—A brief description of the Croning process. See *M.A.*, **20**, 373.—T. G.

Impregnation Improves Casting Quality and Service Life. J. B. Cantwell (*Iron Age*, 1952, **170**, (12), 166–167).—A mixture of 500-mesh metallic flour in a Na_2SiO_4 carrier is introduced into porous castings after air has been exhausted from the voids under high vacuum. Then air at 100 lb./in.^{2} is applied. The castings thus invisibly impregnated below the surface will withstand high operating pressures, and working temp. are higher than when plastics are used for impregnation.—J. H. W.

Plastic-Impregnated Castings Are Improved by Nu-Line Through Proper Testing Methods. I. B. Herr (*Western Metals*, 1952, **10**, (10), 69–71). Three methods of testing castings for pressure-tightness are described.—T. G.

Technique for Vacuum Impregnation of Non-Ferrous Castings Is Used at Northrop [Anaheim, Calif.]. Frank Charity (*Western Metals*, 1952, **10**, (7), 48–49).—See *M.A.*, **20**, 374—T. G.

***Study of the Surface of Castings. IV.—Roughness of Green-Sand Mould and Casting. VI.—Considerations on the Sieving Method for Moulding Sand. VIII.—Considerations on Measuring Methods for Cast Surfaces. IX.—A Report on the Measurement of Ferrous Cast Surfaces.** Kazuo Katori, Tsuneyuki Okakura, and Kenji Hashimoto (*J. Mech. Lab.* (*Tokyo*), 1950, **4**, (3), 126–136; (4), 274–278; (8), 304–308; 1951, **5**, (1), 15–21).—[In Japanese]. [IV.—] The surface roughness of green sand moulds was measured by a light cross-section method and the results compared with the surface of the casting, which was measured by a tracer method. [VI.—] There are many problems and limitations in connection with sieving and testing methods for moulding sand. In order to obtain sands of various grain-sizes that were required for a study of casting surfaces, a sieving method was studied. The results are: (i) the sieving time required depends on the grain distribution, (ii) the opening is more important than the mesh, and (iii) the shape of sand particles must be considered when deciding the size of sieve opening. In general, the mould sand is so rounded or sub-angular that shorter openings of rectangular form can be adopted. [VIII.—] The light cross-section method and the tracer method have been used to measure the characteristics of a cast surface, and the results are compared and discussed. It was found that (i) accuracy decreases remarkably with roughness, (ii) the rougher the surface the greater the number of measurements that must be made, and (iii) there are no differences between the values obtained by the two methods. [IX.—] Results of measurements of the surface roughness of cast Fe specimens are reported. Among the conclusions are: (1) there is a reciprocal relation between the grain-size of the moulding sand and the surface roughness of the casting; (2) there is a proportional relation between sand permeability and the roughness of the casting; and (3) 10–15% clay in the moulding sand gives a finer cast surface.—AUTHORS.

***Metallurgy of Shell Moulding.** B. N. Ames, S. B. Donner, and N. A. Kahn (*Amer. Foundryman*, 1952, **21**, (1), 24–29).—Cast-to-shape test-bars and $10 \times 12 \times \frac{1}{8}$ in. test-plates were moulded by the shell process and by the standard green-sand method, and the soundness and the mech. properties of some typical non-ferrous and ferrous castings compared. The properties of shell-moulded castings were equal to, and frequently superior to, sand-moulded castings, mainly owing to the cleaner castings obtained. Shell moulds were also made by using different types of backing mixtures, viz. metallic shot or granulated refractories. Radiographic and structural examinations of castings suggested that shell moulds have smaller chilling power than green-sand moulds, but the nature of the backing mixture has no effect. An increase in the chilling power of shell moulds can be obtained, however, by using zircon instead of silica sand for compounding the shell-moulding mixture.—V. K.

Zirconite Sand in Foundry Practice. —— (*Light Metals*, 1951, **14**, (154), 50–52).—Phys. properties and average grain-size distribution of zirconite ($ZrSiO_4$) sand are given, and recommendations are made for its use in making up cores and moulds for Al-base and for Mg-base castings. Compn. of core mixes and of sand facings are tabulated. Zirconite flour with added mineral binder is used in washes for moulds and cores.—P. R.

19 — WORKING

Aluminium Die-Forging Design. III.—Standard Dimensional Tolerances. A. E. Favre and A. J. Orazem (*Product Eng.*, 1951, **22**, (1), 125–129).—Cf. *ibid.*, 1950, **21**, (9), 130; *M.A.*, **19**, 155. Design precautions to minimize the effects of shrinkage and warping on dimensional accuracy are discussed, and standard die-forging tolerances are given.—M. A. H.

Hot-Forming Practice [for Aluminium Alloys] at Northrop Aircraft.—I.–II. Gilbert C. Close (*Steel Processing*, 1952, **38**, (1), 23–24; (2), 76–79).—A report is given and discussed of data obtained by studying the hot forming of various Al and Mg alloys, particularly T5S-T6 on a prodn. basis. Various curves illustrate the effect of changing factors on the mech. properties of the metal, and comparison is made of the various methods of heating and their appn. —S. R. W.

Precision Machining of Large Aluminium Alloy Forgings Requires Extreme Care. John J. Locke (*Western Metals*, 1952, **10**, (1), 34–36).—Describes the problems encountered in machining a 3000-lb. Al-alloy forging to a finished weight of 250 lb.—T. G.

Abrasive Belts Taper Aluminium Plate. —— (*Iron Age*, 1952, **170**, (8), 130–131).—Cf. *M.A.*, **20**, 375. A huge abrasive belt has been developed in the U.S.A. for putting the taper and finish on Al plate for aircraft wings. As much as 0·01 in. can be removed in one pass from a 72-in.-wide plate. —J. H. W.

Processing the Clad Steels. Rick Mansell (*Steel Processing*, 1950, **36**, (12), 605–611, 645).—The indust. use, surface protection during handling, cutting, heat-treatment, fabrication, and finishing of clad steel are reviewed.—S. R. W.

***Deformation of Magnesium Alloys.** S. I. Gubkin and M. I. Zatulovsky (*Izvest. Akad. Nauk S.S.S.R.*, **1950**, [Tekhn.], (10), 1537–1549).—[In Russian]. MA-2 and MA-3 alloy billets were subjected to various degrees of preliminary deformation consisting of primary (linear) deformation by a press, and secondary (lateral) deformation by a hammer or press. Results indicate that in the case of secondary deformation the optimum degree is 0·7 min. Preliminary secondary deformation considerably enhances plasticity characteristics and extends the temp. range of deformation. A dimensioned diagram showing the deformation of the test-pieces and various forging-temp. curves are included, and the conditions under which hammer treatment is permissible are stated. The employment of a method in which the basic form change is produced in the press at 420°–340° (for MA-2) or 380°–340° C. (for MA-3) is described, and the exact shapes and dimensions are given of forgings produced in the press or by hammer at 230°–250° (for MA-2) and at 250°–280° C. (for MA-3). 6 ref.—H. W.

Titanium Fabrication Progress. James Joseph (*Light Metal Age*, 1952, **10**, (1/2), 10, 11, 30).—J. describes some experiences at North American Aviation, Inc., with the hot- and cold-working, and tapping of Ti and the RC 130A and RC 130B Ti alloy sheet and bar stock, resp.—H. A. H.

Titanium Takes the Test at Ryan Aeronautical. William S. Cockrell (*Western Metals*, 1951, **9**, (11), 38–39).—A discussion of experimental results obtained on the cold and hot forming of Ti in comparison with those on stainless steel.—T. G.

New Roll-Forming Techniques Facilitate High Production for Close Tolerances. John S. Liefeld (*Western Metals*, 1952, **10**, (4), 49–50).—Various new roll-forming machines are described for Al alloy, brass, bronze, Zn, Cu, and steel.—T. G.

[Rotary Extrusion of] Integral Finned Tubing for Heat-Exchangers. Walter P. Hill (*Product Eng.*, 1951, **22**, (9), 140–142).—A new "rotary extrusion" process produces finned tubing having integral helical fins, in Cu and Cu alloys, Al, steel, and bi-metal. Heat-transfer characteristics are described.—M. A. H.

Reducing Cost of Extrusions. F. McL. Mallett (*Product Eng.*, 1951, **22**, (6), 149–152).—A method of analysing the cost of extrusions by ref. to the "shape factor" (ratio of perimeter: area of section) is outlined, and other cost factors are considered.—M. A. H.

New Dinking Dies Lower Blanking Costs. Thomas A. Dickinson (*Steel Processing*, 1950, **36**, (1), 31–33).—A description of blanking or piercing dies—"dinking" dies—and their advantages over other methods of cutting precision blanks.—S. R. W.

Factors in Selecting a Drawing Lubricant. R. F. Johnston (*Steel Processing*, 1949, **35**, (1), 27–30, 43).—The two most important factors, pressure and max. temp. during drawing, are discussed, and the properties demanded of the lubricants are given. The appn. and advantages of colloidal graphite as a lubricant are described in some detail.—S. R. W.

***The Mechanism of Failure of Highly Alloyed Materials by Upsetting.** M. V. Rastegaev (*Izvest. Akad. Nauk S.S.S.R.*, **1950**, [Tekhn.], (8), 1183–1188).—[In Russian]. Tests were undertaken to ascertain the cause of failure of high-alloy steels when subjected to heat and pressure. A brittle steel test-piece 35 mm. in dia. was subjected to hammer forging and compression at various temp. It was found that in the process of irregular deformation the metal becomes heated unevenly. Upsetting by flat blows causes local overheating, resulting in the formation of a layer at points in the plane of max. displacement. The extent of such overheating ∝ the rate and extent of deformation and inversely ∝ the thermal conductivity of the steel treated. Local overheating of the layer causes local phase reactions when the metal as a whole is subjected to compression. In most high-alloy steels a phase reaction at high temp. is accompanied by the formation of chains of globules and films which break up the cryst. bond of the alloy and thus reduce its plasticity. In brittle high-alloy steels disturbance of the cryst. bond weakens the structure, causing cracks and ultimate failure. The result of flat blows on a round bar is diagrammatically shown. Photographs illustrate the destruction of test-pieces by forging; the microstructure of a high-alloy steel bar heated to seven temp. between 400° and 1270° C.; and the cracks in a bar at different stages of jumping. A table gives results of blows from a 90-kg. breaker on a 21-mm.-dia. test-piece at different temp.—H. W.

***The Effect of the Cutting Fluid Upon Chip/Tool Interface Temperatures.** M. C. Shaw, J. D. Pigott, and L. P. Richardson (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (1), 45–52; discussion, 53–56).—An analysis is presented of the short-circuiting effect of a cutting fluid in conjunction with the chip/tool interface temp. measurements by the tool/work thermocouple technique. An experimental arrangement based upon the results of this analysis and capable of giving reliable cutting temp. in the presence of fluids is described. The temp.-reducing characteristics of a representative group of water-base cutting fluids were studied, and these experiments revealed that the fluids become less effective in reducing the cutting temp. as the vol. of metal removed/unit time is increased. The water-base fluids considered are found to reduce the tool-tip temp. by a cooling action to a greater extent than by the reduction of a friction force on the face of the tool. —H. Pl.

Method X: Electro-Mechanical Machining. Malcolm F. Judkins (*Canad. Metals*, 1952, **15**, (6), 56, 58).—An electro-mech. machining process called Method X is described. Elect. forces are applied between the tool and work of sufficient magnitude to exceed the material's tensile strength, and rupture occurs. The tool is made the negative electrode, and penetrates the work without phys. contact, using a suitable liquid as electrolyte. The principal appn. of the process is for metals which cannot be machined by conventional methods, and the process can be used for drilling, tapping, internal and external shaping, and engraving. A machine for performing these operations is described.—W. A. M. P.

20 — CLEANING AND FINISHING

(See also " Protection " and " Electrodeposition ".)

Finishing Aluminium Castings. Evans Newcomb (*Canad. Metals*, 1952, **15**, (4), 30–31).—N. discusses the finishing of Al castings, using coated abrasive belts.—W. A. M. P.

Finishes for Aluminium Products. R. V. Vanden Berg (*Product Eng.*, 1951, **22**, (10), 179–186).—The characteristics, methods of prodn., and appn., of various electrochem., organic, ceramic, mech., and chem. finishes for Al and Al alloys are described and tabulated.—M. A. H.

How [U.S.] Navy in Seattle Refinishes Salt-Water Pitted [Aluminium-Alloy Aircraft] Propellers. Howard E. Jackson (*Western Metals*, 1952, **10**, (4), 49–50).—A detailed description of the cleaning and refinishing process, including inspection and testing.—T. G.

What Finish for Aluminium? —— (*Canad. Metals*, 1951, **14**, (7), 42).—A brief note on the cleaning and finishing of Al. —W. A. M. P.

Bright Dipping Brass Compacts. Marvin Rubinstein (*Products Finishing*, 1952, **16**, (5), 34–40, 42, 44, 46, 48).—The compacts are degreased, given a 30-sec. dip in hot potash, rinsed, and then bright dipped for 10–40 sec. in H_2SO_4 12, HNO_3 6, H_2O 1½ U.S. gal., HCl 3 fluid oz. A little wood soot is added to this soln. as a brightener. After treatment the work is rinsed in water, then in NaCN (4oz./U.S.gal.), and finally rinsed and treated for 15 sec. in a water-displacing fluid. —G. T. C.

Electrolytic and Chemical Finishes. —— (*Product Eng.*, 1951, **22**, (2), 141–148).—A comprehensive table gives the main uses, characteristics, properties, and techniques of appn. of the principal inorganic finishes for the major metals; and in addn. various inorganic finishes for sp. metals (Al, Cu alloys, Sn, Pb, ferrous metals, Mg, and Zn) are considered in some detail.—M. A. H.

Bicycle Finishing in Japan. Ezra A. Blount (*Products Finishing*, 1952, **16**, (9), 88–98).—Information is included on various cleaning and plating processes used and on the use of a surface conversion dip on bright Zn plate.—G. T. C.

Hand Finishing Cast Toys. —— (*Product Finishing* (*Lond.*), 1952, **5**, (4), 72–77).—The hand finishing of cast " tin " soldiers (Pb–Sb alloy) is described.—H. A. H.

Nikon [Camera-] Finishing Methods Stress Quality. Ezra A. Blount (*Products Finishing*, 1952, **16**, (10), 32–34, 36).—A brief review of finishing methods used in a camera factory in Japan.—G. T. C.

Typewriter Finishes. —— (*Product Finishing* (*Lond.*), 1952, **5**, (1), 40–44).—A very brief account of the Cu, black and white Ni plating, metal pretreatment, and painting processes used by Remington Rand Ltd.—H. A. H.

Organic Finishes for Metal Surfaces. —— (*Product Finishing* (*Lond.*), 1952, **5**, (2), 69–75).—A brief survey of the main types of organic finishes suitable for use on ferrous and non-ferrous surfaces.—H. A. H.

Selecting Organic Coatings for Metal Products. —— (*Product Eng.*, 1951, **22**, (12), 173–180).—Factors to be considered in selecting enamels, lacquers, and varnishes are discussed.—M. A. H.

Textured Organic Coatings. Lester Back (*Product Eng.*, 1951, **22**, (4), 129–136).—The characteristics, methods of appn., and uses of several new coatings with textured surfaces are described.—M. A. H.

Lacquering to Obtain Brilliance and Metallic Lustre. H. H. Vevers and G. E. Gardam (*J. Electrodepositors' Tech. Soc.*, *Advance Copy*, 1952, **28**, (5), 8 pp.).—The importance of brilliance and lustre on metallic objects is discussed, and a method is described for the detn. of graphs using a Guild reflectometer which enable the relative smoothing effect of various lacquers and various methods of appn. to be assessed. Types of lacquer are described both for base coats and for top coats, and the characteristics of various methods of appn. are also mentioned.—G. T. C.

Practical Considerations in the Application of Vacuum Coatings for Metal Finishing. L. Holland (*J. Electrodepositors' Tech. Soc.*, *Advance Copy*, 1952, **28**, (4), 12 pp.).—The process is briefly described, and details are presented of the equipment used. The use of the method to produce Al coatings and the properties and appn. of these coatings are described. The effects of contaminating vapours are also considered. 8 ref.—G. T. C.

How the Plating Industry Will Benefit from Research on Metal Cleaning. E. T. Candee (*Proc. Amer. Electroplaters' Soc.*, 1951, **38**, 181–189).—Suggestions are made for various lines of research into metal cleaning.—G. T. C.

A New Degreasing Evaluation Test: The Atomizer Test. H. B. Linford and E. B. Saubestre (*Proc. Amer. Electroplaters' Soc.*, 1951, **38**, 169–176; discussion, 177–179).—A report is presented on the atomizer test developed under Research Project No. 12 of the American Electroplaters' Society. The report is based on information already published. Cf. *Plating*, 1950, **37**, 1265; 1951, **38**, 60, 158, 254, 367, 713; *M.A.*, **18**, 477, 634, 778; **19**, 233.—G. T. C.

Painting Castings Electrostatically. —— (*Foundry*, 1952, **80**, (3), 240, 242).—A mechanized unit for the electrostatic painting of castings is described.—R. W. R.

A Metal-Cleaning Test Using Radioactive Stearic Acid as Soil.—I.–II. J. W. Hensley, H. A. Skinner, and H. R. Suter (*Metal Finishing*, 1952, **50**, (7), 49–52; (8), 77–80, 83).—Procedures are given for producing a uniform, reproducible, abraded surface on steel test-pieces and for applying very thin, uniform, and highly reproducible films of stearic acid. A special apparatus is used for preparing and cooling the test-pieces, and this is described. Detailed results are presented showing the removal from steel of the stearic acid soil by single-component alkaline cleaners with variations in time and concentration and by an organic detergent soln. with variation in time. When used for still cleaning at 90° C., distinct differences in the rate of soil removal and amount of residual soil left on the specimen were noted. With each cleaning soln. the residual soil on the test-piece tended to reach an equilibrium value after a certain period of cleaning and did not decrease further with increased cleaning time. With a particular alkaline soln., variation in concentration over the range 0·1–1·0% Na_2O had little effect on cleaning results (2 min. immersion at 90° C.), but increasing the concentration above 1·0% Na_2O caused a very pronounced decrease in cleaning efficiency. 5 ref.—G. T. C.

Metal Degreasing. J. Groome (*Metalen*, 1952, **7**, (17), 282–285).—[In English]. A survey of the main methods of degreasing metal surfaces. Mech., chem., soln., vapour, and boiling-liquor methods are compared. General hints are given on cleaning for electroplating and on the selection of C_2HCl_3 degreasing plant.—I. S. M.

Surface Preparation for Metallizing. Rick Mansell (*Steel Processing*, 1949, **35**, (5), 241–244, 267).—The following methods are briefly described and discussed: (1) blasting method; (2) rough threading method; (3) dove-tailing method; (4) elect. bonding method—a projected spot-welding machine with a high-amperage and low-voltage transformer is applied; the electrode contact heats up the metal surface, and the metal is then fused by a contacting movement; (5) Fusebond process—an elect. process in which the electrode metal applied to the base metal is exploded into a foam by heat produced by elect. resistance; and (6) Sprabond process in which a special metallic undercoat spray is first applied.—S. R. W.

Felt for Metal Finishing and Polishing. Leon D. Gruberg (*Metal Finishing*, 1952, **50**, (9), 58–63).—Information is presented on the use of felt in metal finishing. Particular ref. is made to the finishing of Al, Cu, Sn, and steel. Details are given on the manufacture, setting up, and care of felt wheels.—G. T. C.

Barrel Finishing. III.—Planning a Barrel-Finishing Department. Morris S. Shipley (*Plating*, 1952, **39**, (8), 860–861).—Cf. *ibid.*, (3), 257; *M.A.*, **19**, 882. A number of suggestions are made on various aspects of the planning of a barrel-finishing department.—G. T. C.

Characteristics of Industrial Primers. H. J. Testro (*Product Finishing* (*Lond.*), 1952, **5**, (5), 48–52).—A short survey of the essential characteristics of priming paints for use on wood or metal surfaces (both ferrous and non-ferrous).—H. A. H.

The Last Word?: Metal Finishing Productivity Report. —— (*Product Finishing* (*Lond.*), 1952, **5**, (2), 82–86).—A report of a meeting of the Institute of Metal Finishing at which the Anglo-American Productivity Report on Metal Finishing was discussed.—H. A. H.

Marking Processes.—XI.–XX. —— (*Product Finishing* (*Lond.*), 1951, **4**, (8), 82–90; (9), 75–83; (10), 83–91; (11), 83–92; (12), 84–94; 1952, **5**, (1), 63–69; (2), 88–98; (3), 78–85; (5), 79–83; (6), 57–61).—Cf. *ibid.*, 1951, **4**, (7), 82; *M.A.*, **19**, 235. [XI.—]. Al labels, name plates and designs. [XII.—] Printed metal panels and plates. [XIII.—] Sign-writing and other hand processes. [XIV.—] Cast metal name plates. [XV.—] Printed circuits and elect. conductive patterns. [XVI.—] Abrasive methods of marking. [XVII.—] Marking by pressure sensitive tapes. [XVIII.—] Markings and patterns on vitreous enamel. [XIX.—] Hand marking by machine. [XX.—] Direct photographic printing processes.—H. A. H.

Handbuch der Metallbeizerei. I.—Nichteisenmetalle. (Vogel). See col. 448.

---

21 — JOINING

Hard Soldering in an Electric Furnace with a Reducing Atmosphere. A. P. Gulyaev and I. I. Kobin (*Avtomobil. i Traktor. Prom.*, 1950, (8), 11–14; *C. Abs.*, 1951, **45**, 526).—[In Russian]. The strength of bonds produced by hard soldering exceeds the strength of pure Cu and may be taken as 15 kg./mm.^{2} for calculations. The use of peat gas is recommended for the reducing atmosphere.

Aluminium Brazing Saves Weight in Aircraft Parts. W. H. Perry (*Western Metals*, 1952, **10**, (9), 48–51).—T. G.

Brazing of Steel. I.—Elements of Brazing. II.—Physical Chemistry of Brazing. III.—Metallurgy of Brazing. S. L. Case (*Steel Processing*, 1950, **36**, (3), 125–129, 161; (4), 183–185, 199, 206; (5), 238–243, 257, 265).—[I.—] Definitions and appn. of torch, dip, elect., and furnace brazing, the forms of brazing alloy and their appn., the design of joint, surface prepn., and the use of flux are given. [II.—] Wetting of the base metal by the brazing alloy, the effect of surface condition on wettability and evaluation of wettability by the Kollagraph (Earle, *J. Inst. Metals*, 1945, **71**, 45; *M.A.*, **12**, 420) are described and discussed. Capillary-flow characteristics of brazing alloys, as well as the ability to wet and spread over the base metal are associated with surface tension. These are discussed, and the basic principles of two methods of measuring the surface tension of liq. metals are briefly described. [III.—] After defining the function of the flux the nature of the bond formed is discussed, and a review given of the results of various experiments that have been made to determine the effect of film thickness of filler metal and brazing temp. on the strength of the bond. The cause of voids, their detection, and the effect of their presence on the quality of brazed joints are also discussed.—S. R. W.

A New Method of Brazing and Soldering. Charles Boyden, Sr. (*Weld. J.*, 1950, **29**, (10), 907–909).—Directions are given for brazing or soldering with a " gun " fed with wire of suitable alloy at a const. selected speed.—P. R.

How to Bronze Weld. C. H. Wanamaker (*Weld. J.*, 1950, **29**, (3), 235–237).—Detailed directions for bronze welding are given; the importance of cleaning and of low heat is emphasized.—P. R.

A Comparison of Heating Methods for Brazing.—I.–II. Lester F. Spencer (*Steel Processing*, 1951, **37**, (12), 617–623; 1952, **38**, (1), 29–34).—The compn. and properties of brazing alloys are given, and methods of brazing by furnace, salt-bath, torch, induction, elect. resistance, and incandescent C heating are described and compared.—S. R. W.

Ravenna Metal Products Uses Rotating Fixtures to Speed Hand Silver Brazing. Howard E. Jackson (*Western Metals*, 1952, **10**, (5), 46–47).—T. G.

***Tests of Fillet Welds on Aluminium Alloy 61S-T6 Plate.** R. L. Moore and J. M. Pickett (*Weld. J.*, 1950, **29**, (5), 224S–230S).—Static tensile and shear tests were carried out on A-shielded W-arc transverse and longitudinal fillet welds in Al alloy 61S-T6 plate to determine (*a*) reduction in static tensile strength likely to result from fillet welding ⊥ direction of load; and (*b*) static shear strength of fillet welds. Results (tabulated and summarized graphically) indicate that weld size influences plate strengths, which were higher for $\frac{3}{16}$-in. than for $\frac{1}{4}$-in. fillets. Increased ratio of specimen width to weld width gave greater strength. Size of fillet and total length of weld should thus be kept to the min. consistent with shear strength and sound work. Transverse-welded specimens failed by ductile shear-type fractures at or near the edge of the weld. Measurements of elongation at the edges after failure indicated that discontinuities due to welds had little effect on the resistance to plastic deformation. The size of the weld did not appreciably affect the ultimate shear strength. The first yield in shear took place at $\sim\frac{1}{3}$ of load to failure, both values being slightly below those for specimens made by unshielded metal-arc welding.—P. R.

***Helium versus Argon for Welding of Aluminium.** A. J. Last (*Weld. J.*, 1950, **29**, (9), 770–771).—Operation records for continuous inert-arc welding of Al hot-water tanks showed that at resp. optimum adjustments He welding permits lower gas consumption, higher welding speed, and stronger, narrower, and better reinforced welds than with A. The latter is retained for manual welding because of the easier manipulation of the arc.—P. R.

How Giant-Size Aluminium Trays Were Fabricated. —— (*Weld. J.*, 1949, **28**, (12), 1173).—Dehydrating trays 22 ft. long for dried-egg manufacture were built up from 4 lengths of Al sheet by constructing halves by riveting, welding, and bonding, and welding the halves together in position. —P. R.

Ryan Simplifies Fabrication of Giant Wing Tank [in Aluminium Alloy] Requiring 30,000 Spot Welds. Joerg Litell (*Western Metals*, 1952, **10**, (5), 41–42).—Large external fuel tanks of Al alloy for the new Boeing B-478 are resistance welded. The equipment for welding and testing is described. —T. G.

Large Structures and Heavy-Gauge Metals [Aluminium Alloy and Steel] Joined at Ryan with Improved Spot Welding. —— (*Western Metals*, 1952, **10**, (1), 37–38).—A main feature of the equipment is the electronic control, which is briefly described.—T. G.

Some Basic Factors in Design for Aircraft Welding. J. Koziarski and K. B. Gillmore (*Weld. J.*, 1950, **29**, (9), 709–714).—Problems associated with fusion welding in aircraft construction include the mech. effects of the discontinuity caused by the weld and corrosion due to entrapped flux or incomplete filling. These can be minimized or obviated by suitable modifications in design, especially the type of joint; recommendations are made for the design of joints and tube-end fittings.—P. R.

***Observations of Electrode Tip Pick-Up and Tip Life in the Spot Welding of Magnesium Alloy Sheet.** W. F. Hess, T. B. Cameron, and R. A. Wyant (*Weld. J.*, 1947, **26**, (8), 433S–442S, 484S).—Factors influencing tip pick-up (fouling of electrodes by transfer of material being welded) are enumerated, and the development of pick-up during spot-welding of Mg–Mn alloy sheet is described. The life of Elkalloy tips used with stated types of welding is recorded, for 2 types of sheet of given gauge and surface condition, in terms of (*a*) number of welds to pick-up of Cu, and (*b*) number of welds to sticking and failure. A bench test for the presence of Cu is described. Pick-up is much less marked with Mg–Al–Zn sheet than with Mg–Mn alloys, the hardness of welded material possibly being an important factor. Results of earlier work are summarized, as are the effects of pre-cleaning method, weld size, direction of current, material treated, and other factors on pick-up.—P. R.

***Some Unusual Factors Encountered in Investigating Cracked Welds in 35 : 15 [Nickel–Chromium Steel] Magnesium Retorts.** H. J. Nichols (*Weld. J.*, 1947, 26, (10), 881–884).—A series of failures by cracking in the circumferential weld joining condenser to barrel in alloy-steel retorts for the prodn. of Mg is attributed after met. examination, chem. analysis, and measurements of service temp., to: (*a*) concentration of stresses in weld metal owing to the rigidity of the high-C alloy steel (Ni 35, Cr 15, C 0·4% min.) retort itself; (*b*) embrittlement of the weld through carbide pptn. at service temp.; and (*c*) faulty maintenance of the supporting system and consequent excessive stressing of the bottom of the retort.—P. R.

***Effect of Weld-Metal Composition on the Strength and Ductility of 35 : 15 Ni–Cr [Steel] Welds.** David Rozet, Hallock C. Campbell, and R. David Thomas, Jr. (*Weld. J.*, 1948, 27, (10), 481S–491S; discussion, (12), 610S–612S).—Low ductility and cracking of the weld metal in the arc-welding of 35 : 15 Ni–Cr steel with lime-coated electrodes is attributed to the formation of intergranular slag films when C exceeds, or Si is below, certain limits. Recommendations are made on the compn. of electrodes where high Si and C contents are desired, and the influence of S, P, Mn, and Nb on the strength of the weld metal is tabulated.—P. R.

Hints for White-Metal Welding. E. V. Olsen (*Weld. J.*, 1950, 29, (10), 904–907).—Detailed recommendations are made for the repair of white-metal castings by welding. —P. R.

Welding Galvanized Iron. F. J. Pilia (*Weld. J.*, 1950, 29, (6), 491–492).—Galvanized Fe can be fusion-welded or bronze-welded in the same way as uncoated material, but the presence of the Zn coating necessitates protection from ZnO fumes and the use of min. welding heat.—P. R.

Spot-Welding Galvanized Steel. M. L. Begeman, M. L. Hipple, and L. Cullum, Jr. (*Weld. J.*, 1949, 28, (9), 385S–395S).—Operating conditions for welding galvanized and ordinary steel are compared, major factors in the former process being the low m.p. of Zn, its affinity for Cu contained in electrodes, variations in thickness of the Zn coating, and the phys. properties of Fe–Zn compound. The influence of the following factors in spot-welding galvanized sheet have been studied at the University of Texas: cooling of electrodes, welding current, welding period, tip pressure, and nature of material being welded. Effects on the strength of the weld are tabulated, and characteristic structures are described. Increased welding current increases the weld strength; tip life is prolonged by using refrigerated coolant, but the accompanying increase in strength is negligible, and the method is slow and costly. High welding speeds and specially designed electrodes are desirable. 17 ref.—P. R.

***Electrode Tip Life in Spot-Welding Galvanized Steel.** Frank W. McBee, Jr. (*Weld. J.*, 1950, 29, (10), 531S–535S).—Spot welds were made in galvanized steel sheet with a commercial welding machine under simulated indust. conditions to determine the factors influencing the life of electrode tips and the quality of the weld. Domed or truncated-cone tips were used (*ibid.*, 1949, 28, 385S; 1950, 29, (1), 31S). Increased contact area obtained with the latter type caused decreased c.d. and lower weld strength. Poor appearance and strength of the weld were mainly due to deformation of the tip, which should preferably be of hard material. Pick-up of Zn was without appreciable ill-effects, and sometimes limited the loss of coating. Flashing and sticking were due to use of excessive energy rather than to the number of welds made. The shape of the tip should remain const. and the length of stroke should be kept to a min.—P. R.

***Electrodes for Spot-Welding Galvanized Steel.** Harry B. Spore (*Weld. J.*, 1950, 29, (1), 31S–36S).—The influence of electrode material and tip shape on the phys. properties of spot welds in galvanized steel was studied with 3 forms of tip and 4 compn. of electrode under fully stated conditions. Optimum machine settings and welding times are quoted for resp. shapes; tips of 2-in. radius, requiring min. time for satisfactory performance, are considered preferable to pointed, domed, or flat tips. Welding current/weld strength graphs for the different electrode materials used show little difference between the 4 types.—P. R.

Stud Welding to Ferrous and Non-Ferrous Metals. T. L. Hulfert, P. F. Powder, and J. Nater (*Product Eng.*, 1951, 22, (3), 89–91).—A description of the method and its fields of appn. is given in the form of answers to questions likely to arise when considering its use.—M. A. H.

Resistance Welding of Jet Engines. H. E. Lardge (*Weld. J.*, 1949, 28, (3), 249–254).—Development of welding technique as applied to the prodn. of jet engines is described, with numerous examples. Phys. and mech. properties influencing welding performance are tabulated for mild steel, austenitic 18 : 8 steel, Inconel, Nimonic, and a high-alloy steel.—P. R.

Welding and Low-Temperature Brazing of Air-Conditioning and Refrigeration Parts. Ward Swarthout (*Weld. J.*, 1948, 27, (7), 511–516).—Intricate design and arrangement of parts, and character of materials used, make oxy-acetylene welding and brazing particularly suitable for fabrication of air-conditioning and refrigerator plant. Appn. are described and illustrated, with details of support, welding rod, brazing alloy, flux, and gas supply.—P. R.

***Weld-Metal Properties at Low Temperatures.** T. N. Armstrong and W. L. Warner (*Weld. J.*, 1950, 29, (6), 295S–302S).—Impact (Charpy) tests at several temp. from −238° F. (−150° C.) to +175° F. (80° C.) were carried out on butt welds made under stated conditions with 9 types of ferritic Ni steel electrodes, a 25 : 20 stainless steel, a Ni–Cr–Fe–Nb alloy, and a special Monel for welding steel. Results of preliminary tensile tests are described and tabulated. Certain welds were also subjected to cracking sensitivity tests, as used by the U.S. Navy Department. Results are shown graphically, and impact values at stated temp. are tabulated, together with the temp. at which the energy to fracture was 15 ft.-lb. Low-temp. properties adequate to most commercial requirements are obtainable at temp. down to −150° F. (−100° C.) with many low-Ni steel electrodes. The marked differences between various types are attributable to welding current, &c., rather than to compn. Low-temp. properties of welds examined were not appreciably improved by preliminary stress-relieving treatment. Impact strengths of 2 high-Ni alloy deposits were satisfactory down to −320° F. (−195° C.). —P. R.

Heliarc and Railroad Applications. H. E. Gannett (*Weld. J.*, 1949, 28, (12), 1147–1151).—Inert-gas arc welding, preferably under A, ensures fluidity and high capillarity of molten metal and obviates dipping the filler rod deeply into the melt. The process is therefore particularly suitable for welding cast Fe or Al, and its use on Diesel railway engine and coach parts in these metals and in high-alloy steel is described.—P. R.

***Preliminary Results of Welding-Peening Investigation.** J. Lyell Wilson (*Weld. J.*, 1950, 29, (3), 97S–108S).—Factors examined included the weight of hammer, adjustment of air pressure, and temp. at which treatment is applied. Calibration of equipment is described. The riding load on the hammer may have an effect equal to that of air pressure, but the latter is a major factor in hammer performance, accurate pressure control hence being essential. Relationship between riding load, mass of the hammer, and mass of piston is important, particularly as influencing natural frequencies in the system. Recommendations are made for hammer loads and air pressure for certain conditions.—P. R.

***Radiography of Spot Welds in Various Sheet Gauges and Dissimilar Gauge Combinations.** (McMaster, Lindvall, and Gaard). See col. 433.

***Stress Distribution Round Spot Welds.** A. O. Bergholm, P. W. Swartz, and G. S. Hoell (*Weld. J.*, 1950, 29, (5), 217S–223S).—Flat bars of low-C steel with 2 or 3 spot welds were subjected to tensile tests to explore the stress distribution round the welds. The direction of the principal stresses was studied on photo-elastic models, observations on "stresscoat" enamel coatings supplying some confirmation. Strain-gauge readings are tabulated. Single laps of spot-welded joints are shown to be inherently weak, since bending of plates is unavoidable; double laps reduce high stress concentrations at welds. Stresses may be relieved by drilling holes near the weld. For important structures carrying greatly varying loads or liable to stress reversals, a single row of spot welds is undesirable.—P. R.

Metal Spraying in the Reclamation of Worn Engine Parts. Donald E. Harris (*Australasian Eng.*, 1952, (July), 58–61; discussion, 61).—Spraying methods are briefly reviewed. Zn and Al are sometimes used for reclamation work, but hard steel is more common.—T. A. H.

Metallizing of Packing Areas. J. E. Wakefield (*Weld. J.*, 1949, 28, (9), 875–876).—Worn packing areas on pump shafts, rams, &c., can be effectively repaired by metallizing if the applied metal has adequate resistance to wear and corrosion. Repairs to bronze pump-sleeves, turbine shafting, and hydraulic press rams are described.—P. R.

New Production Applications of Hard-Facing. E. C. Hurt (*Weld. J.*, 1950, 29, (12), 1082–1086).—Inert-gas shielded-arc welding with automatic or semi-automatic equipment, giving satisfactory bond with min. dilution, has extended the appn. of hard-facing. Examples include the treatment of valves for steam, aircraft and automobile engines, pump fittings, conveyor surfaces, and die faces. Equipment is described and illustrated.—P. R.

Automatic Electric Hard-Facing. Turner G. Brashear, Jr. (*Weld. J.*, 1950, 29, (5), 391–398).—Hard-facing with automatic welding equipment is economical because of its high speed, soundness of deposit, continuous operation, and possibility of using semi-skilled labour. Coiled welding rod of various compn. is available. Factors influencing choice of equipment and of flux are enumerated, and several appn. are described. Recommendations are made on bonding difficult materials with stainless steel and on controlling penetration and dilution.—P. R.

Developments in Automatic Hard-Facing. H. W. Sharp (*Weld. J.*, 1949, 28, (11), 1037–1039).—Worn or corroded parts can be repaired by automatic hard-facing with tubular alloy wire, produced by adding granular hard-alloy ingredients to mild-steel strip. Many electrode alloys are now available as hard-facing wire; ranges of alloy content and Rockwell C hardness are given.—P. R.

Hard-Facing Applications in the Steel Industry. J. J. Barry (*Weld. J.*, 1950, 29, (3), 198–203).—Selection of hard-facing alloys depends on accurate knowledge of optimum conditions, which may include abrasion, impact, heat and/or corrosion (see *ibid.*, 1949, 28, 31; abstract below). Hard-facing materials are classified as: (*a*) ferrous alloys, suitable for most purposes; (*b*) non-ferrous alloys, mainly Co-base materials resisting wear at high temp.; and (*c*) composite WC prepn. conferring high abrasion-resistance. Appn. of each group are described.—P. R.

The Economics of Hard-Facing. J. J. Barry and Albert Muller (*Weld. J.*, 1949, 28, (1), 31–37).—Optimum results from hard-facing (maintenance of dimensions, longer service, reduction in cost because process is applicable to cheap base materials, repairs are facilitated, and spares can be kept to a min.) depend on full understanding of service conditions, leading to selection of the right surfacing material. Recommendations are made for resistance to abrasion, impact, heat, corrosion, and erosion, and for the most economical use of the method. Appn. are described and illustrated.—P. R.

Hard-Facing with Inert-Gas-Arc Welding. K. H. Koopman (*Weld. J.*, 1949, 28, (1), 46–52).—Hard-facing by inert-gas shielded-arc welding and with various materials was carried out under stated conditions on mild steel, cast Fe, stainless steel, Cu, and Cu-base alloys. Quality of deposits usually approached that obtained by oxy-acetylene welding, and distortion was less. Forehand position was best for large areas and massive deposits, and backhand for smaller work. Recommendations are made on optimum A flow.—P. R.

Hard-Surfacing of Cast-Steel Propeller Blades. K. B. Young, H. J. Nichols, and M. J. Nolan (*Weld. J.*, 1949, 28, (2), 153–157).—Performance of cast-steel propeller blades hard-surfaced with Co–Cr–W alloy is compared with that of unsurfaced and of Pb-wiped blades. Apart from pitting of the steel near the boundary of the second coating, and some local pitting, at junctions of successive coatings applied at different times, the coatings and the leading and trailing edges of the hard-surfaced blades were in good condition after 2 to 3 years' service at different sea-water temp. One Pb-wiped coat had disappeared after 31 months, a Sn-base alloy coat being in good condition after 15 months. Unsurfaced blades were rapidly attacked. For hard-facing, the atomic-H process is preferred to use of oxy-acetylene.—P. R.

Reclamation of Diesel Locomotive Parts. La Motte Grover and R. L. Rex (*Weld. J.*, 1950, 29, (11), 973–980).—Many railway authorities in the U.S.A. maintain specially equipped units for welding or hard-facing worn or damaged Diesel-engine parts (*ibid.*, 1948, 27, 1034; *M.A.*, 20, 301). Detailed accounts are given of repairs to Al pistons, cast Fe cylinder heads and liners, parts of bearing assemblies, valves, bodies, framework, &c., with reasons for choice of resp. procedures.—P. R.

How Dodge Hard-Faces Gear Forging Dies. John C. McComb (*Steel Processing*, 1951, 37, (5), 231, 257).—A brief description of a process by which a hard surface $\frac{1}{16}$ in. thick is prepared by welding alloys contg. approx. C 1, Cr 28, W 4·5, and Co 62% on to the surface.—S. R. W.

25 — BOOK REVIEWS

Imperfections in Nearly Perfect Crystals. Edited by W. Shockley, J. H. Hollomon, R. Maurer, and F. Seitz. A Symposium sponsored by the Committee on Solids, Division of Physical Sciences, National Research Council, and held at Pocono Manor, 12–14 October 1950. 8½ × 5½ in. Pp. xii + 490, illustrated. 1952. New York: John Wiley and Sons, Inc. ($7.50); London: Chapman and Hall, Ltd. (60*s.*)

The theme of this symposium, contained in the title "Imperfections in Nearly Perfect Crystals", derives from the realization that many properties of crystals are largely due to, or greatly influenced by, small deviations from the structure which idealized perfect crystals have. Accordingly, the seventeen papers in this book discuss the imperfections that have been suggested to exist, the experimental evidence for the different types, and their effects on properties, the second two aspects in particular overlapping. Although the book has a general title, it deals very largely with metals; two chapters are devoted to ionic crystals, and other types of crystalline solid receive only passing mention. It follows that the imperfections which receive most attention are dislocations, grain boundaries, and vacant lattice sites. Although some papers include new material, the main purpose of the symposium was evidently to bring together information which already existed but which was scattered throughout the literature, and to subject this material to critical examination.

The book is divided into four main sections. The first deals with "The Nature of Imperfections" and contains a paper by F. Seitz, the longest in the book, which attempts to consider all the various main types of imperfection and their interaction with one another. It is not clear that the importance of the conclusions reached is commensurate with the space taken up, but one notes the sense of realism which includes impurity atoms in the list of six "primary imperfections". The only other paper in this section is by W. T. Reid and W. Shockley and gives a short account of the geometry of dislocations.

Section II contains five papers dealing with "The Role of Imperfections in Deformation", the imperfections here being mainly dislocations. The paper by N. F. Mott discusses mechanical strength in terms of dislocation theory; the portion dealing with the strength of solid solutions contains a new treatment of this problem which starts from different assumptions from those made in an earlier treatment by Mott and Nabarro. No reason is given for putting forward the new treatment in preference to the former one, and the reviewer was therefore intrigued to find on comparing the predictions of the two with some experimental data which chanced to be at hand that the new treatment gave considerably better agreement. The interpretation of X-ray data on cold-worked metals is considered in two papers from widely different points of view. Other papers in this section deal with slip lines and the influence of dislocations on damping and on elastic constants.

In Section III are two papers on ionic crystals and two on diffusion. That by J. Bardeen and C. Herring discusses diffusion from the point of view of thermodynamics, and that by C. Zener discusses the same problem from the point of view of statistical mechanics; it is a pity that the editors found no way of comparing the conclusions of these two treatments in the same terms. Section IV contains six papers dealing with external surfaces and grain boundaries. The data on surface and grain-boundary energies is critically reviewed by J. C. Fisher and C. G. Dunn, who most helpfully give a list of "best" values. Of the other papers in this section, that by A. Guinier on "Substructures in Crystals" in one sense is the most complete in the book, for it fits the experimental data with perfect clarity into one consistent framework. It remains to add that not the least valuable part of the book is the discussion which follows most papers. The remark can safely be made that the conclusions reached in any paper should not be accepted until the discussion has been read!

In such a book there are bound to be omissions. For example, two subjects which qualify for inclusion are the problem of fracture—fracture may always, and sometimes does, start at imperfections—which is not discussed at all, and the problem of the influence of grain boundaries on mechanical properties, the various ramifications of which are not gone into. Other points to note are that in one or two papers there are misprints, such as a reference in the text to something not present in a diagram, which will cause some difficulty; some authors collect their references at the end of the paper, others put them at the bottom of the page; of the 20 authors, 1 is French, 2 English, and 17 American, and the references accordingly have an American bias.

It will by now be clear who is likely to be interested in this book. It will be extremely valuable both as reference book and source book to research workers in a variety of subjects, and some parts of it should be useful to students. Since, moreover, it is probably the only book of such comprehensiveness in a field that is becoming more and more important, some others may read it for general information.

D. McLean.

Ternäre Legierungen des Aluminiums. Beispiele für die Kristallisation ternärer Legierungen. Von Heinrich Hanemann and Angelica Schrader. (Atlas Metallographicus, Band 3, Teil 2.) 29 × 21 cm. Pp. 170, with 140 illustrations and 502 photomicrographs. 1952. Düsseldorf: Verlag Stahleisen m.b.H. (Geb., DM 80.—.)

This book is not merely a review of published equilibrium diagrams of aluminium-rich ternary alloys: much of it deals with the results of original work carried out in the laboratories of the Technische Hochschule, Berlin. It opens with a section, some 40 pages long, describing the phenomena of solidification in ternary alloys rich in aluminium, polishing and etching methods that have been found satisfactory, and the appearance of the intermetallic phases occurring. Next follows a detailed account of 38 ternary systems, in which earlier work is reviewed in the light of the authors' own investigations. The published diagrams include, where possible, the liquidus surface isothermals, selected vertical sections through the ternary models, and selected horizontal (isothermal) sections. The literature has been surveyed up to 1951, and as far as the reviewer has been able to judge, no major publication has been overlooked: some 220 references are included.

The book concludes with 55 double-page tables and plates, consisting of selected micrographs of the various systems, with descriptions of the structures. The quality of the micrographs is superlative and the reproduction excellent.

H. W. L. Phillips.

Strength of Materials. By Arthur Morley. Tenth edition. 8½ × 5½ in. Pp. x + 583, with 265 illustrations. 1952. London: Longmans, Green and Co. (25*s*.)

This famous text-book, which has steadily maintained its reputation amongst engineering students, has now reached its tenth edition. It is unnecessary to comment generally on a book which has so signally stood the test of time, but the reviewer may profitably examine the revisions which have been made in this edition.

The main changes are that a section has been added on resistance strain gauges; an appendix deals with the analysis of strain, with particular reference to strain rosettes; and a further appendix, on close-coiled helical springs, supplements an earlier chapter on this subject. Apart from a few minor changes elsewhere, these are the only real additions, and it is a little disappointing that the opportunity was not taken to bring several other topics up to date.

The section on resistance strain gauges is too short to be helpful, and no references are given to books or papers containing details of practical experience in their use. The whole chapter on testing technique and extensometers might well have been revised: there are very few references later than 1912, and some progress has perhaps been made in the intervening years.

Hugh Ford.

Handbuch der Metallbeizerei. Band I. Nichteisenmetalle. Von Otto Vogel, neubearbeitet von Hermine Vogel *et al.* Zweite Auflage. 27 × 19 cm. Pp. xv + 410, with 263 illustrations. 1951. Weinheim/Bergstrasse: Verlag Chemie. (Geb., DM 49.20.)

This book is a comprehensive treatise on all aspects of the pickling and etching of non-ferrous metals. The historical development of the subject is first outlined, and there is a chapter which deals in considerable detail with the equipment needed and with the arrangement, ventilation, &c. of the building in which these processes are carried out. Information is given on disposal of waste liquors and on hazards to health. A chapter deals with the preparation of metals before etching and includes information on the use of protective atmospheres to avoid scale formation during heat-treatment and on degreasing by all the usual methods. Several chapters are devoted to detailed recipes for the treatment of individual metals and alloys, and there is one which deals solely with the treatment of antiquities, &c. The book is well indexed, and provides extensive literature references.

G. T. Colegate.

APPOINTMENTS VACANT

ABSTRACTOR required to compile and edit the Metal Powder Report published by Powder Metallurgy, Ltd. Some technical knowledge of the subject desirable. Must be able to prepare clear, well-written abstracts from technical articles published in English, German, and preferably one other language. Salary according to experience. Pension scheme. Write giving full details, age, education, experience to date, and salary required to Powder Metallurgy, Ltd., (Ref. DW), 59/62 High Holborn, London, W.C.1.

METALLURGIST of B.Sc. standard or equivalent required for laboratory investigations. Aged 20 to 30. Applicant required to work in West London to start and transfer to Maidenhead at a later date. Canteen. Profit-Sharing Scheme. Write age, experience, and salary required. Box 347, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

SENIOR EXPERIENCED ANALYTICAL CHEMIST required for chemical section of laboratory (London) dealing with metals, ores, cement, concrete, fuels, and engineering materials in general. Applications giving age, qualifications, experience, and salary required to Box No. 348, Institute of Metals, 4, Grosvenor Gardens, London, S.W.1.

THE BRITISH CAST IRON RESEARCH ASSOCIATION has a vacancy on its research staff for an investigator for the determination of solidification sequences in cast iron by temperature-gradient measurements, as part of an investigation on the soundness of iron castings. Applicants should possess an honours degree or its equivalent, and some previous research and/or industrial experience would be an advantage. The appointment will be in the Scientific Officer (£418–£676 per annum) or Senior Scientific Officer (£785–£890 per annum) grade, according to age, experience, and qualifications. Applicants should submit full details of qualifications, age, and experience to the British Cast Iron Research Association, Alvechurch, Birmingham.

ZEITSCHRIFT FÜR METALLKUNDE

Official Organ of the Deutsche Gesellschaft für Metallkunde e.V.

This Journal, now in its 42nd annual volume, is the official organ of the Deutsche Gesellschaft für Metallkunde, and as such it is the foremost medium of publication for German metallurgical research.

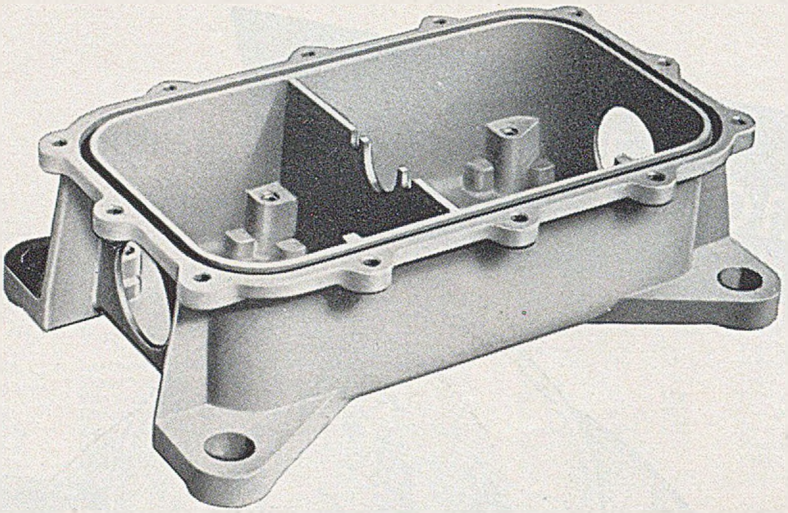
It consists entirely of papers describing original work carried out in the Universities, the Max-Planck Institut, and industrial research establishments.

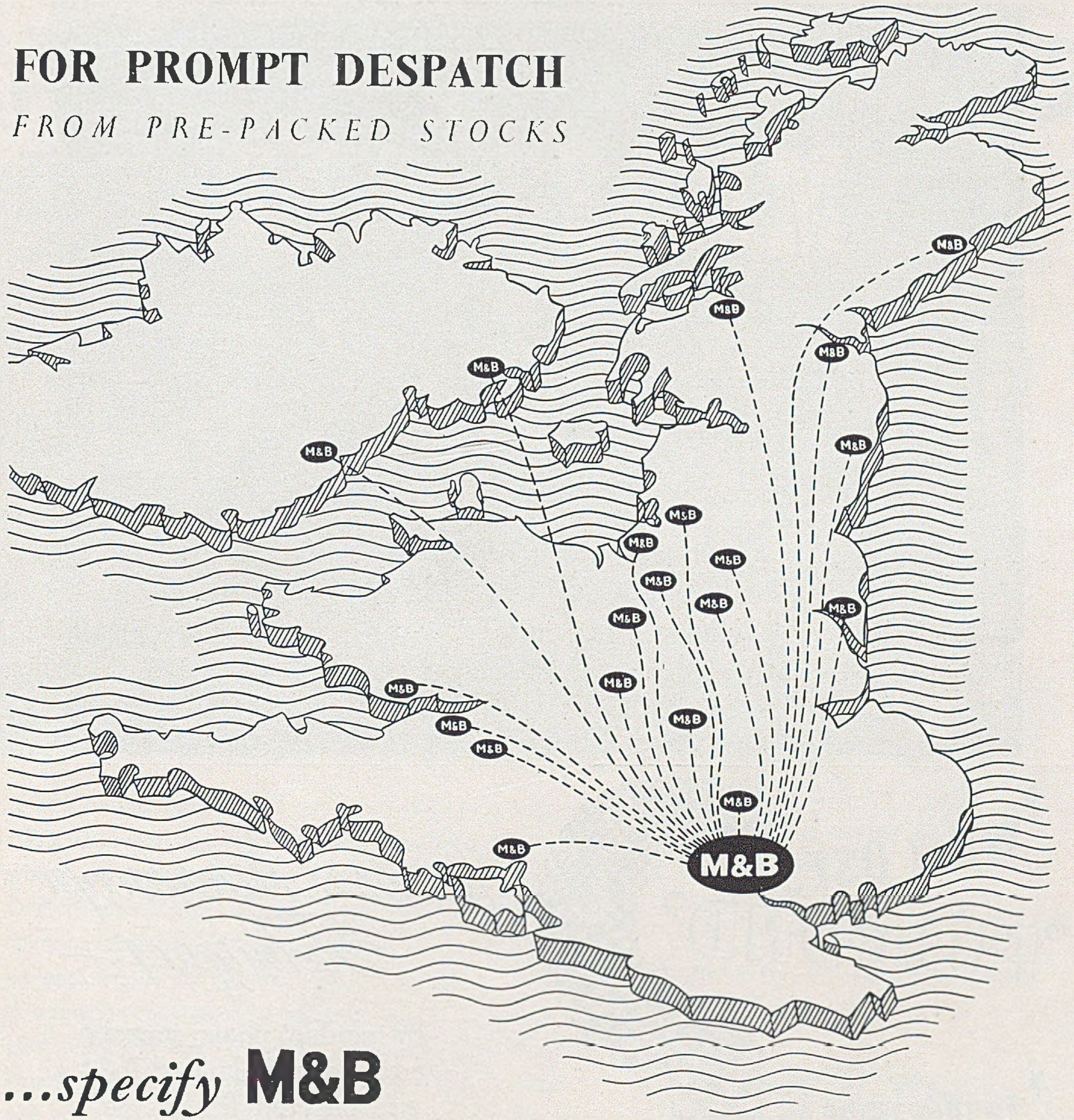
Price per issue DM 4.–

Annual Subscription (12 issues) DM 48.– Postage and packing extra

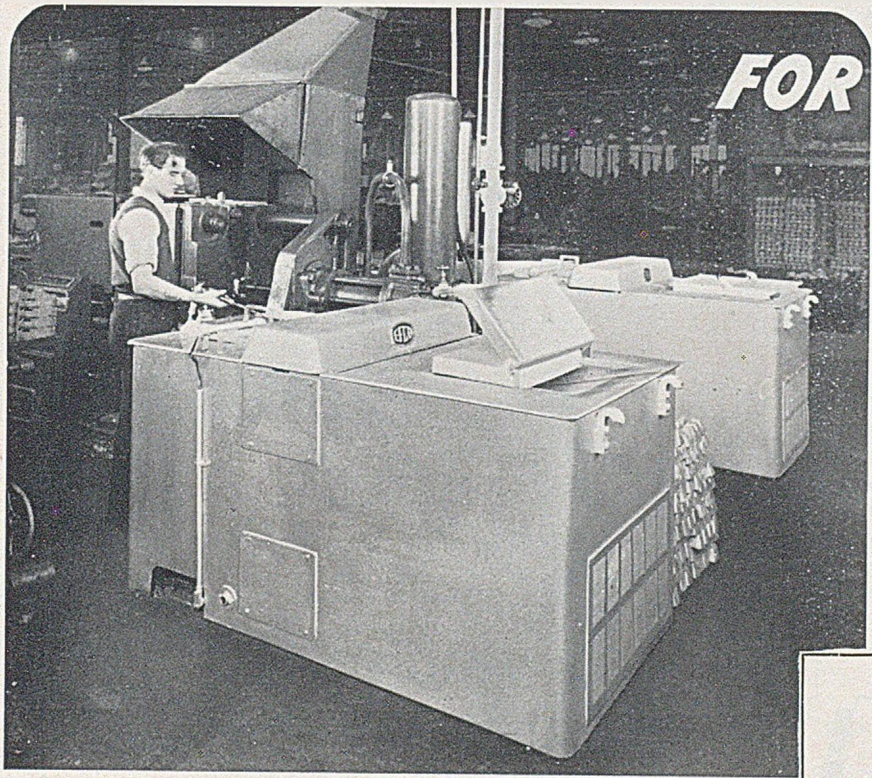
DR. RIEDERER-VERLAG G.m.b.H., STUTTGART S.

METALLURGICAL ABSTRACTS AUTHOR INDEX




M&B

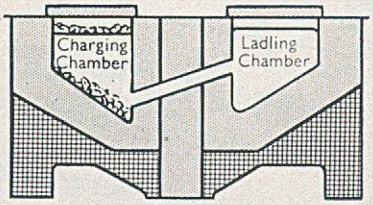
ALUMINIUM MELTING



FOR DIE CASTING

A MELTING AND HOLDING FURNACE IN ONE UNIT

Left
EFCO-Lindberg Aluminium Melting Furnaces. Illustration by courtesy of The British Thomson-Houston Co. Ltd.



Schematic diagram showing relationship between the two chambers.

A new design with a difference is the keynote to this EFCO-Lindberg melting furnace. It has two chambers which are connected by induction melting channels. One chamber is for receiving ingots and known as the charging chamber and the other which contains only molten metal is the ladling or pouring chamber. Use of this furnace eliminates transferring metal from melting unit to holding unit. It can, of course, be charged with liquid metal thus greatly increasing the output. You will have cleaner molten metal, less scrap loss, less maintenance. Running costs are low and there is no freeze-up or overheating. It is available in three sizes 40kW 200 lb. per hour, 80kW 400 lb. per hour, 120kW 600 lb. per hour.

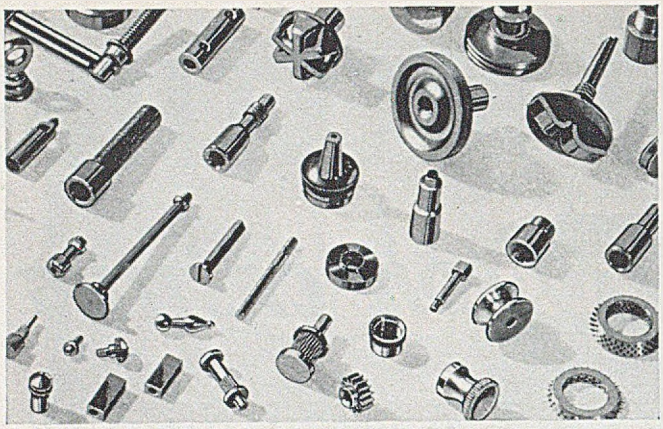
EFCO

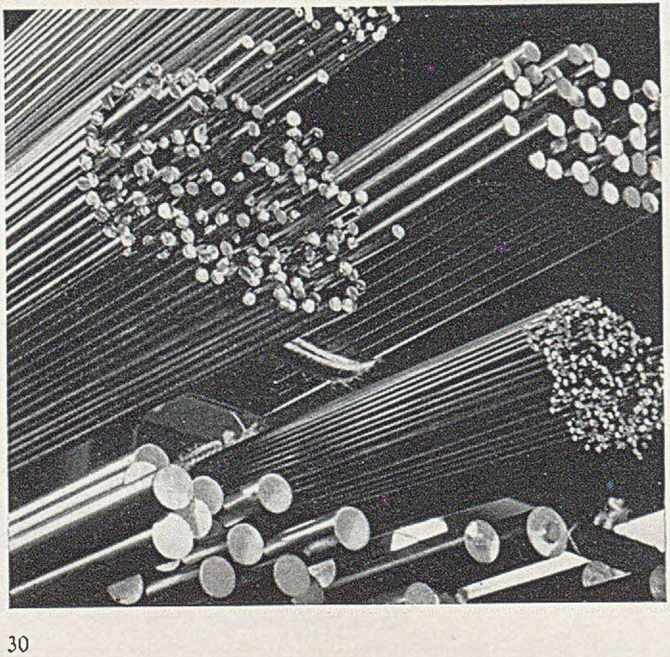
LINDBERG

Write for further details of this new design induction furnace

ELECTRIC FURNACE CO. LTD., Netherby, Queens Road, Weybridge, Surrey

Associated Companies: Electric Resistance Furnace Co. Ltd., Electro-Chemical Engineering Co. Ltd. Telephone: Weybridge 3891





Index to Advertisers

BETTER WELDS at lower cost

FAR LESS TIME AND MATERIALS REQUIRED

Much of the advantage gained by the use of high quality welding equipment can be thrown away—unless you also use top-grade rods and fluxes. ALDA rods and fluxes are the finest you can get. They include only those materials prepared to the formulæ of B.O.C metallurgists, or those proved by repeated tests to fulfil their specific purpose. All consignments are analysed before issue. As a result, you can rely on ALDA rods and fluxes every time—for better welds, made more quickly, and with the minimum consumption of essential materials. Use ALDA rods and fluxes—and *get* efficiency with economy.

FREE

. . . and very useful indeed!

The comprehensive Welding Guide and ALDA Data Sheet provides a list of 54 different rods and 21 different fluxes with full details of their applications and characteristics.
Also: Six different methods of removing flux residue, according to the types of metal involved.

WRITE NOW TO YOUR LOCAL B.O.C BRANCH FOR YOUR COPY

ALDA

WELDING RODS AND FLUXES

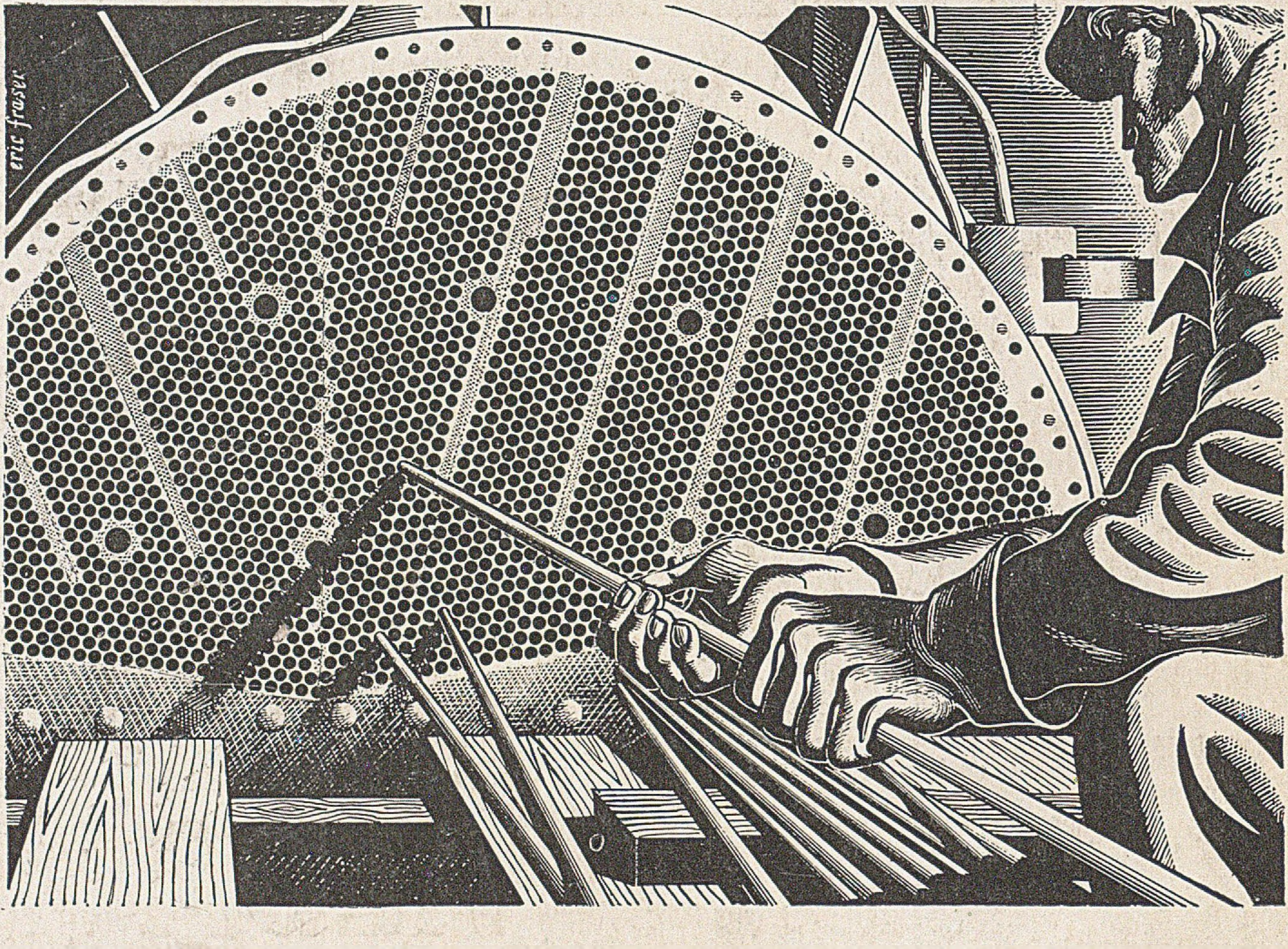
BOC **THE BRITISH OXYGEN CO LTD**

LONDON AND BRANCHES

NON-FERROUS

PLATE

and condensers



Tube plates fitted to condensers in marine and land installations must offer high resistance to corrosion and erosion. The Metals Division of I.C.I. manufactures condenser plates to customers' requirements in Naval Brass, Muntz Metal, 70/30 Brass, Aluminium Brass and Cupro-nickel, in accordance with any of the British, American or other recognised standards.

The consistently high demand for I.C.I. condenser plates and tubes is ample evidence of their quality and reliability—the outcome of many years' production experience, coupled with rigid technical control. The resources of the Division are freely available to help in the solution of technical problems concerning the use of such materials.

IMPERIAL CHEMICAL INDUSTRIES LIMITED, LONDON, S.W.1

M.193

ALL COMMUNICATIONS REGARDING ADVERTISEMENTS SHOULD BE ADDRESSED TO THE ADVERTISEMENT DEPARTMENT, 4 GROSVENOR GARDENS, LONDON, S.W.1.

PRINTED IN GREAT BRITAIN BY RICHARD CLAY AND COMPANY, LTD., BUNGAY, SUFFOLK.