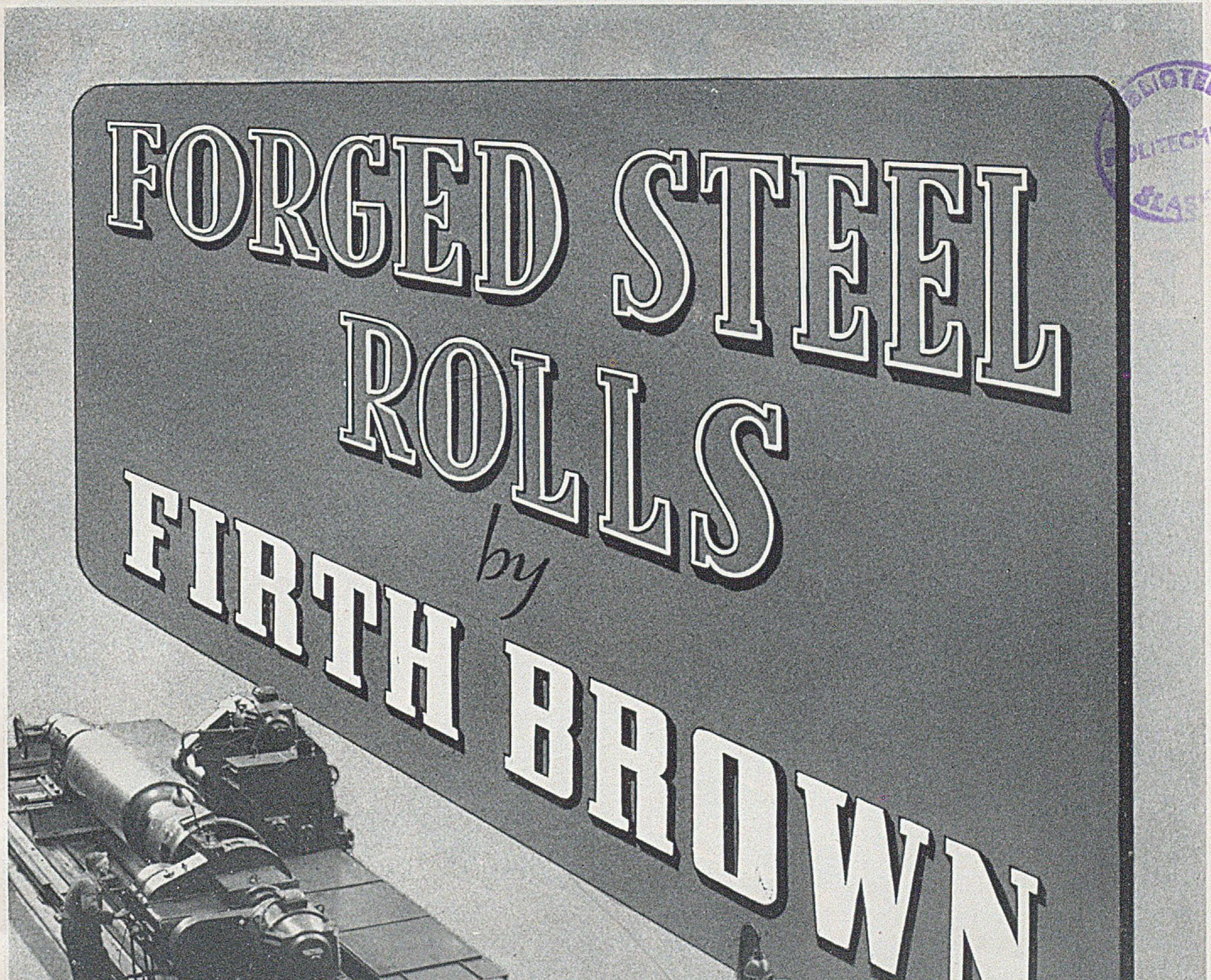
THE JOURNAL OF THE INSTITUTE OF Metals

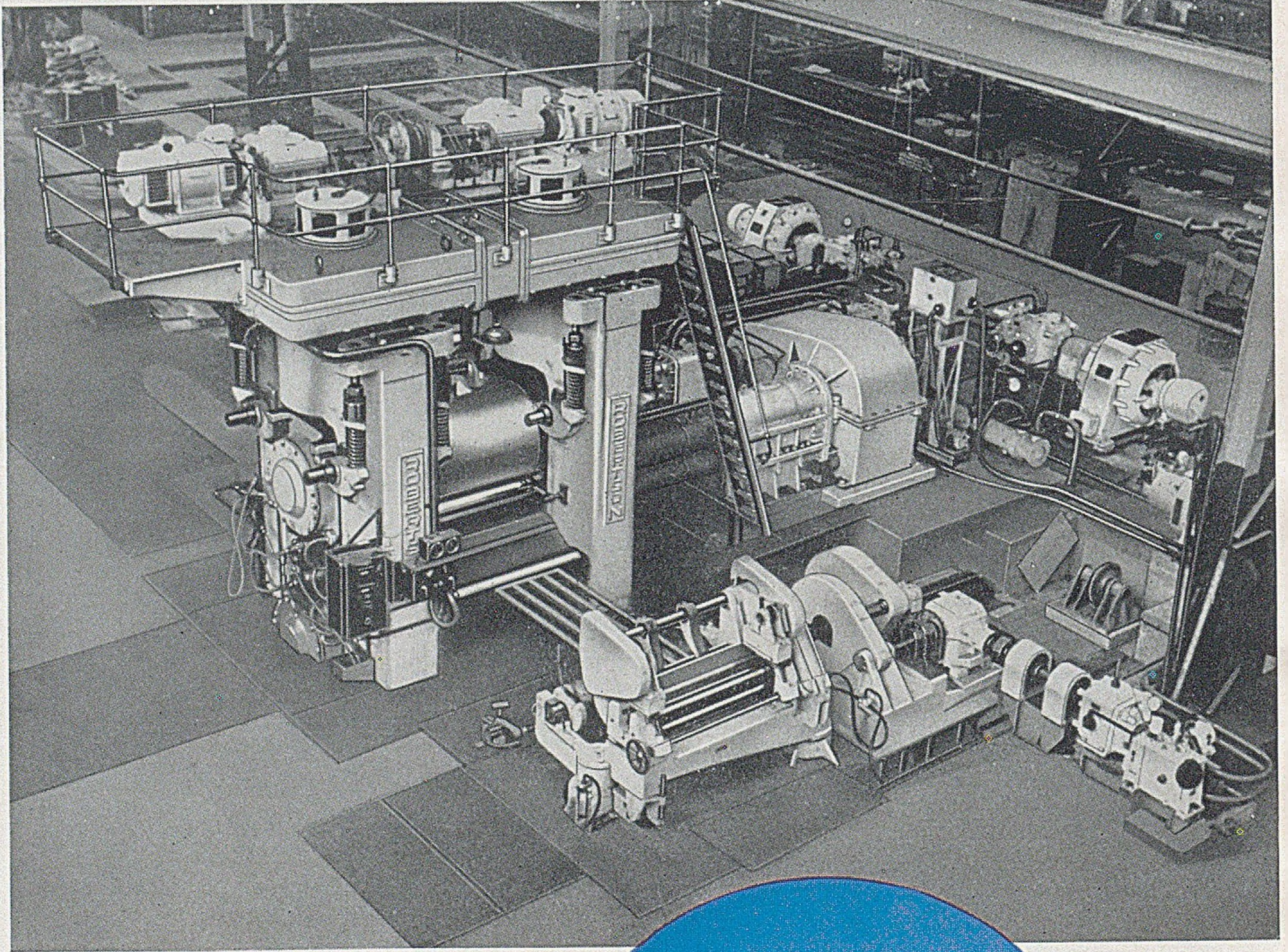
WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents:

PRICE: TEN SHILLINGS AND SIXPENCE

JANUARY 1953


FORGED STEEL ROLLS
by
FIRTH BROWN



Just Published

ATOMIC THEORY FOR STUDENTS OF METALLURGY

(SECOND, REVISED EDITION)

By WILLIAM HUME-ROTHERY, O.B.E., F.R.S.

(*Institute of Metals Monograph and Report Series, No. 3*)

This well-known book was originally written for the Honours student in Metallurgy, though it has also been of great value to students in Chemistry and Physics. The original edition has, with only minor alterations, been reprinted twice, and more than 8000 copies have been sold. In the present revised Second Edition, most of the original text has been left unchanged, but substantial alterations and additions have been made to Part VI—Electrons, Atoms, Metals, and Alloys, and the chapter on Transition Elements has been rewritten; additions have also been made to the section dealing with Brillouin zones, some mistakes have been corrected, and throughout the book minor alterations have been made to bring the subject matter up to date. The length of the book has been increased by 36 pages.

Cloth, 331 pp. with 129 illustrations.

Price 21s. or $3.50, post free. (*Members, one copy at 10s. 6d. or $2.00.*)

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

THERMODYNAMICS OF ALLOYS

By JOHN LUMSDEN, B.Sc.

(*Research Department, Imperial Smelting Corporation, Ltd.*)

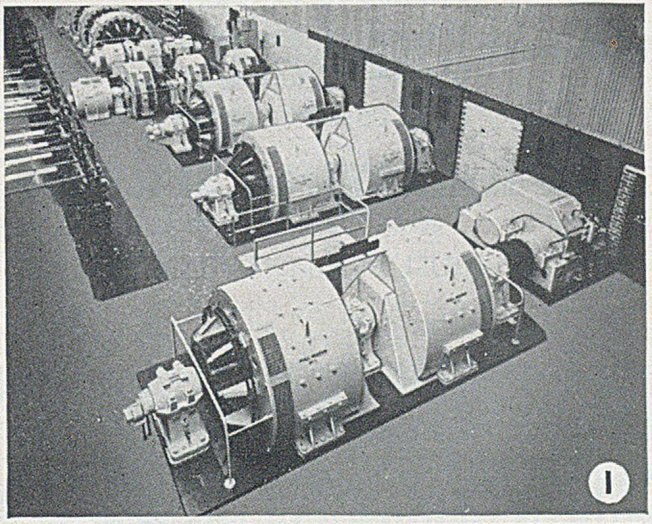
Institute of Metals Monograph and Report Series, No. 11 *Cloth,* 390 *pp. with* 110 *illustrations*

Price 35s. or $5.50 post free. (Members, one copy at 17s. 6d. or $3.00 post free.)

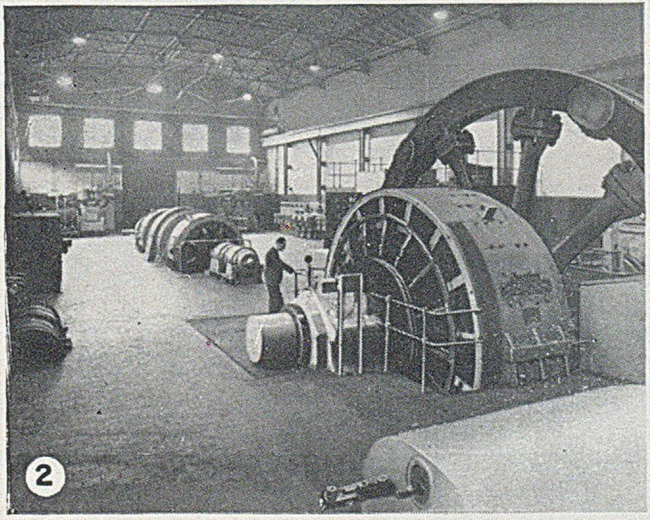
This monograph deals with the application of thermodynamics to the quantitative study of metals and alloys. The fundamental theory of the subject is presented first, and is followed by some illustrative calculations on pure metals. Then the theory of solutions is developed and exemplified by calculations on solid and liquid alloys. Emphasis is laid on the usefulness of thermodynamics for the accurate correlation of various equilibrium properties. Attention is particularly directed to the information that can be deduced from phase boundaries. Detailed calculations are presented on some typical systems for which measurements of vapour pressures, electrode potentials and heats of mixing are available; these results are combined with a study of liquidus, solidus and solid-solubility curves to derive the activities of both components in the liquid and solid phases. It is shown how a free-energy equation for a system, deduced from a suitable set of accurate experimental measurements, furnishes a reliable basis for calculating other equilibrium properties. To assist in numerical calculations, tables of functions that continually recur in thermodynamic problems are provided.

Elementary statistical mechanics is introduced for calculating the entropy of gases and for interpreting the form of free-energy curves in solid and liquid alloys. A critical study is made of the extent to which liquid alloys behave as " regular solutions."

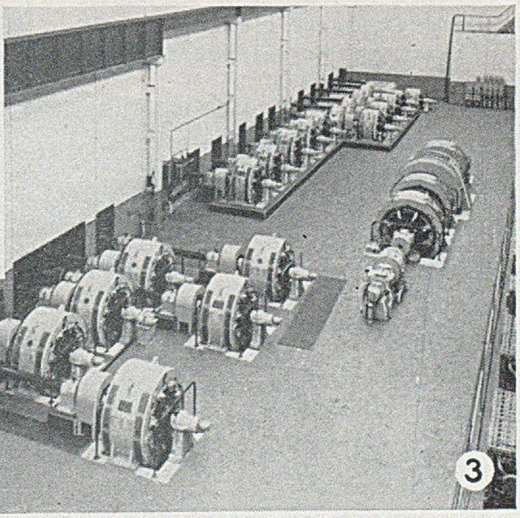
THE INSTITUTE OF METALS · 4 Grosvenor Gardens, London, S.W.1



①



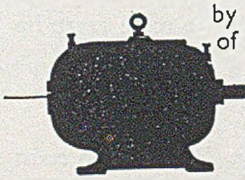
②

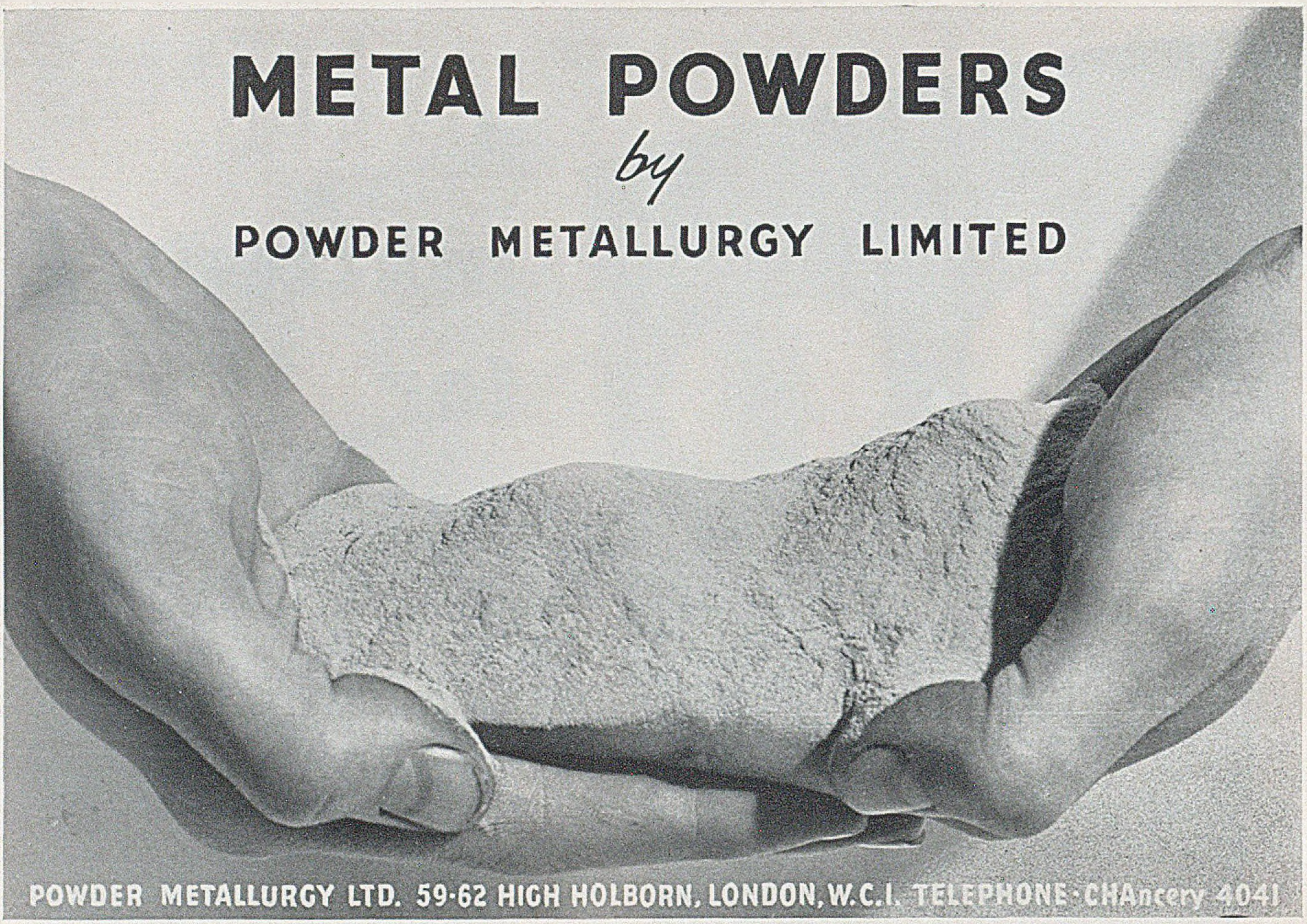


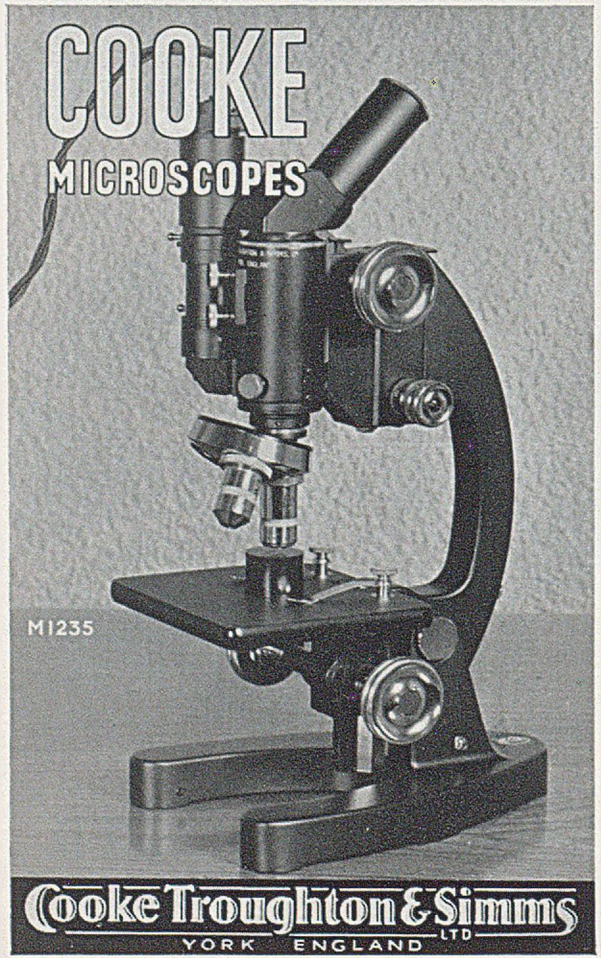
③

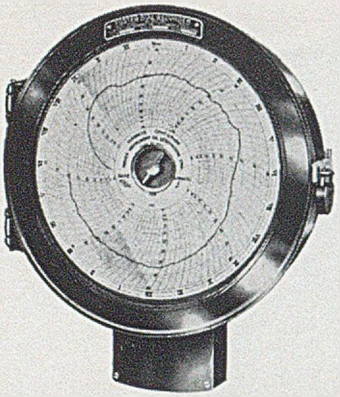

It pays to specify
McKECHNIE
'TANK' BRAND
non-ferrous
INGOTS
MKB
TANK
BRAND
VELVET OTTO
ANTIFRICTION METAL
No 1
BRAND
BIRMINGHAM
PHOS BRONZE
MW


THOMAS
BOLTON
& SONS LTD







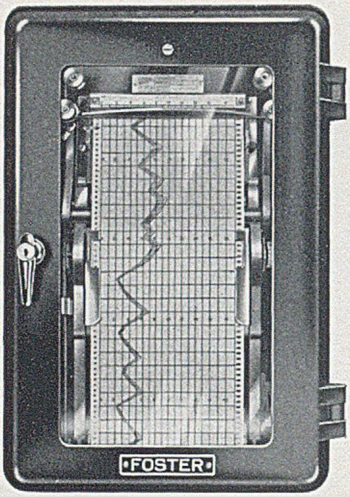


Dial Recorder

FOSTER

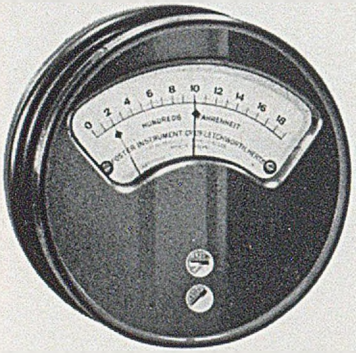
PYROMETERS

For over 40 years we have specialised in the design, production and application of pyrometers. This experience has enabled us to produce a range of instruments to satisfy the needs of practically every modern industrial process. Your problem in pyrometry is our problem and, having solved it, we ensure that the instruments supplied continue to give satisfactory and reliable service.



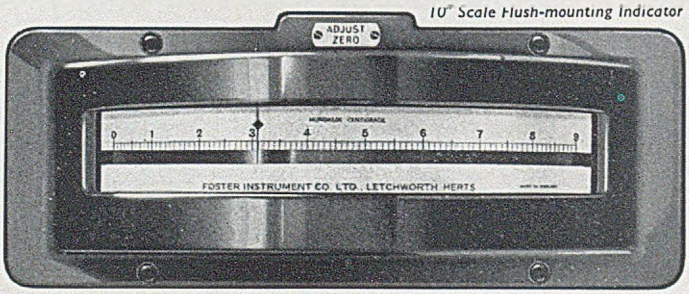
Multi-Point Strip Chart Recorder

5" Scale Circular Wall Type Indicator



★ We also make the FOSTER INTROSCOPE for internal inspection of inaccessible parts, including pipe welds, diesel jets, etc.

10" Scale Flush-mounting Indicator



FOSTER INSTRUMENT CO., LTD., LETCHWORTH, HERTS.

Telephone: LETCHWORTH 984 (3 lines) ENGLAND Telegrams: RESILIA, LETCHWORTH

FIRE!

MY LIFE LOST BUT FOR

NU-SWIFT!

"A match struck...an explosion... petrol-sprayed I became a living torch... but my 13-year old Nu-Swift Extinguisher saved me from agonizing death." Everyone has too many fire risks. What would you have done?

NU-SWIFT LTD. · ELLAND · YORKS

In Every Ship of the Royal Navy

★ WHO USES 'FULBOND'?

TRADE MARK

★Malleable ironfounders to avoid hot tears

For service & information write to:—

THE FULLERS' EARTH UNION LTD

Patteson Court, Redhill, Surrey

Telephone: REDHILL 3521

CMF.2

Electrify production—

use G.E.C.

ELECTRIC Process Heating

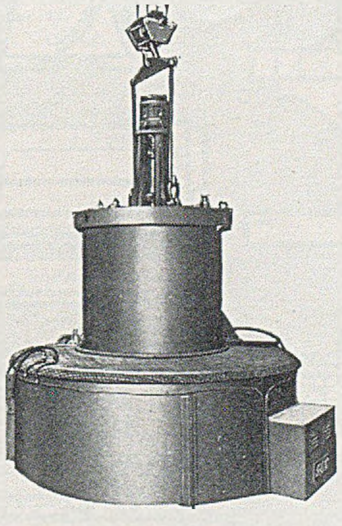
FURNACES · INFRA RED · HIGH FREQUENCY

● Bold thinking about bigger production demands new methods of heat treatment. G.E.C. *electric* process heating ensures maximum output and true economy. The many heating equipments in its range are clean, safe, simple to operate, and heat rapidly to constant correct and controlled temperature. Its versatility is being proved every day.

● Its applications are many : for heat treating steel, setting resin glues, drying paint and numerous other processes. For example, the vertical cylindrical furnace is widely used in heavy industry for the bright annealing of most metals.

● G.E.C. has had many years' experience in the design and construction of heating plant. The Company's specialists welcome opportunities to discuss potential uses with engineers and metallurgists.

Furnace Plant



G.E.C.
VERTICAL CYLINDRICAL FURNACE

The charge is carried in the heat-resisting steel container. Rapid and uniform heating is ensured by the forced convection set up inside the container by the high-speed fan combined with the correct distribution of heating elements on the furnace wall.

Vertical cylindrical furnaces provide a perfect bright anneal for most ferrous and non-ferrous metal and alloys in coils of wire or strip.

● ELECTRICITY ENSURES TRUE COAL ECONOMY

THE GENERAL ELECTRIC CO. LTD., MAGNET HOUSE, KINGSWAY, LONDON, W.C.2

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 5 PUBLISHED MONTHLY JANUARY

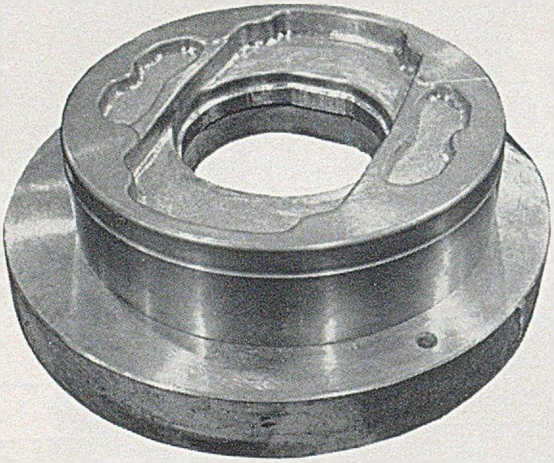
CONTENTS

BULLETIN

METALLURGICAL ABSTRACTS

JOURNAL

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



BULLETIN
OF THE INSTITUTE OF METALS

VOLUME 1 JANUARY 1953 PART 17

NEW YEAR MESSAGE FROM THE PRESIDENT

It gives me great pleasure to take this opportunity of wishing all members of the Institute a very Happy and Prosperous New Year. The Institute will continue to render you all the service it can, and in turn counts on your support.

C. J. Smithells

LONDON.
DECEMBER, 1952.

INSTITUTE NEWS

Spring Meeting, London, Monday–Thursday, 23–26 March 1953

The 1953 Annual General Meeting will be held in London from Monday to Thursday, 23–26 March, inclusive. All the scientific and technical sessions and a Dinner-Dance will be held at the Park Lane Hotel, Piccadilly, W.1.

The meeting will begin with the Annual May Lecture by Sir Christopher Hinton.

On the following day the Institute's Medals will be presented, and the new President will be inducted into the Chair and will deliver an Address. As a measure of economy, no Luncheon will be held in connection with this meeting, but there will be a Dinner-Dance, at which it is hoped that there will be a large attendance.

On the Wednesday there will be held the first of a series of three annual all-day Symposia on the control of quality in the production of wrought non-ferrous materials, and on the evening of that day a Conversazione and Exhibition will be held at the Institute's headquarters at 4 Grosvenor Gardens, S.W.1.

The meeting will end with a series of visits and, as an experiment, an all-day tour for ladies has been arranged, which it is hoped will be popular.

In view of their success at the 1952 Spring Meeting and at the Autumn Meeting at Oxford, simultaneous sessions on widely differing subjects will also be held.

The programme of the meeting is printed on pp. 134–136 of this issue. No separate programme is being printed in connection with this meeting, a **Meeting Reply Form for which is sent with this issue of the Journal.**

The Council hopes that there will be a large attendance at this meeting. Members are encouraged to bring as many friends as possible to the May Lecture, for which tickets of admission are not required.

It is particularly requested that those who intend to take part in the meeting will return the Reply Form at the earliest possible moment (and, if possible, not later than 18 March). The arrangements for the meeting are rendered difficult if members leave the return of their Reply Forms until the very last moment.

Members requiring hotel accommodation in London are advised to book their rooms early.

PROGRAMME

Monday, 23 March

Location : Royal Institution, Albemarle Street, W.1.

6.0 p.m.—Forty-Third Annual May Lecture, by Sir Christopher Hinton, M.A., M.I.C.E., M.I.Mech.E. (Deputy Controller of Atomic Energy (Production), Ministry of Supply) on " The Present and Future Metallurgical Requirements of the Chemical Engineer ".

Tuesday, 24 March

Location : Park Lane Hotel, Piccadilly, W.1 (Tudor Room).

10.30 a.m.—Annual General Meeting.

Report of Council.

Report of the Honorary Treasurer.

Presentation of:

- (*a*) Institute of Metals (Platinum) Medal.
- (*b*) W. H. A. Robertson Medal.
- (*c*) Rosenhain Medal.
- (*d*) Students' Essay Competition Prizes.

Induction of the new President, Professor F. C. Thompson, D.Met., M.Sc.

Votes of thanks to retiring officers.

Presidential Address, by Professor F. C. Thompson.

12.45 p.m.—Adjournment.

2.30 p.m.—Scientific session. The following papers (the references are to the *Journal*) will be discussed :

(1) *Jointly* : V. A. Phillips and A. Phillips : " The Effect of Certain Solute Elements on the Recrystallization of Copper " (No. 1431; Dec. 1952); and H. M. Miekk-oja : " Segregation of Iron and Phosphorus at the Grain Boundaries in 70 : 30 Brass During Grain Growth " (No. 1382; June 1952).

(2) *Theme : Young's Modulus of Alloys.*

Jointly : A. D. N. Smith : " A Study of Some Factors Influencing the Young's Modulus of Solid Solutions " (No. 1370; May 1952); and N. Dudzinski : " The Young's Modulus, Poisson's Ratio, and Rigidity Modulus of Some Aluminium Alloys " (No. 1414; Sept. 1952).

5.0 p.m.—Adjournment.

7.0 for 7.30 p.m.—Dinner-Dance, at the Park Lane Hotel, Piccadilly, W.1. *Dress :* Evening dress (tails or dinner jacket) or uniform with decorations. (Tickets price 32*s*. each, inclusive of coffee and gratuities, but exclusive of drinks.) Tables can be booked for parties of from 6 to 12 (incl.); others are requested to state a preference regarding seating.

Wednesday, 25 March

Two concurrent scientific and technical sessions, and an all-day programme for Ladies.

SESSION " A "

Location : Park Lane Hotel, Piccadilly, W.1. (Tudor Room).

9.30 a.m.—Symposium on " The Control of Quality in the Production of Wrought Non-Ferrous Metals and Alloys. Part I.—The Control of Quality in Melting and Casting ", arranged by the Metallurgical Engineering Committee. An all-day General Discussion, based on the following papers, which will be published in the March issue of the *Journal* (to be distributed to Members at the end of February) :

(1) " The Principles of Technical Control in Metallurgical Manufacture ", by A. R. E. Singer, Ph.D., B.Sc. (Senior Lecturer, Department of Industrial Metallurgy, University of Birmingham).

(2) " Control of Quality in Melting and Casting Aluminium Alloys for Working ", by R. T. Staples and H. J. Hurst (T.I. Aluminium, Ltd., Birmingham).

(3) " Control of Quality in the Melting and Casting of Copper and High-Conductivity Copper-Base Alloys ", by J. Sykes, F.I.M. (Enfield Copper Refining Co., Ltd., Brimsdown, Enfield).

(4) " Control of Quality in Melting and Casting Brass ", by Maurice Cook, D.Sc., Ph.D., F.I.M., and C. L. M. Cowley, B.Sc., A.I.M. (Imperial Chemical Industries, Ltd., Metals Division, Witton, Birmingham).

(5) " Control of Quality in the Melting and Casting of Magnesium and Its Alloys ", by R. G. Wilkinson, B.Sc. (Magnesium Elektron, Ltd., Clifton Junction).

(6) " The Control of Quality in the Casting of Zinc and Zinc Alloy Rolling Slabs and Extrusion Billets ", by C. W Roberts, B.Sc., A.I.M., and B. Walters, M.A. (Imperial Smelting Corporation, Ltd., Avonmouth).

12.45 p.m.—Adjournment.

2.30 p.m.—Symposium resumed.

5.0 p.m.—Conclusion of Symposium.

SESSION " B "

Location : Park Lane Hotel, Piccadilly, W.1 (Smoke Room).

9.30 a.m.—Scientific session. The following papers (references are to the *Journal*) will be discussed :

(1) *Theme : Corrosion of Alloys.*

Jointly : P. Brenner and G. J. Metcalfe : " The Effect of Cold Work on the Microstructure and Corrosion-Resistance of Aluminium–5% Magnesium Alloys Containing 0–1% Zinc " (No. 1441; Jan. 1953); G. J. Metcalfe : " Atmospheric Corrosion and Stress-Corrosion of Aluminium–Copper–Magnesium and Aluminium–Magnesium–Silicon Alloys " (Feb. 1953); and C. W. Roberts : " Intercrystalline Corrosion in Cast Zinc–Aluminium Alloys " (Feb. 1953).

(2) *Theme : High-Temperature Oxidation of Alloys.*

Jointly : A. Preece and G. Lucas : " The High-Temperature Oxidation of Some Cobalt-Base and Nickel-Base Alloys " (No. 1435; Dec. 1952); and J. P. Dennison and A. Preece : " High-Temperature Oxidation Characteristics of a Group of Oxidation-Resistant Copper-Base Alloys " (No. 1436; Jan. 1953).

12.45 p.m.—Adjournment.

2.30 p.m.—Scientific session. Discussion of the following papers :

Theme : Creep and Plastic Deformation.

A General Discussion based on the following papers :

W. A. Rachinger : " The Effect of Grain-Size on the Structural Changes Produced in Aluminium by

Slow Deformation" (No. 1363; April 1952); W. A. Wood and J. W. SUITER: "Stress-Recovery in Aluminium" (No. 1373; May 1952); D. MCLEAN: "Creep Processes in Coarse-Grained Aluminium" (No. 1374; May 1952); J. TROTTER: "Electron-Microscopic Studies of Slip in Aluminium during Creep" (No. 1375; May 1952); K. E. PUTTICK and R. KING: "Boundary Slip in Bicrystals of Tin" (No. 1378; June 1952); G. B. GREENOUGH, (Mrs.) C. M. BATEMAN, and (Mrs.) E. M. SMITH: "X-Ray Diffraction Studies in Relation to Creep" (No. 1379; June 1952); W. A. RACHINGER: "Relative Grain Translations in the Plastic Flow of Aluminium" (No. 1412; Sept. 1952); J. A. RAMSEY: "The Recovery of Polycrystalline Aluminium" (No. 1416; Oct. 1952); S. BHATTACHARYA, W. K. A. CONGREVE, and F. C. THOMPSON: "The Creep/Time Relationship under Constant Tensile Stress" (No. 1420; Oct. 1952); A. E. JOHNSON and N. E. FROST: "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" (No. 1421; Oct. 1952); D. MCLEAN: "Crystal Slip in Aluminium During Creep" (No. 1425; Nov. 1952); (Mrs.) V. M. URIE and H. L. WAIN: "Plastic Deformation of Coarse-Grained Aluminium" (No. 1427; Nov. 1952); J. W. SUITER and W. A. WOOD: "Deformation of Magnesium at Various Rates and Temperatures" (No. 1430; Dec. 1952); J. A. RAMSEY: "The Sub-Grain Structure in Aluminium Deformed at Elevated Temperatures" (No. 1434; Dec. 1952); D. MCLEAN: "Crystal Fragmentation in Aluminium During Creep" (Feb. 1953); D. MCLEAN: "Grain Boundary Slip During Creep of Aluminium" (Feb. 1953).

5.0 p.m.—Conclusion of session.

Location: 4 Grosvenor Gardens, S.W.1.

8.0 p.m.—CONVERSAZIONE AND EXHIBITION. Guests (including ladies) will be welcome. *Dress:* Lounge suit. (Tickets, price 4*s*., including light refreshments.)

Note: The Secretary will be pleased to receive offers of suitable exhibits.

LADIES' PROGRAMME

10.0 a.m.—ALL-DAY SIGHT-SEEING TOUR, with Guide, to WINDSOR, including the State Apartments and St. George's Chapel, and HAMPTON COURT. Lunch at Windsor; tea at Hampton Court. (Price of ticket, 25*s*., including lunch, tea, transport, guide, gratuities, and admission.)

Note: If the State Apartments are NOT open, the tour will NOT take place.

Thursday, 26 March

PROGRAMME "A"

Location: Park Lane Hotel, Piccadilly, W.1 (Tudor Room).

10.0 a.m.–12.30 p.m.—Informal Discussion on "Liquid Metals" arranged by the Metal Physics Committee.

PROGRAMME "B"

9.45 a.m.—Visits to the following works and laboratories, particulars of which are published below.

(1) BRITISH NON-FERROUS METALS RESEARCH ASSOCIATION, Euston Street, N.W.1.

(2) THE PYRENE CO., LTD., Brentford, Middlesex (morning); followed by

HOOVER, LTD., Perivale, Greenford, Middlesex (afternoon). (Lunch will be arranged.)

(3) VICKERS-ARMSTRONGS, LTD., Weybridge. (Lunch by invitation.)

Transport will start promptly from outside the Park Lane Hotel. The price of tickets for each visit is stated on the reply form.

BRIEF DESCRIPTIONS OF WORKS AND LABORATORIES TO BE VISITED

The British Non-Ferrous Metals Research Association

Euston Street, London, N.W.1

The Association started in 1921 in Birmingham; the first year's income was under £6000 and the membership 20. Its own laboratories were first established in 1930 on the present site, but have been greatly enlarged to serve an increasing membership, which now totals about 575 firms producing, manufacturing, or using non-ferrous metals. The annual income is over £100,000. The staff number 155, including about 50 graduates.

The laboratories include, in addition to equipment for general metallography and heat-treatment, apparatus for the determination of gases in metals and for the study of gas/metal reactions; a Physics Section, in which the techniques of X-ray crystallography, electron diffraction, and electron microscopy are applied to metallurgical research; a Mechanical Testing Section, equipped for tensile, fatigue, and creep testing of metals; an Experimental Foundry; an Experimental Metal-Working Shop, equipped with a 500-ton hydraulic press, 12-in. rolling mill, draw-bench, furnaces, and ancillary equipment; an Electrodeposition Laboratory; a Section devoted to the study of the corrosion of non-ferrous metals and to the development of corrosion-resistant alloys for special applications; laboratories for chemical and spectrographic analysis; and a fully-equipped workshop in which much of the Association's research equipment is made.

A comprehensive technical library and information service is maintained; loans to members exceed 9000 items a year. Through the Liaison and Technical Service Department, a free confidential technical enquiry service is maintained for members and assistance is given in the practical application of the results of the Association's researches.

Hoover, Ltd., Perivale

Hoover cleaners first came to Britain after World War I. At first all the cleaners came from the United States and later from Canada, but in 1931 it was decided to manufacture in Britain, and the factory at Perivale was begun. During World War II, 17 British Hoover factories produced electrical and other equipment for the war effort. Since the war there has been a great expansion in the production of domestic and light electrical products.

The Perivale factory remains the home of the famous electric cleaner, and some 2000 people work there. Among the operations that will be seen during the visit are: the pressure die-casting of aluminium alloys; the making and heat-treatment of the tools for die-casting and also for Bakelite moulding; pressworking of ferrous and non-ferrous materials;

production of laminations for electric motors; metal-finishing processes, including plating and enamelling; and the assembly of motors and vacuum cleaners.

The Pyrene Co., Ltd., Brentford

Although this firm is perhaps most widely known for its fire extinguishers, its factory on the Great West Road also has a large metal-finishing department where the "Bonderizing", "Parkerizing", and "Pylumin" processes are carried out.

Vickers-Armstrongs, Ltd., Weybridge

At these works civil and military aircraft and guided weapons are designed and manufactured.

In addition to the current production of such aeroplanes as the *Viscount* and the *Valiant*, considerable research work is carried out. Some of a fundamental nature is concerned with aerodynamics, structures, and properties of materials under varying conditions of temperature, pressure, humidity, &c.

The research equipment additional to a Materials Laboratory includes a supersonic duct, a near-sonic-speed wind-tunnel, and a very large chamber (25 by 50 ft) in which it is possible to produce atmospheres varying from one equivalent to ground level at the equator to one 100,000 ft. above the earth.

Institute of Metals (Platinum) Medal

Mr. W. S. Robinson, to whom the Institute of Metals (Platinum) Medal for 1952 was awarded last March, recently received the Medal from the hands of the Governor-General of Australia at a private meeting.

Election of Members

The following 11 Ordinary Members, 4 Junior Members, and 15 Student Members were elected on 19 November 1952:

As Ordinary Members

Bishop, Edward, B.Sc., Technical Assistant to Metals Economy Adviser, Ministry of Supply, Shell-Mex House, London, W.C.2.

Buzzard, Robert W., B.S., Research Metallurgist, National Bureau of Standards, U.S. Department of Commerce, Washington 25, D.C., U.S.A.

Chisholm, Donald William, Standards Engineer, The English Electric Co., Ltd., Rugby.

Conard, George P., Sc.D., M.S., Assistant Professor of Metallurgy, Lehigh University, Bethlehem, Pa., U.S.A.

Crussard, Charles, Head, Metal Physics Department, L'Institut de Recherches de la Sidérurgie, St. Germain-en-Laye (S. et O.), France.

Houdremont, *Professor* Eduard, Dr.Ing., Professor of Metallurgy, Technische Hochschule, Aachen, Germany.

Köhn, Wilhelm, Technical Director, Hindrichs-Auffermann A.G., Wuppertal-Oberbarmen, Germany.

Lachenaud, René, Chef du Laboratoire Central, Société Nationale de Constructions Aéronautiques du Sud-Ouest, 2 rue de Larnac, Courbevoie (Seine), France.

Levy, Frank M., B.Chem., Technical Superintendent, Mueller Brass Company, Port Huron, Mich., U.S.A.

Waddell, Thomas, Metallurgist, Royal Ordnance Factory, Patricroft, Manchester.

Winsor, Reginald, Commissioner for Railways, Department of Railways, 19 York Street, Sydney, N.S.W., Australia.

As Junior Members

Brown, Gordon T., B.Sc., Junior Metallurgist, Quasi-Arc Co., Ltd., Bilston, Staffs.

Klein, Wesley Graham, Assistant Chief Engineer, P.O. Box 127, Cowra 3W, N.S.W., Australia.

Leach, Anthony, B.Sc., A.R.I.C., Research Laboratories of The General Electric Co., Ltd., North Wembley, Middlesex.

Martin, Edmund John, Glacier Metal Co., Ltd., Alperton, Wembley, Middlesex.

As Student Members

Berkeley, Kenneth Gordon Charles, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Duff, Thomas, Student of Metallurgy, Royal Technical College, Glasgow.

Elliott, John Edward, Metallurgical Analyst, Electro-Refined Irons, Ltd., Taylor Street, Darwen, Lancs.

Gardiner, Keith M., Student of Metallurgy, University of Manchester.

Green, Frank Alan, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Heppleston, Colin, Student of Metallurgy, University of Sheffield.

Knaggs, Kenneth, Student of Metallurgy, University of Sheffield.

Locke, David Harold, B.Sc., Research Student, University of Birmingham.

Partridge, Peter G., Student of Metallurgy, Bristol Aeroplane Co., Ltd., Filton, Bristol.

Roberts, Peter Rowe, Student of Metallurgy, University of Manchester.

Sillitoe, John Frederick, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Tiplady, Geoffrey Rowland, Student of Metallurgy, University of Sheffield.

Vickers, Walter, Student of Metallurgy, University of Sheffield.

Watkinson, F., Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

Wright, Mark, Student of Metallurgy, University of Liverpool.

The following 7 Ordinary Members, 3 Junior Members, and 18 Student Members were elected on 12 December 1952:

As Ordinary Members

Deisinger, Walter, Dr. Ing., Director, Vacuumschmelze A.G., Hanau, Western Germany.

Dewsnap, Noel F., B.Met.E., Instructor, School of Mines, Ballarat, Vic., Australia.

Inshaw, Howard Austen, Works Manager, Maudslay Motor Co., Ltd., Castle Maudslay, Alcester, Warwickshire.

Metzger, Guinn E., Met.Eng., Royal School of Mines, Prince Consort Road, London, S.W.7.

Rogers, B. A., B.S., M.S., Ph.D., Senior Metallurgist, Institute for Atomic Research, Iowa State College, Ames, Iowa, U.S.A.

Steven, Gary, B.S., M.S., Research Metallurgist, Armour Research Foundation, 35 W. 33rd Street, Chicago 16, Ill., U.S.A.

Terry, William Joseph, Chairman and Managing Director, London Electric Wire Company and Smiths, Ltd., 24 Queen Anne's Gate, London, S.W.1.

As Junior Members

Craik, Robert Laidlaw, B.Sc., A.R.T.C., Postgraduate Student, Metallurgy Department, Royal Technical College, Glasgow, C.1.

Marton, Stephen, B.Met., Research Metallurgist, Sheepbridge Engineering, Ltd., Chesterfield, Derbyshire.

Wragg, Vernon, B.Met., Assistant Lecturer in Metallurgy, County Technical College, Wednesbury, Staffs.

As Student Members

Appleton, Arthur Stanley, Student of Metallurgy, University of Liverpool.

Baird, J. Sheila M., Assistant Metallurgist, Bull's Metal and Marine, Ltd., Bulldale Street, Glasgow, W.4.

Baker, Dennis Walter Clifford, B.Sc., Research Metallurgist, Development and Research Department, The Mond Nickel Co., Ltd., Bashley Road, London, N.W.10.

Bean, John Kenneth, Student of Metallurgy, University of Leeds.

Bonham, Ian C., Student of Metallurgy, University of Leeds.

Clare, John William Henry, Student of Metallurgy, University of Birmingham.

Gough, John Richard Cox, B.Sc., Scientific Officer, Ministry of Supply Establishment, Aldermaston, Berks.

Jopling, John Denis, Student of Metallurgy, University of Leeds.

King, Robert James, Student of Metallurgy, Constantine Technical College, Middlesbrough.

Livey, D. T., B.Sc., A.R.T.C., Research Student, Metallurgy Department, Royal Technical College, Glasgow, C.1.

Mehd, Dinesh P., Student of Metallurgy, University of Leeds.

Pearce, Jai, Student of Metallurgy, College of Mining and Metallurgy, Benares Hindu University, Benares, U.P., India.

Pucknell, David John, L.I.M., Apprentice, Royal Ordnance Factory, Nottingham.

Sharp, John David, Student of Metallurgy, University of Cambridge.

Tate, Reginald, B.Met., Student of Metallurgy, National Testing Laboratory, Winnipeg, Manitoba, Canada.

Taylor, J. L., B.Sc., Student of Metallurgy, University College, Cardiff.

West, John Michael, Student of Metallurgy, University of Cambridge.

Wright, Gordon Cunliffe, Student of Metallurgy, University of Leeds.

Representation on Other Bodies

Dr. C. J. Smithells has been appointed to be the Institute's representative on the Parliamentary and Scientific Committee, in succession to the late Sir William Griffiths.

Colonel Sir Paul Gueterbock and Mr. H. W. L. Phillips have been reappointed as the Institute's representatives on the Empire Council of Mining and Metallurgical Institutions.

Mr. W. A. Baker has been appointed as representative of the Institute in connection with the publication of "Acta Metallurgica", in succession to the late Sir William Griffiths.

"Metallurgical Abstracts" Volumes Wanted

The Institute does not now require any more copies of the monthly parts of the *Journal and Metallurgical Abstracts* for 1940. The following monthly parts are, however, still required, and the Institute will pay 7*s.* 6*d.* for each copy in good condition.

1941: July, Aug.
1942: July, Aug., Oct.
1943: May, July, Aug., Sept., Oct.

The Institute will also pay £1 each for bound volumes, in good condition, of *Metallurgical Abstracts* 1940 (Vol. 7), 1941 (Vol. 8), 1942 (Vol. 9), 1943 (Vol. 10).

Copies offered for sale should be addressed to the Secretary.

Members' Addresses Missing

The Secretary has lost touch with the following members, and would be glad to receive information regarding their present whereabouts and addresses:

Name	*Last known address*
Atkins, L. J.	c/o W. B. Stagg Macey, Box 55, Ndola, Northern Rhodesia.
Britton, J.	14 St. Albans Road, Swansea. (*Student Member.*)
Clark, R.	15 Lindsey Avenue, Ontario, Canada. (*Junior Member.*)
Dodridge, N. R.	30 Joydon Drive, Chadwell Heath, Essex.
Ellıethy, N. F.	Zaid St. 12, Prince Mohamed Ali Street, Cairo, Egypt.
Halliwell, V. T. H.	66, Bromwich Street, Haulgh, Bolton, Lancs. (Now believed to be in Australia.)
Kostroun, Ing. V.	c/o Sandrik, Mikulasovice, Czechoslovakia.
McKnight, I. B.	43 Halesworth Road, Lewisham, London, S.E.13.
Moorby, F. H.	P.O. Box 9587, Johannesburg, South Africa.
Naor, Ing. P.	53 Lightwoods Hill, Birmingham, 41. (*Student Member.*)
Prytherch, W. E.	"Little Hockeridge", Ashley Green, Chesham, Bucks.

PERSONAL NOTES

Mr. R. W. Adam has been awarded the B.A. degree of Cambridge University and has taken up a post in the Research Laboratories of Imperial Chemical Industries, Ltd., Metals Division, Birmingham.

Professor Dr. P. Brenner has been elected President of the Deutsche Gesellschaft für Metallkunde in succession to Professor Dr. G. Masing.

Mr. D. J. Bullivant has been appointed a Director of Shardal Castings, Ltd., Inge Street, Birmingham 5.

Mr. A. C. J. Burningham has resigned his Directorship of Durham Chemicals, Ltd., and Associated Companies, and has joined the Board of The Egglescliffe Chemical Company, Egglescliffe, Co. Durham.

Mr. John Cartland resigned his directorships of Fry's Metal Foundries, Ltd., and The Eyre Smelting Co., Ltd., on 31 December. He will continue to act as a Consultant to the Fry Organization, with which he has been connected for 33 years. Mr. Cartland, who is a Past-Chairman of the London Local Section, served as an Ordinary Member of Council of the Institute from 1940 to 1947 and as a Vice-President from 1947 to 1950.

Dr. F. A. Fox has joined the Australian Government Department of Supply as Superintending Scientist in charge of the Department's Chemical and Physical Research Laboratories, Maribyrnong, Victoria. His address is: c/o Department of Supply, 339 Swanston Street, Melbourne, Australia.

Dr. G. E. Gardam is the first recipient of the Hothersall Medal of the Institute of Metal Finishing.

Dr. P. Greenfield is now a Research Associate in the Department of Metallurgy at the University of Illinois, Urbana, Ill.

Dr. J. L. Haughton acted as Secretary to a group of metallurgists from various European countries that recently visited the United States under the auspices of the Organization for European Economic Co-operation to study productivity in the American magnesium industry.

Mr. S. T. M. Johnstone has returned to the Aeronautical Research Laboratories, Melbourne, after working for a time at the National Physical Laboratory, Teddington.

Mr. E. F. Maillard was a member of a team that recently visited the United States to study productivity in the magnesium industry.

Mr. R. J. Maitland is now Lecturer in Metallurgy at the Borough Polytechnic, London, S.E.1.

Professor R. F. Mehl (Corresponding Member to the Council for the United States) has been awarded the Gold Medal of the American Society for Metals.

Mr. N. M. Mitchell has left Farnborough and returned to the Defence Research Laboratory, Maribyrnong, Victoria, Australia.

Dr. S. Mocarski has left England and taken a post with the General Electric Company, Toronto, Canada.

Dr. J. N. Pratt has resigned his post as Research Fellow in the Department of Metallurgical Engineering, University of Toronto, to become I.C.I. Research Fellow in the Department of Metallurgy, Manchester University.

Mr. A. Prince has been appointed Lecturer in Metallurgy at University College, Southampton.

Mr. P. Rowley has been appointed Metallurgist to The Massey-Harris Co., Ltd., Kilmarnock.

Mr. M. B. Sanders has been awarded the B.Sc. degree of the University of Wales and is now in the employ of the Bristol Aeroplane Co., Ltd.

Dr. Amos J. Shaler has been appointed Professor of Metallurgy and Chief of the Division of Metallurgy at the Pennsylvania State College. During 1951 Dr. Shaler served as Scientific Liaison Officer with the U.S. Office of Naval Research in London, and since then he has been Associate Professor of Metallurgy at the Massachusetts Institute of Technology.

Mr. H. J. Sharp, Chief Chemist and Metallurgist to Hoover, Ltd., Perivale, has been awarded the degree of M.Sc. by London University for a thesis on die-casting.

Sir Arthur Smout (Past-President) has recently received the City and Guilds of London Institute Insignia Award in Technology from the President of the Institute, The Duke of Edinburgh. The Award, which has just been established, is intended to be a mark of distinction for those who have combined with a sound practical training an adequate knowledge of the fundamental scientific principles of their industry, and who possess a capacity for leadership and administration.

Professor Cyril Stanley Smith (Honorary Corresponding Member to the Council for the United States) has been awarded the Francis J. Clamer Medal by the Franklin Institute for his metallurgical contribution to the development of atomic energy during and since World War II.

Mr. B. N. H. Thornely has returned to England from Canada.

Mr. J. F. Whitfield has relinquished his post as Chief Engineer of Birmetals, Ltd., to take up the appointment of General Manager of B. Thornton, Ltd., Huddersfield.

Mr. M. E. Williams has resigned his post with McKechnie Bros., Ltd., and taken up an appointment at the Widnes works of Thomas Bolton and Sons, Ltd., as Superintendent of the Smelting and Refining Department.

Death

The Editor regrets to announce the death of:

Dr. Gilbert Henry Gulliver on 24 October. He was an Original Member of the Institute and until his retirement he was associated with David Kirkaldy and Son, London.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

Birmingham Local Section

The Birmingham Local Section is arranging to hold, on Friday, 27 February 1953, an all-day meeting at The College of Technology, Suffolk St., Birmingham, when papers concerned with "Making the Best of Metals (Resources, Recovery, Refining, Uses)" will be presented.

An introductory address will be given by Mr. R. Lewis Stubbs, O.B.E., on "The Resources of Non-Ferrous Metals" and other papers have been promised by Mr. T. Burchell, Mr. S. Hands, Dr. V. Kondic, Mr. H. J. Miller, and Dr. E. Scheuer. There will be an open discussion during the afternoon session, when short supplementary contributions will be welcomed.

A synopsis of the papers will be sent to all who register for the meeting (registration fee 7*s.* 6*d.*, lunch 8*s.* 6*d.*). Communications and enquiries should be addressed to the Hon. Secretary, Mr. A. W. Matthews, 124 Hay Green Lane, Bournville, Birmingham 30.

London Local Section

At a meeting of the Section held at 4 Grosvenor Gardens, London, S.W.1, on 2 October, Dr. C. E. **Ransley** of The British Aluminium Co., Ltd., delivered his Chairman's Address entitled:

Some Useful Techniques in Metallurgical Research

The talk was mainly concerned with experimental methods for the determination of the equilibrium constitution of alloys, and some of the work described has not been published.

The lecturer said that for the investigation of phase boundaries in certain solid alloy systems, a new "sandwich" technique had proved very useful. In this, alloys were prepared with particular phases segregated into layers by hot rolling together plates of appropriate compositions; the composite material was rolled down to thin sheet, annealed until equilibrium was established by diffusion, and the final distribution of elements then determined by analysis.

Reference was made to the thermal analysis of alloys, and attention drawn to the advantages of C. S. Smith's constant-heat-flow method. A form of apparatus, which had been constructed for cooling-curve determinations on aluminium alloys, was described, and some features of the inverse-rate curves obtained with it were discussed.

The speaker said that constitutional problems could frequently be attacked by indirect methods; the absorption of hydrogen by solid alloys, for example, was sometimes instructive, and he described a high-vacuum analysis system capable of dealing with very small volumes of gas, which was very suitable for such measurements. The application of this technique was illustrated by reference to some work on the aluminium–sodium and aluminium–silicon–sodium alloys.

This type of vacuum system was also the basis of a convenient apparatus for measuring the gas content of metals, and a brief description was given of the procedures for determining the hydrogen present in aluminium and aluminium alloys, and mechanically extracting gas from blisters in wrought material.

JOINT ACTIVITIES

Joint Committee on Metallurgical Education

The Joint Committee on Metallurgical Education of the Iron and Steel Institute, the Institution of Mining and Metallurgy, the Institute of British Foundrymen, the Institute of Metals, and the Institution of Metallurgists has recently published a 56-page report on "The Education and Training of Metallurgists". Copies are available from the Secretary of the Committee, 4 Grosvenor Gardens, London, S.W.1.

Mond Nickel Fellowships Awards

The Mond Nickel Fellowships Committee has announced the following awards 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.

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.

DIARY

The Institute

8 January. Informal Discussion on "Rolls and Their Maintenance in the Non-Ferrous Metals Industry". (The University, Birmingham 15, at 11.30 a.m.). Before the meeting there will be a visit to the Aitchison Metallurgical Laboratories, The University, Birmingham 15, at 10.30 a.m.

Local Sections and Associated Societies

13 January. South Wales Local Section. "The Metallurgical Problems Arising from Stratospheric Flight", by Major P. L. Teed. Official Visit of the President and Secretary to the Local Section. (Metallurgy Department, University College, Singleton Park, Swansea, at 6.30 p.m.)

19 January. Scottish Local Section. Visit to Remington Rand, Ltd., Hillington, Glasgow.

21 January. Liverpool Metallurgical Society. "Alloys Resistant to Oxidation and Creep at High Temperature", by Dr. S. J. Kennett. Joint Meeting with the North Wales Metallurgical Society. (Liverpool Engineering Society, 9 The Temple, Dale Street, Liverpool, at 6.30 p.m.)

21 January. London Local Section. "Analysis of Metals by Spectroscopy", by Dr. A. C. Menzies. Joint Meeting with the London Section of the Society of Chemical Industry and the London and South-Eastern Counties Section of the Royal Institute of Chemistry. (Large Chemistry Theatre, University College, Gower Street, London, W.C.1, at 7.0 p.m.)

21 January. Manchester Metallurgical Society. "Indentation Hardness", by Dr. D. Tabor. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

23 January. Sheffield Local Section. "Observations on Electroplating Research and on Flatware Manufacture in the U.S.A.", by Dr. G. E. Gardam. Joint meeting with the Institute of Metal Finishing, Sheffield and North East Branch. (Grand Hotel, Sheffield, at 6.30 p.m.)

26 January. North East Metallurgical Society. Discussion on "The Training of a Metallurgist". Opening Speaker: Dr. A. D. Merriman. (Cleveland Scientific and Technical Institution, Middlesbrough, at 7.15 p.m.)

29 January. Birmingham Local Section. Discussion on "Modern Technique in Spectrographic Analysis". (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

3 February. Oxford Local Section. Symposium on Metal Pressing: (1) Film supplied by the Aluminium Development Association. (2) A speaker from Aluminium Laboratories, Limited. (3) A speaker from Pressed Steel Co., Ltd. (Black Hall, St. Giles, Oxford, at 7.0 p.m.)

4 February. Manchester Metallurgical Society. "Some New Laboratory Techniques", by Dr. F. Ashworth, F. G. Haynes, and J. Johnson. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

5 February. Birmingham Local Section. "Substitution", by Professor A. J. Murphy. (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

5 February. Leeds Metallurgical Society. "The Continuous-Casting Processes", by Dr. E. Scheuer. (Chemistry Department, The University, Leeds 2, at 7.15 p.m.)

5 February. London Local Section. "Titanium", by Dr. N. P. Allen. (4 Grosvenor Gardens, London, S.W.1, at 7.0 p.m.)

9 February. Scottish Local Section. "The Production of Non-Ferrous Castings", by R. F. Hudson. (Institution of Engineers and Shipbuilders in Scotland, 39 Elmbank Crescent, Glasgow, C.2, at 6.30 p.m.)

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 6.30 p.m.)

Other Societies

14 January. Institute of British Foundrymen, Southampton Section. "Synthetic Resins in the Foundry", by P. G. Pentz. (Southampton Technical College, St. Mary Street, Southampton, at 7.0 p.m.)

15 January. Institution of Mining and Metallurgy. General Meeting. (Rooms of the Geological Society, Burlington House, Piccadilly, London, W.1, at 5.0 p.m.)

16 January. Institution of Mechanical Engineers. "Welding in Marine Engineering", by H. N. Pemberton. (Institution of Mechanical Engineers, Storey's Gate, St. James's Park, London, S.W.1, at 5.30 p.m.)

19 January. Institute of Metal Finishing, London Branch. "The Economics of Industrial Painting", by J. N. T. Adcock. (Northampton Polytechnic, St. John Street, London, E.C.1, at 6 p.m.)

19 January. Institution of Production Engineers, North-Eastern Section. "Drop Stamping", by A. Chilton. (Neville Hall, Westgate Road, Newcastle-upon-Tyne, at 7.0 p.m.)

20 January. Society of Chemical Industry, Corrosion Group. "The Electrochemical Behaviour of Metal and Corrosion", by Dr. M. Pourbaix. (The Chemical Society, Burlington House, London, W.1, at 6.30 p.m.)

21 January. Institute of British Foundrymen, East Midlands Branch. "System of Studying Casting Defects", by G. W. Nicholls and D. T. Kershaw. (College of Arts and Crafts, Derby, at 6.0 p.m.)

21 January. Institution of Production Engineers, South Essex Sub-Section. "Modern Electroplating and Metal-Finishing Processes", by H. Cann. (Ilford Bowling Club, nr. Ilford Station, Essex, at 7.30 p.m.)

22 January. Institute of Welding, North London Branch. "The Joining of Aluminium and Its Alloys", by W. V. Binstead. (The Polytechnic, Regent Street, London, W.1, at 7.30 p.m.)

22 January. Institution of Production Engineers, London Section. "Recent Developments in Metal Machining", by K. J. B. Wolfe. (Royal Empire Society, Northumberland Avenue, London, W.C.2 at 7.0 p.m.)

24 January. Institute of British Foundrymen, Bristol and West of England Branch. "Mechanical Aids in the Foundry", by J. Blakiston. (Grand Hotel, Bristol, at 3.0 p.m.)

28 January. Institute of British Foundrymen, London Branch. "Runners and Risers". Three papers by E. D. Daybell, P. A. Russell, and R. W. Ruddle. (Waldorf Hotel, London, W.C.2, at 7.30 p.m.)

29 January. Royal Aeronautical Society. "Titanium", by Major P. Litherland Teed. (Institution of Mechanical Engineers, Storey's Gate, London, S.W.1, at 6.0 p.m.)

30 January. Institute of British Foundrymen, Falkirk Section. "Synthetic Resins", by R. Carswell. (Temperance Café, Lint Riggs, Falkirk, at 7.30 p.m.)

3 February. Institute of Metal Finishing, Midland Branch. "Properties of Electrodeposits in Relation to Engineering Requirements", by R. A. F. Hammond. (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

5 February. Society of Chemical Industry, Corrosion Group. "Atmospheric Corrosion and Atmospheric Pollution", by Dr. J. C. Hudson and J. F. Stanners. Joint meeting with the Society of Chemical Industry, Nottingham Section. (Gas Showrooms, Parliament Street, Nottingham, at 7.15 p.m.)

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.)

APPOINTMENTS VACANT

A VACANCY exists in a large engineering company in the Midlands for a tropicalization expert. The successful applicant must be an engineer with at least seven years' industrial experience, and he should also possess a considerable knowledge of chemistry and metallurgy. He will be required to investigate and initiate processes to prevent corrosion of manufactured articles being shipped abroad. Please reply, quoting reference AJBH, to Box 344, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

COMPETENT ANALYST required with several years' experience of Metallurgical Analysis, for Aluminium Alloy Manufacturers in North London area. Age preferably 20–30 years. Write giving full particulars, Box 346, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

EXPERIENCED CHIEF CHEMIST required for mine assay office in Cyprus. Salary not less than £100 per month, but according to qualifications and experience. Free quarters, family passage paid. Three years' initial contract renewable for suitable man. Three months' paid home leave after three years. Write Box No. 345, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGICAL CHEMIST required for East London factory. An interesting and progressive position is offered to the right applicant and entails process control investigational work, using chemical metallurgical analytical methods. B.Sc. preferred, but not essential. Age 25–35. Commencing salary in the range £600–£700 p.a. Write Box Z.X.511, Deacon's Advertising, 36 Leadenhall Street, E.C.3.

METALLURGIST, with University qualifications, required for development work on light alloy welding in metallurgical research department. Preference given to applicant with some industrial experience in light alloy welding or founding. Excellent salary for suitable applicant. Apply in writing to the Personnel Manager, Murex Welding Processes, Ltd., Waltham Cross, Herts.

METALLURGIST, with University qualifications, required for investigational work on problems associated with the welding of steels. Some experience in research or industry essential and a knowledge of arc welding an advantage. Excellent salary and prospects for suitable applicant. Apply in writing to the Personnel Manager, Murex Welding Processes, Ltd., Waltham Cross, Herts.

MULLARD BLACKBURN WORKS, LTD., require a metallurgist for laboratory investigations into the properties of, and the manufacturing processes for, tungsten and molybdenum wires. Applicants should possess a degree in metallurgy or the Associateship of the Institution of Metallurgists; experience in powder metallurgy or previous work on the recrystallization of metals would be an advantage, but a new graduate would be considered for the post. Salary in accordance with age and qualifications. Apply in writing, giving full details, to the Works Personnel Officer, Mullard Blackburn Works, Ltd., Philips Road, Blackburn, Lancs.

HIGH-TEMPERATURE OXIDATION CHARACTERISTICS OF A GROUP OF OXIDATION-RESISTANT COPPER-BASE ALLOYS *

1436

By J. P. DENNISON,† Ph.D., B.Sc., JUNIOR MEMBER, and
PROFESSOR A. PREECE,‡ M.Sc., F.I.M., MEMBER

SYNOPSIS

The influence of small separate additions of aluminium, beryllium, chromium, magnesium, and silicon on the high-temperature oxidation of copper has been examined, these alloying elements having been chosen because their oxides have a high electrical resistivity. The composition and microstructure of the scales formed were studied and correlated with the rates of oxidation.

The effectiveness of the additions in conferring oxidation-resistance was in the decreasing order: beryllium, aluminium, magnesium, silicon, chromium.

Alloys having rates of oxidation less than that of copper 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 copper and of the alloying element through the oxide layers.

I.—INTRODUCTION

THE results described in the present paper were obtained in a general investigation of the high-temperature oxidation characteristics of a selection of binary copper-base alloys. The alloying elements chosen were aluminium, beryllium, chromium, magnesium, and silicon.

Earlier investigations by Wagner [1] and by Price and Thomas [2] have demonstrated the importance, among other things, of a high electrical resistivity in the oxide film if protection against continued oxidation is desired. The elements chosen have a much higher affinity for oxygen than that of copper; their oxides are refractory, and with the exception of that of chromium, possess high electrical resistivity.

Relatively little detailed information is available concerning the exact nature of the oxide/metal interface or the constitution of the oxides formed on these alloys, especially in the temperature range 400°–700° C. It was decided therefore to begin with the simple binary alloys before proceeding to the more complex compositions.

In view of the recent publication of a comprehensive review of the literature on the oxidation of copper by Tylecote,[3] only references directly connected with the present work are included.

II.—EXPERIMENTAL WORK

1. MATERIAL AND PROCEDURE

The alloys listed in Table I were prepared by melting electrolytic copper under charcoal and adding the alloying element in the form of a master alloy. Bars of 1-in. dia. were cast by the Durville process, hot rolled to ½-in. dia., and machined to a smooth finish.

TABLE I.—*Compositions of Alloys Used.*

Element	Alloy Content		Copper, %	Remainder, by Difference, %
	Nominal, %	Actual, %		
Aluminium	2 4 6 8 10	2·05 4·03 5·96 8·08 9·90	97·90 95·91 94·00 91·87 90·04	0·05 0·06 0·04 0·05 0·06
Beryllium	1 2	1·05 2·00	98·92 97·96	0·03 0·04
Chromium	0·5 1·5	0·50 1·20	99·46 98·77	0·04 0·03
Magnesium	1	0·90	99·07	0·03
Silicon	2 3·5	2·03 3·55	97·93 96·40	0·04 0·05

Specimens cut from these bars were exposed at the required temperatures to ordinary air and, in the preliminary part of the investigation, to the products of combustion of paraffin containing 2% sulphur burnt with a 60 : 1 air-to-fuel ratio to give an atmosphere of the following composition: N_2 78, O_2 17, H_2O 3, CO_2 2, and SO_2 0·11%.

It was found that only when the fuel contained sulphur was there any difference in the rate of oxidation. This difference, which was slight, was directly proportional to the sulphur content of the combustion atmosphere, and was apparent only at temperatures

---

* Manuscript received 24 June 1952.

† Lecturer in Metallurgy, University College, Swansea.

‡ Metallurgy Department, King's College, Newcastle-on-Tyne.

below 700° C. All the results reported below relate to exposure in air.

After exposure the specimens were cooled in a desiccator. The rate of oxidation was expressed as weight increase/unit of original surface area.

In most cases the scale produced was non-adherent; small particles of any scale adhering were removed by compressing the specimens in a confined space. The latter method was also adopted for removing adherent scales.

Before being sampled for analysis, the scales were powdered and thoroughly mixed. Analysis was carried out on approximately 0·5-g. samples. Copper was estimated electrolytically and the remaining solution analysed for the alloying element by standard gravimetric methods. Assuming the formula for the oxide of the alloying element, the percentage of copper in the copper oxide was used to calculate the amount of cupric oxide formed.

In microscopic examination polarized light was extremely useful in differentiating between cuprous and cupric oxides, the former appearing ruby red and the latter jet black. The crystalline appearance of cupric oxide was accentuated by partial rotation of the polarizer.

2. Results

(a) *Copper–Aluminium Alloys*

Oxidation rates for the copper–aluminium alloys at temperatures between 400° and 1000° C. are shown in Fig. 1. The double inflection in the curves for the 2 and 4% aluminium alloys in the range 700°–800° C. will be noted. The rate of oxidation does not follow the parabolic law, as shown by Figs. 2 and 3.

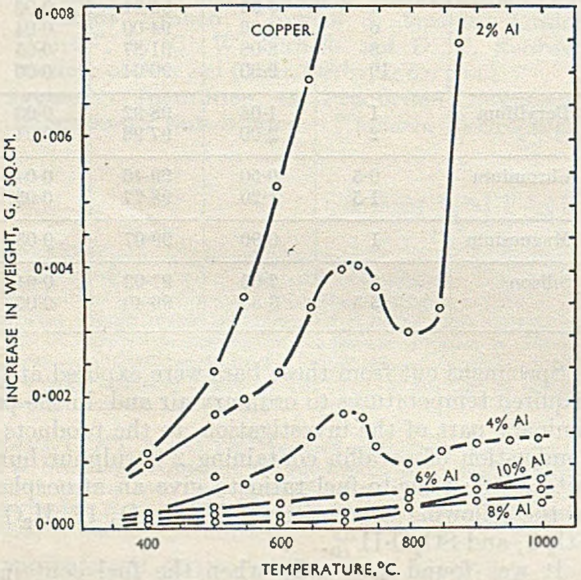


Fig. 1.—Oxidation of Copper and Copper–Aluminium Alloys After 24 Hr. at 400°–1000° C.

It seems that oxidation proceeds rapidly until a protective film of alumina is built up at the oxide/metal interface, and then oxidation practically ceases. Fröhlich[4] reported a similar result with additions of 1–3% aluminium at 800° C.

In the following discussion the scaling properties of a 2% aluminium alloy are dealt with in some detail, and those of higher aluminium content are referred to only in so far as they differ from the behaviour of the 2% aluminium alloy. In all cases the rates of attack are considerably less than those for pure copper (Fig. 1).

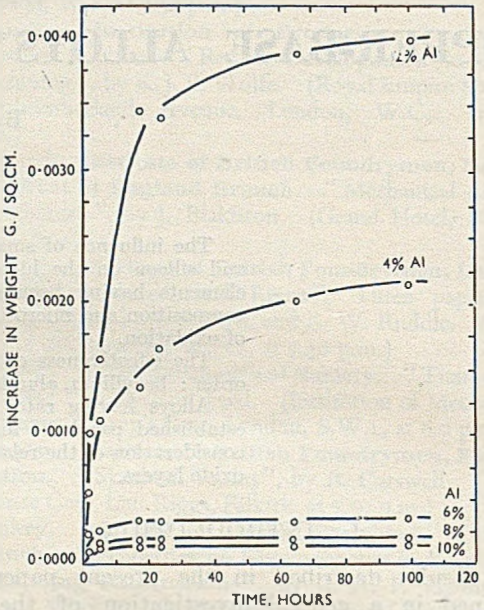


Fig. 2.—Oxidation of Copper–Aluminium Alloys After Different Times at 650° C.

It is interesting to note that an inflection similar to that in the rate curve at 750°–850° C. (Fig. 1) is found in Fig. 4, which shows the proportion of cupric oxide in the scale formed during a 24-hr. exposure. The curve for copper in Fig. 4 also shows a similar inflection. With the 2% aluminium alloy the minimum in the rate curve at about 850° C. corresponds to a more adherent type of scale and to the presence of a more continuous and compact film of alumina at the scale/metal interface.

In the temperature range 400° to approximately 800° C. the scale consists of cuprous oxide containing alumina particles in the initial stages; during further oxidation cupric oxide also forms, and with prolonged periods of exposure the proportion of cuprous oxide diminishes until finally the scale consists only of cupric oxide and alumina (Fig. 5). Above 875° C. cupric oxide becomes unstable and is not found in the scale. This scale is non-adherent, and consists of an outer layer of cuprous oxide and an inner duplex layer of cuprous oxide containing alumina particles distributed along definite boundaries, which probably correspond to the grain boundaries of the original metal (Fig. 11, Plate XXXII). In a thin, uneven subscale, in which alumina particles are found mainly at the

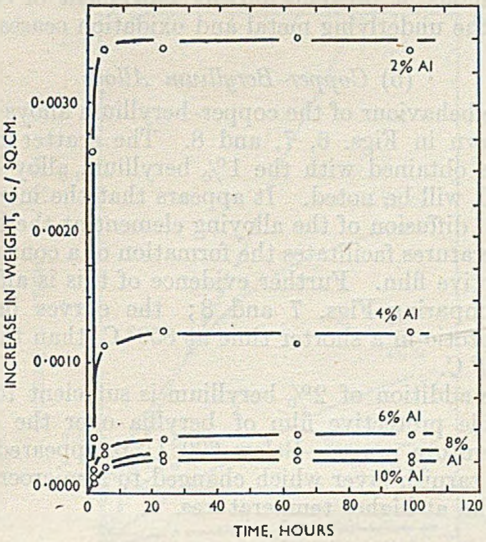


Fig. 3.—Oxidation of Copper–Aluminium Alloys After Different Times at 850° C.

grain boundaries, the impoverished metal appears to consist of pure copper. The structure of this subscale indicates that aluminium diffuses to the crystal boundary and is oxidized ahead of the advancing metal/oxide interface. The junction between the outer layer of cuprous oxide and the inner duplex layer corresponds with the original surface of the specimen.

No subscale is apparent between 750° and 850° C., and in this temperature range the proportion of alumina in the scale is greater than would correspond with the aluminium content of the alloy (Table II). (Below this temperature range there is appreciable subscale formation.) The adherent scales formed between 750° and 850° C. exhibit a continuous white film of alumina adhering to the underlying metal.

Apart from an increase in protection to be expected from a greater concentration of the alloying element, the only notable result obtained with the 4% aluminium alloy is the very considerable protection afforded above 850° C. There is also a much greater proportion of cupric oxide in the scale produced on this alloy owing to the more rapid formation of a protective film of alumina at the scale/metal interface. A corresponding improvement in the adherence of the scale takes place, especially at temperatures above 750° C.

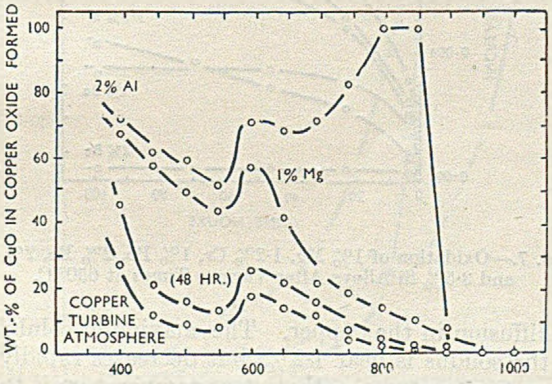


Fig. 4.—Percentage CuO in Scale Formed on Cu, 2% Al Alloy, and 1% Mg Alloy Heated for 24 Hr. at 400°–1000° C.

Table II.—*Percentage of Alloying Element Converted into Scale During 24 Hours' Oxidation in Air.*

Alloy	Oxidation Temperature, °C.									
	400	500	600	650	700	750	800	850	900	1000
2% Al .	1·9	2·1	2·0	1·9	1·9	2·5	3·1	2·4	2·0	1·9
1·2% Cr .	1·5	1·6	1·3	1·3	1·1	...	1·2	...	...	1·3
1% Mg .	0·9	...	0·9	...	1·0	1·2	1·7	1·9	2·0	...
2% Si . .	2·0	1·8	2·1	2·0	2·2	2·3	2·7	1·8	0·4	0
3·5% Si .	...	3·2	...	4·1	...	2·8	0·8	0	...	...

The scales produced on the higher-aluminium alloys were adherent, and consisted very largely of alumina over most of the temperature range, and above 750° C. were almost entirely of alumina.

The interesting feature in the oxidation of the 2% aluminium alloy is its unusual behaviour in the range 700°–850° C. Below this range oxidation follows the usual pattern, in which preferential oxidation of aluminium occurs in advance of the oxide/metal interface, giving rise to a subscale. The alumina particles give only partial protection, since they are located very largely at the pre-existing crystal boundaries of the alloy and do not form a continuous film at the oxide/metal interface.

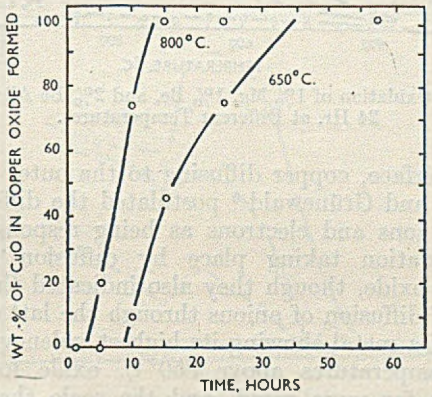


Fig. 5.—Percentage CuO in Scale Formed on 2% Al Alloy After Various Times at 650° and 800° C.

It is evident that some change occurs in the mechanism of oxidation at 700° C., since the aluminium is not preferentially oxidized in advance of the oxide/metal interface as at lower temperatures; there is, consequently, no subscale formation, and the alumina forms as a compact film between the oxide and the metal. This change in the disposition of alumina in the scale gives a higher degree of protection and results in the double inflection of the rate curve in the range 700°–850° C., as shown in Fig. 1.

As the copper oxide portion of the scale in this temperature range eventually consists entirely of cupric oxide, either oxygen must diffuse inwards to the cuprous oxide/cupric oxide interface, or cuprous oxide must dissociate into cupric oxide and copper at that interface, copper diffusing to the outer surface. Wagner and Grünewald [5] postulated the diffusion of cuprous ions and electrons as being responsible for the oxidation taking place by diffusion through cuprous oxide, though they also indicated the possibility of diffusion of anions through the lattice of an oxide of a metal showing its highest valency.

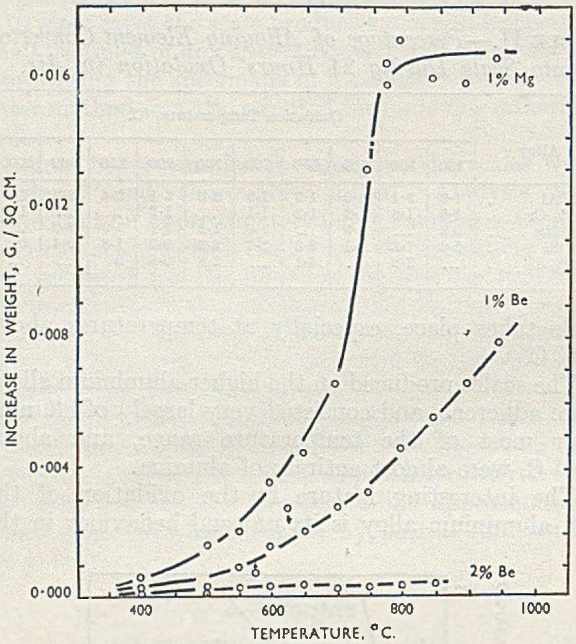


Fig. 6.—Oxidation of 1% Mg, 1% Be, and 2% Be Alloys After 24 Hr. at Different Temperatures.

At temperatures above 850° C. oxide formation becomes far more rapid, and the scale that forms consists only of cuprous oxide and alumina; a thin, irregular subscale in which the alumina particles appear as a network in the inner layer of cuprous oxide, also forms.

From the results obtained it is clear that complete protection against continued oxidation is obtained by the formation of a continuous film of alumina. Under present conditions 4% aluminium is not sufficient to provide such a film, though the irregular manner in which the 4% aluminium alloy oxidizes appears to indicate that this content is near to that required to give the almost complete protection conferred by a slightly higher aluminium content.

The progress of oxide formation with these alloys would seem to be in accord with the mechanism postulated by Mott,[6] in that the duplex layer of alumina and cuprous oxide formed in the initial stages of oxidation is transformed into an alumina film as a result of the replacement of cuprous ions by those of aluminium, the replaced copper ions diffusing outward to the oxide/gas interface. As the alumina film becomes continuous, it prevents further movement of copper from the underlying metal and oxidation ceases.

(b) *Copper–Beryllium Alloys*

The behaviour of the copper–beryllium alloys in air is shown in Figs. 6, 7, and 8. The scatter in the results obtained with the 1% beryllium alloy below 800° C. will be noted. It appears that the increased rate of diffusion of the alloying element at the higher temperatures facilitates the formation of a continuous protective film. Further evidence of this is afforded by comparing Figs. 7 and 8; the curves become asymptotic in a shorter time at 850° C. than they do at 650° C.

The addition of 2% beryllium is sufficient to give a stable protective film of beryllia over the entire temperature range. Below 600° C. it appeared as a slight tarnish layer which changed to a microcrystalline film at higher temperatures.

(c) *Copper–Chromium Alloys*

The addition of 1·2% chromium to copper confers no resistance to oxidation (Fig. 9). The structure of the oxide layer is similar to that formed on pure copper, except that the inner layer of cuprous oxide contains particles of Cr_2O_3. This alloy possesses a high susceptibility to internal oxidation of a peculiar form at high temperatures, as illustrated in Fig. 12 (Plate XXXII). The tendency for Cr_2O_3 to give a Nessler-ring effect is interesting.

The lack of protection is presumably due to the low solid solubility of chromium, coupled with a low rate of diffusion in the copper. The maximum solubility at the solidus is near 1·2%, but decreases rapidly at lower temperatures so that at no temperature is there sufficient in solution to form a continuous protective film of chromic oxide when heated.

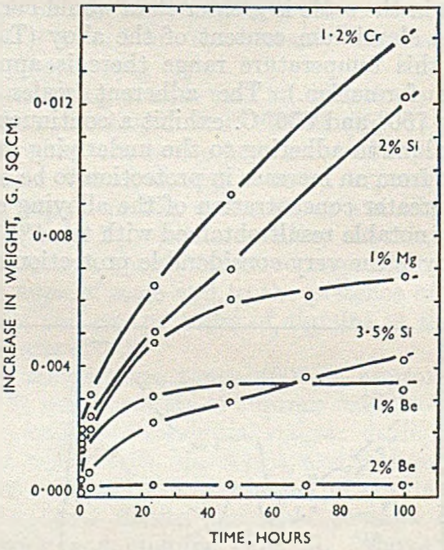


Fig. 7.—Oxidation of 1% Mg, 1·2% Cr, 1% Be, 2% Be, 2% Si, and 3·5% Si Alloys After Various Times at 650° C.

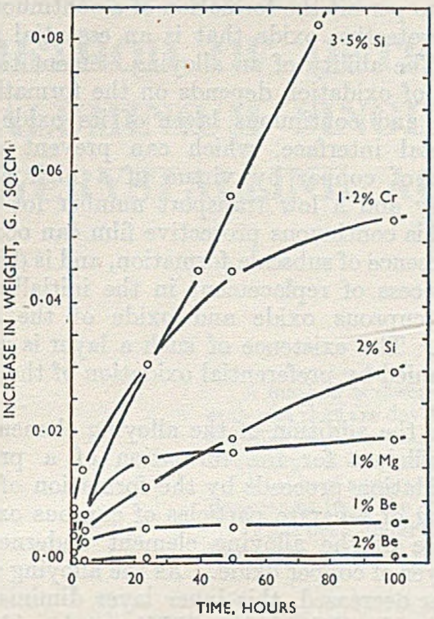


Fig. 8.—Oxidation of 1% Mg, 1·2% Cr, 1% Be, 2% Be, 2% Si, and 3·5% Si Alloys After Various Times at 850° C.

(d) *Copper–Magnesium Alloys*

Only the alloy containing 1·0% magnesium was examined in this system. Although this addition does not give a high degree of resistance, the change in direction in the oxidation curve shown in Fig. 6 at 750° C. is interesting. Apart from occurring at a higher temperature, it is similar to that found in the copper–aluminium and copper–silicon alloys.

Extensive subscale formation occurs at all temperatures up to approximately 750° C., but above this temperature it is slight and irregular (Fig. 13, Plate XXXII). The scales formed were non-adherent at all temperatures below 750° C.

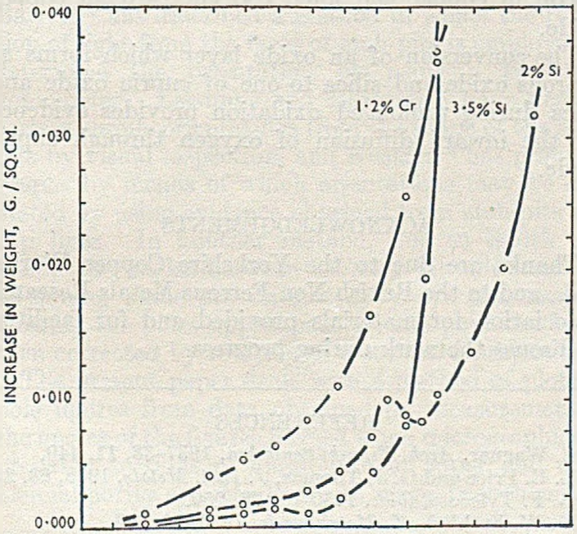


Fig. 9.—Oxidation of 1·2% Cr, 2% Si, and 3·5% Si Alloys After 24 Hr. at Different Temperatures.

(e) *Copper–Silicon Alloys*

The oxidation properties of copper–silicon exhibit some interesting features. As with the 2% aluminium alloy, the rate curve for the 2% silicon alloy (Fig. 9) shows a double inflection in the range 775°–825° C. At temperatures near 800° C. and below, a film of silica was found at the scale/metal interface that could be easily separated from the cuprous and cupric oxides by dissolving away the latter in hydrochloric acid. Above 825° C. this film of silica was no longer evident, and the inflections became more marked with increased time of exposure. Above 900° C. there is evidence of partial melting at the surface of the metal, although the melting point of the 2% silicon alloy is given in the literature as 1010° C.

Analyses of the scales formed on the 2% silicon alloy showed that cuprous and cupric oxides and silica were present in scales formed below 850° C., but that at higher temperatures the silica and cupric oxide are absent.

The changes occurring in the composition of the oxide layer formed on this 2% silicon alloy over the temperature range from 400° to 900° C. are shown in Fig. 10. The curve has two minima, one at 550° C., similar to that found in the oxidation of pure copper, and a second at 775° C., which corresponds with the maximum in the oxidation-rate curve shown in Fig. 9.

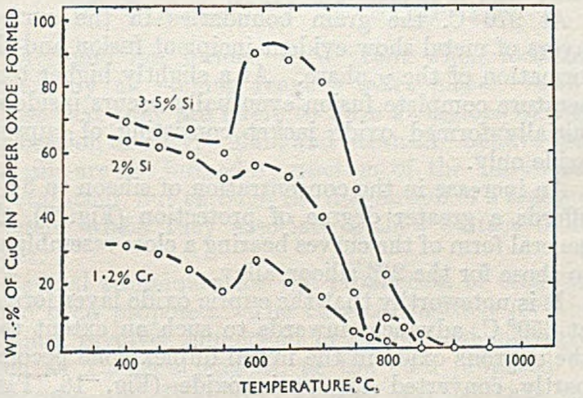


Fig. 10.—Percentage of CuO in Scale Formed on 1·2% Cr, 2% Si and 3·5% Si Alloys After 24 Hr. at Different Temperatures.

It is only in the limited temperature range 750°–850° C. that appreciable preferential oxidation of silicon occurs (Table II, p. 231), and this is at a maximum between 800° and 825° C. These observations are in agreement with those of Dunn,[7] who showed that at 725° C. a film of silica could be detected at the oxide/metal interface, but that scales formed at 827° C. contained practically no silica.

At temperatures up to 750° C. subscale formation occurs, mainly along grain boundaries, and is extensive even at 550° C. (Fig. 14, Plate XXXII). Between 750° and 775° C. no subscale was observed, but the scale/metal interface was very irregular, with deep localized penetration of scale consisting of cuprous oxide and silica (Fig. 15, Plate XXXII).

The scale/metal interface is comparatively even at 800° C. and corresponds with the minimum in the rate curve in Fig. 9 (p. 233). This even attack is presumably due to the build-up of the interface film of silica between the metal and the copper oxide layer and, as would be expected, this was associated with a much greater proportion of cupric oxide (Fig. 16, Plate XXXII). At 850° C. very little cupric oxide occurs in the scale, and the scale/metal interface again becomes irregular and the extent of the duplex zone is much diminished.

With the absence of silica in scale formed at 900° and 950° C., there is an increase in the concentration of silicon in the metal at the scale/metal interface which results eventually in the formation of a two-phase zone in the underlying metal (Fig. 17, Plate XXXII). The hard, bluish constituent in this zone appears to consist of the γ phase. Analysis of the metal in the surface layers shows a large increase in the proportion of silicon present, and also that only copper and silicon are present, thus ruling out the possibility of melting being due to either formation of copper silicate or some kind of subscale.

At 970° C. the grain boundaries in the surface layers of metal show evident incipient fusion and the formation of the γ phase. At a slightly higher temperature complete fusion eventually occurs inside an initially formed oxide jacket consisting of cuprous oxide only.

An increase in the concentration of silicon to 3·5% affords a greater degree of protection (Fig. 9), the general form of the curves bearing a close resemblance to those for the 2% silicon alloy.

It is noteworthy that the cupric oxide layer formed at 650° C. advances inwards to such an extent that the cuprous oxide in the initial duplex zone becomes partly converted to cupric oxide (Fig. 18, Plate XXXII). This may be regarded as further evidence of film growth continuing by inward diffusion of oxygen through cupric oxide.

There is a general similarity in the oxidation behaviour of the copper–aluminium and copper–silicon alloys, especially in the occurrence of maxima and minima in the rate curves. One important difference, however, is the more extensive subscale formation in the copper–silicon series at lower temperatures, which results in a higher scaling rate than that found with the copper–aluminium alloys.

III.—CONCLUSIONS

The results obtained in the present investigation demonstrate certain features in the oxidation of copper-rich alloys containing small additions of beryllium, aluminium, magnesium, silicon, and chromium. The effectiveness of these additions in conferring resistance to oxidation decreases in the order given. Chromium affords no protection whatsoever, the solid solubility of chromium in copper being too low to permit the formation of a continuous layer of the protective oxide that is an essential requirement. The ability of an alloying element to reduce the rate of oxidation depends on the formation of a separate and continuous layer of its oxide at the scale/metal interface, which can prevent outward diffusion of copper by virtue of a high electrical resistivity and a low transport number for cuprous ions. This continuous protective film can occur only in the absence of subscale formation, and is dependent on a process of replacement in the initially duplex film of cuprous oxide and oxide of the alloying element. The existence of such a layer is generally accompanied by preferential oxidation of the alloying element.

Where the addition of the alloying element is not quite sufficient for the formation of a protective film, oxidation proceeds by the formation of a scale consisting of separate particles of cuprous oxide and the oxide of the alloying element underneath the outer layer of copper oxide. As the alloying addition is further decreased, this inner layer diminishes and may be replaced by a zone of internal oxidation in which preferential oxidation of the added element occurs at the grain boundaries of the alloy.

The behaviour of the copper–silicon is most remarkable in that at 900° C. and above the silicon diffuses into the metal away from the oxide/metal interface, whereas at lower temperatures, e.g. 800° C., the silicon diffuses towards the surface of the oxidizing metal to form a layer of silica at the oxide/metal interface.

In considering the results obtained with these alloys reference must be made also to the inflections found in the oxidation-rate and oxide-composition curves with increasing temperature. It is considered that these inflections result from an interplay of temperature-sensitive properties such as diffusion of silicon in copper and ionic transfer through copper oxide.

The conversion of an oxide layer which forms as cuprous oxide and silica to one of cupric oxide and silica during prolonged oxidation provides evidence for the inward diffusion of oxygen through cupric oxide.

ACKNOWLEDGEMENTS

Thanks are due to the Yorkshire Copper Works, Ltd., and to the British Non-Ferrous Metals Research Association for materials provided and for facilities to discuss the work during progress.

REFERENCES

1. C. Wagner, *Arch. Eisenhüttenwesen*, 1937–38, **11**, 449.
2. L. E. Price and G. J. Thomas, *J. Inst. Metals*, 1938, **63**, 21.
3. R. F. Tylecote, *ibid.*, 1950–51, **78**, 259.
4. K. W. Fröhlich, *Z. Metallkunde*, 1936, **28**, 368.
5. C. Wagner and K. Grünewald, *Z. physikal. Chem.*, 1938, [B], **40**, 445.
6. N. F. Mott, *Nature*, 1940, **145**, 996.
7. J. S. Dunn, *J. Inst. Metals*, 1931, **46**, 25.

1437

A METHOD OF DETERMINING ORIENTATIONS IN ALUMINIUM SINGLE CRYSTALS AND POLYCRYSTALLINE AGGREGATES

By G. E. G. TUCKER,† B.Sc., STUDENT MEMBER, and P. C. MURPHY,† B.Sc., A.I.M., MEMBER

SYNOPSIS

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 aluminium. Suitable etching reagents are indicated, and the technique of angular measurement using a metallurgical 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.

I.—INTRODUCTION

It has been known for many years that when metals and alloys are etched with certain reagents, attack takes place at isolated points within the grains rather than by general surface etching. This localized etching produces pits whose traces on the etched surface are rectilinear figures, commonly known as etch-pits. It has also been shown that in most cases these pits are formed by attack along definite crystallographic planes; for instance, in most face-centred cubic metals etching is along {100} planes. It will therefore be seen that the positions of the sides or faces of etch-pits are related to the orientations of the grains on which they are formed.

This property has been used by many investigators to determine crystal orientations. For instance, Barrett [1] has described a method in which the reflection of light from the faces of etch-pits is employed to determine the orientation by using a two-circle goniometer. Lacombe and Beaujard [2] have derived certain simple orientations from the shape of the etch-pits by visual inspection, and Kostron [3] has prepared graphs by means of which orientations may be estimated by using evidence obtained from etch-pits and slip lines. In another method, due to Smith and Mehl,[4] the angles between the etch-pit sides and the sectioning plane are estimated by eye and used in plotting the poles approximately, the positions being then corrected by rotation of a standard projection.

The present paper deals with a method of plotting pole figures from data obtained by measurement of the angles of the figures formed when micrographically prepared sections of aluminium of super and commercial purity are attacked with reagents such as that of Tucker.[5] The measurements may be carried out on any ordinary metallurgical microscope.

II.—FORMATION OF ETCH-PITS

Etch-pits are formed in a grain when localized attack by an etching reagent takes place. Cahn [6] states that they are likely to form at regions of high dislocation density. Mahl and Stranski [7] believe that the pits are not formed by reaction of the aluminium crystal itself, but through the dissolution of a reaction product which they associate with a surface oxide layer.

Several techniques for the production of etch-pits have been proposed. The classical method for aluminium involves immersion in strong acid mixtures, such as Tucker's reagent, or the etch proposed by Lacombe and Beaujard.[2] Other methods are the electrolytic etch of Jacquesson and Manenc [8] and the gaseous etch used by Mahl and Stranski.[9]

If the faces of the etch-pits are in fact cube planes, the pits will appear as simple rectilinear figures; Mahl and Stranski, however, consider that in some cases the true cube planes are not revealed, but instead "vicinal surfaces", which may vary by as much as 10° from the cube planes, and Kostron [3] shows photographs of etch-pits with curved sides. In such cases orientation determinations based on the assumption that cube planes are revealed could not be made.

It has been found that these objections can be avoided by the use of suitable etching reagents on materials of fairly high purity such as commercial and superpure aluminium; any appreciable quantity of alloying elements gives many pits that are bounded by curved surfaces.

---

* Manuscript received 14 August 1952.

† Research Metallurgist, Aluminium Laboratories Limited, Banbury.

Assuming that the etching reagent attacks along the {100} planes of the lattice, it can be seen that the etch-pits must be sections of cubes. If two of the crystallographic axes lie parallel to the sectioning plane then the etch-pit will be square or rectangular, depending upon the relative rates of attack parallel to these axes; in any case the pit formed will have a flat base parallel to the sectioning plane. If only one of the crystallographic axes is parallel to the sectioning plane, the etch-pit will be rectangular in form but will have a bottom edge, i.e. two of the faces of the pit will slope down to the base and meet in a line parallel to two of the sides of the rectangle and normal to the other two. If none of the three axes is parallel to the sectioning plane (the most general case), the etch-pit will have the form of a triangle with the three faces running down to meet in a point. The three types of ideal pits and their formation are illustrated diagrammatically in Fig. 1 (*a*), (*b*), and (*c*) and photographs are shown in Figs. 9, 10, and 11 (Plate XXXIII). Fig. 9 shows (001)-type pits under dark-field illumination, Fig. 10 (011)-type pits (the microscope being focused on the "bottom edge" of the pits), and Fig. 11 (111)-type pits. These pit shapes are, of course, formed only under ideal conditions; in practice less perfect shapes are frequently observed, especially in aluminium of low purity. The greatest divergence from the ideal shapes occurs in triangular etch-pits, which are frequently truncated at one or more of the apices producing 4-, 5-, or 6-sided figures. These truncations are, however, always parallel to the sides opposite to them and obviously occur by simultaneous attack on parallel {100} planes. The formation of such truncated pits on an equilateral triangle is shown diagrammatically in Fig. 2 (*a*), and some truncated pits found in practice are illustrated in Fig. 12 (Plate XXXIII). A rather less obvious case is illustrated in Fig. 2 (*b*), which shows a truncation of a very acute triangle. The etch-pit produced in this way may appear to be rectangular, but careful angular measurement (as described later) will distinguish between the two. A still more misleading case occurs when attack along one of the {100} planes fails to reach the surface of the specimen, the etch-pit then having a rhombic instead of a triangular form. This is illustrated in Fig. 2 (*c*), and is shown on some of the pits in Fig. 12 (Plate XXXIII), where the truncations have resulted in the almost complete disappearance of the third side. It is not possible to measure such pits with great accuracy, but it has been found in practice that an assumption that the triangle is isosceles is rarely in great error. It is advisable to carry out a diligent search of any grain in which rhombic etch-pits occur in case there are some isolated instances of incomplete truncation which reveal the ideal triangle.

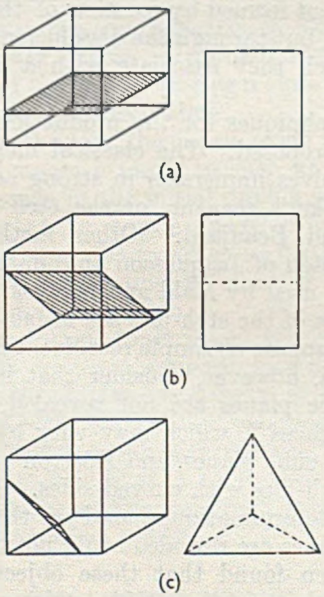


Fig. 1.—Formation of Simple Etch-Pits by Intersection of Sectioning Plane (shaded) with Lattice Cube, and Shape of Etch-Pit as seen under a Microscope. Broken lines indicate bottom edges. (*a*) Two axes in sectioning plane—(001) type. (*b*) One axis in sectioning plane—(011) type. (*c*) No axis in sectioning plane—(111) type.

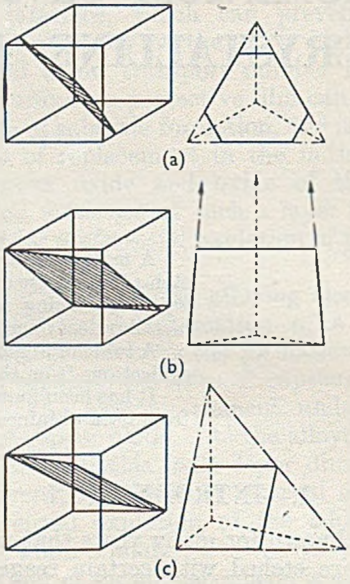


Fig. 2.—Formation of Truncated Etch-Pits by Intersection of Sectioning Plane (shaded) with Lattice Cubes, and Shape of Etch-Pits (full lines) as seen under a Microscope. Dashed lines indicate ideal triangles; dotted lines indicate bottom edges. (*a*) Truncated (111) type. (*b*) Truncated "near (011)" type. (*c*) Rhombic—truncated (112) type.

III.—TECHNIQUE OF MEASURING ANGLES OF ETCH-PITS

1. Preparation of Specimens

In order to obtain reliable data it is essential that the pits should be straight sided, as large as possible, and well spaced on any grain. These requirements are met only by specialized polishing and etching techniques. Each sample to be examined presents a different problem, and only the simpler cases can be prepared by any general technique.

The easiest samples to prepare satisfactorily are those with a relatively large grain-size, e.g. as-cast aluminium of commercial- or super-purity. These can be given a good metallographic polish either electrolytically or mechanically and etched in strong acid mixtures such as the etch proposed by Lacombe and Beaujard,[2] which contains 47% fuming nitric acid, 50% chemically pure hydrochloric acid, and 3% pure hydrofluoric acid, or a similar etch used by the present authors consisting of 35% nitric acid, 61% hydrochloric acid, and 4% hydrofluoric acid (all of Analar quality).

It is generally advisable to use electrolytic polishing, but reasonably good results can be obtained from a mechanically polished specimen if care is taken to avoid undue cold working of the surface. The time of etching depends mainly on the method of preparation and the temperature of the reagent. It is advisable to use a container made of a relatively inert substance such as polythene to avoid possible contamination of the etch. It has also been found desirable to cool the reagent to a temperature of 0°–8° C. in order to increase the etching time to 5–15 sec.

A more difficult problem is presented by relatively fine-grained polycrystalline aggregates, such as cold-rolled or annealed commercial-purity aluminium sheet. In these the etch-pits developed by the above reagents are often of very poor shape, especially if the sample has been mechanically polished. For such materials it has been found advisable to use an electrolytic polish such as that developed by Hone and Pearson [10] and to etch, after washing thoroughly in cold water, in a reagent such as 71% hydrochloric acid, 4% hydrofluoric acid, and 25% ethyl or methyl alcohol, or 42% hydrochloric acid, 41% glycerol, 15% nitric acid, and 2% hydrofluoric acid (all of Analar quality). With the latter etching solution, the reagents should be mixed in the order given, to avoid undue rise in temperature.

These reagents are also used at 0°–8° C., and the time of etching for commercial-purity sheet is about 7–10 sec. For super-purity sheet the reagents recommended for large-grained specimens are more suitable, as those containing organic additions are too slow.

The section examined should be related to some logical reference directions, as is usual in X-ray work. For example, in investigating fibre textures in rods and wires, a transverse section is desirable; in sheet, a section in the rolling plane and related to the rolling direction is most useful. The polished section can be marked with a scribed line parallel to a reference direction.

2. Measurement of Etch-Pits

As described in Section II, etch-pits in aluminium fall broadly into two classes: rectangular pits and those based on triangles. In the latter type, the internal angles of the ideal triangle must be measured, whereas in the former type it is necessary to measure the relative distances of the bottom edge from the two sides of the etch-pit parallel to it. In addition, with both types some measurement must be made to distinguish between geometrically similar pits differing only by a rotation about an axis perpendicular to the sectioning plane.

Consider first a rectangular pit. In this case it is necessary to measure the relative lengths QU and UT (Fig. 3). The ratio QU/UT will for the time being be arbitrarily referred to as the ratio l^2/k^2, for reasons which will be given in the next section. In large pits it may be possible to measure the lengths directly by means of an eye-piece graticule, but this is frequently impracticable and it is necessary to estimate the ratio by eye. This can be done fairly accurately for values of the ratio near to unity, and since the error in pole position decreases with increasing values of the ratio, estimation is generally satisfactory. The ratio l^2/k^2 is sufficient to determine the crystallographic indices of the sectioning plane, and the actual orientation is then defined uniquely by the angle between the reference direction and the bottom edge.

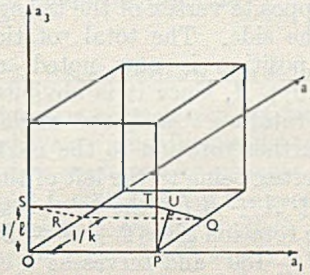


Fig. 3.—Intersection of Sectioning Plane $QRST$ of Indices $(0kl)$ with Lattice Cube.

Consider next the triangular pits. In this case it is necessary and sufficient to measure two angles of the triangle (the ideal triangle if the etch-pit is truncated) to determine the crystallographic indices of the sectioning plane, and the actual orientation may be specified with the knowledge of the angle between the reference direction and one of the etch-pit sides.

The actual process of measuring and recording for a particular sample may now be described in general terms, and may be adapted to suit individual circumstances. First, the reference direction must be marked on the polished and etched sample. This is best accomplished by scribing a fine line on the specimen. The specimen is then mounted on the stage of the microscope so that the reference direction lies parallel to a vertical cross-hair in the eye-piece or on the projection screen. Measurements may then be made at the desired number of points on the specimen, but in moving from one point to another the specimen must not be rotated relative to the stage. At each point, the specimen is rotated until one side of the etch-pit being measured is parallel to the cross-hair and the angle of rotation is recorded. Throughout this description it is assumed that the specimen is rotated in such a way that the field of view rotates clockwise relative to the cross-hair. In rectangular pits this is the only measurement required (other than the ratio l^2/k^2) and is made such that the rotation brings the bottom edge parallel to the cross-hair, so that the ratio l^2/k^2 when taken from left to right is greater than unity. For example, in Fig. 3 the rotation would bring OP vertical rather than PO since $QU/UT > 1$. In rectangular (square) pits with no bottom edge $l^2/k^2 = \infty$, and any side may be taken as the bottom edge. In triangular pits, the vertex opposite to the side brought parallel to the cross-hair must be to the left of this side. In all cases the angle of rotation is measured and will be denoted here by θ.

Since all measurements necessary for a rectangular pit have been taken at this stage, no further rotation of the specimen is necessary. The specimen is returned to its original position (zero rotation) and moved to the next pit to be measured.

For triangular pits further rotation is necessary. In this case, after the angle θ has been noted, the specimen is rotated farther until the next side of the etch-pit, in order, is parallel to the cross-hair, but in this case the opposite vertex of the triangle will be to the right of the side. The total rotation from the original zero position is now noted, and will be referred to as $\theta + A$, since it is obviously made up of the initial rotation θ and one angle (A) of the triangle. A further rotation to the next side of the triangle (the vertex being to the left of this side) gives the reading $\theta + A + B$ for the total rotation so far, while one more rotation gives $\theta + A + B + C$, which should equal $\theta + 180°$ and serves as a check on the accuracy of the measurements. The angles A, B, and C of the triangle may be obtained by subtraction. The transformation necessary to plot poles on a stereographic projection is described in the next section.

It is necessary to measure the angles as accurately as possible, particularly for triangles which have angles near to 0° or 90°, since in these regions the error in pole position may be up to ten times the error in measurement. If, however, the pits are bounded by perfectly straight sides, it is found that measurements can be reproduced to within $\frac{1}{4}°$.

IV.—RELATIONSHIPS BETWEEN ETCH-PIT SHAPES AND ORIENTATION

It is proposed to demonstrate that the positions in space of the cube axes of the lattice of any grain are unambiguously related to the angles of triangular pits and the ratio l^2/k^2 in rectangular pits as discussed above.

1. Triangular Etch-Pits

If the indices of the sectioning plane are (hkl), then the intercepts made by that plane on the three crystallographic axes will be proportional to $\frac{1}{h}, \frac{1}{k}, \frac{1}{l}$, respectively (Fig. 4). The directions AB, BC, CA are thus $[\bar{k}h0]$, $[0\bar{l}k]$, $[l0\bar{h}]$ respectively, leading to the relations:

$$\left.\begin{aligned} \cos C\widehat{A}B &= \frac{kl}{\sqrt{(k^2+h^2)(l^2+h^2)}} \\ \cos A\widehat{B}C &= \frac{hl}{\sqrt{(h^2+k^2)(l^2+k^2)}} \\ \cos B\widehat{C}A &= \frac{hk}{\sqrt{(k^2+l^2)(h^2+l^2)}} \end{aligned}\right\} \quad . \quad . \quad (1)$$

These three angles will be denoted by A, B, and C, respectively. From equations (1) the following relation, which is explicit in h, k, and l, is found:

$$\frac{h^2}{\tan A} = \frac{k^2}{\tan B} = \frac{l^2}{\tan C} \quad . \quad . \quad . \quad (2)$$

Relation (2) makes it possible to calculate the values of $\left|\frac{k}{h}\right|$ and $\left|\frac{l}{h}\right|$ from the measured values of A, B, and C.

The impossibility of assigning either sign or absolute magnitude to h, k, or l other than by definition is to be expected. Consequently, h may be assigned any convenient value, preferably chosen so that h, k, and l are integral or nearly so.

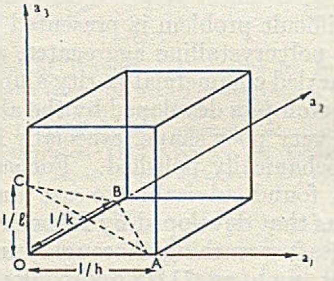


Fig. 4.—Intersection of Sectioning Plane ABC of Indices (hkl) with Lattice Cube.

Thus, from the measurement of its etch-pit angles, the texture in any grain exhibiting triangular pits may be described as possessing an (hkl) plane in the sectioning plane. In order to specify the texture more precisely, information is required on the rotation from some standard position of this (hkl) plane about the normal to the sectioning plane. This information is usually given by stating which crystallographic direction [uvw] is parallel with some reference direction. In the case of etch-pits, the angle θ between the reference direction and one side of the etch-pit is known. Suppose that this side is AB. Then, if the reference direction is [uvw],

$$\left.\begin{aligned} \cos\theta &= \frac{-uk + vh}{\sqrt{(u^2+v^2+w^2)(k^2+h^2)}} \\ 0 &= uh + vk + wl \end{aligned}\right\} \quad . \quad . \quad (3)$$

Solution of equations (3) leads to $\frac{v}{u}$ and $\frac{w}{u}$. Two solutions will generally be obtained for [uvw]. The correct one may be discovered by inspection, or more rigorously by taking that solution which satisfies the relation :

$$\cos(\theta + A) = \frac{-ul + wh}{\sqrt{(u^2 + v^2 + w^2)(l^2 + h^2)}} \quad . \quad (4)$$

Thus it is possible to represent the orientation of each grain by a code similar to that used for quoting rolling textures, i.e. (hkl) $[uvw]$, where the plane (hkl) lies parallel to the sectioning plane and the direction $[uvw]$ lies parallel with the reference direction, which lies in the sectioning plane.

It is often convenient to present the results in the form of a pole figure, preferably without first having to calculate (hkl) and $[uvw]$ and then rotating a standard projection. This may be done by consideration of the geometry of the etch-pits and pole figures. It is most convenient to plot (100) pole figures projected on to the sectioning plane of the specimen, since the traces of the {100} planes in the sectioning plane are already known as the sides of the etch-pit. The pole of any {100} plane lies on that great circle of the reference sphere which projects stereographically as a diameter of the pole figure perpendicular to the appropriate etch-pit side, its position along that diameter being determined by the angle between the sectioning plane and the particular {100} plane concerned. The direction to be taken from the centre along the appropriate diameter has yet to be fixed, and this may be accomplished by arbitrary rules. It is convenient to proceed by imagining a drawing of the etch-pit to be superimposed on the pole figure in such a position that the circumcentre of the triangle is coincident with the centre of the basic circle, as illustrated in Fig. 5; the diameters may then be replaced by radii which are perpendicular bisectors of the sides of the etch-pit, there now being only one position on each radius which lies at any given angle from the centre of the pole figure. The opposite set of directions is equally correct, the two corresponding to plotting from alternative hemispheres of the original reference sphere. In addition, the two may be taken together to give a plot corresponding to that of both hemispheres superimposed, as is common in X-ray pole-figure determinations. The latter procedure is, of course, not permissible with single crystals, or groups of only a few crystals.

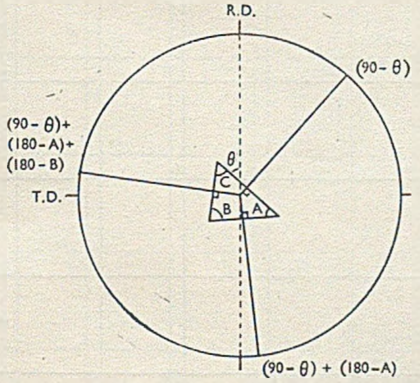


Fig. 5.—Positions of Radii Perpendicular to Sides of Etch-Pit.

It remains to determine the angles that define the positions of the poles along the radii. Since these angles are in fact the angles between the sectioning plane and the three cube planes, they may be determined from h, k, and l, and since h, k, and l are known in terms of the original angular measurements of A, B, and C, it follows that the required angles may be expressed in terms of A, B, and C. The precise relations are :

$$\left.\begin{aligned} \sec^2\alpha &= \tan B \,.\, \tan C \\ \sec^2\beta &= \tan A \,.\, \tan C \\ \sec^2\gamma &= \tan A \,.\, \tan B \end{aligned}\right\} \quad . \quad . \quad . \quad (5)$$

where α, β, and γ are the angles between the sectioning plane and the cube planes whose traces lie opposite angles A, B, and C, respectively. Hence the three cube poles plot at angles α, β, and γ from the centre of the pole figure along the radii $450 - (\theta + A + B)$, $90 - \theta$, $270 - (\theta + A)$, respectively, measured clockwise from the reference direction.

2. Rectangular Etch-Pits

If an etch-pit is rectangular, at least one of the crystallographic axes must lie parallel to the sectioning plane. Consider such an arrangement, which is illustrated in Fig. 3 (p. 237), where the sectioning plane is $QRST$ with crystallographic indices $(0kl)$ and the three axes are $a_1a_2a_3$. The axis a_1 is parallel to the plane $QRST$. Now the true etch-pit shape, in three dimensions, is the right triangular prism $ORSPQT$, and as shown earlier, when viewed from a direction perpendicular to the plane $QRST$, the bottom edge OP appears as a line, parallel to ST and RQ, cutting TQ at U, where PU is perpendicular to TQ. Since triangles PQT, UQP, and UPT are all similar, it follows that :

$$\frac{QU}{UT} = \left(\frac{PQ}{PT}\right)^2 \quad . \quad . \quad . \quad . \quad (6)$$

i.e.

$$\frac{k^2}{UT} = \frac{l^2}{QU} \quad . \quad . \quad . \quad . \quad . \quad (7)$$

Since the ratio QU/UT may be determined it follows that the value of $\left|\frac{l}{k}\right|$ may also be determined. In this case $h = 0$, and k may be taken as any convenient value. The texture may then be quoted as $(0kl)[uvw]$ in a manner similar to that employed for triangular etch-pits. In this case :

$$\left.\begin{aligned} \cos\theta &= \frac{+u}{\sqrt{u^2 + v^2 + w^2}} \\ 0 &= vk + wl \end{aligned}\right\} \quad . \quad . \quad (8)$$

the complementary relation being :

$$\sin\theta = \frac{-vl + wk}{\sqrt{(u^2 + v^2 + w^2)(l^2 + k^2)}} \quad . \quad (9)$$

As before, the poles can be plotted on a stereographic projection, and in this case there are four radii instead of three on which poles lie. This arises because the sectioning plane is at 90° to the (100) plane. The angles from the centre at which the other two poles lie are, from Fig. 3, $\widehat{UTP}$, $\widehat{UQP}$, which are respectively equal to $\tan^{-1}\frac{l}{k}$ and $\tan^{-1}\frac{k}{l}$.

Thus, the poles of the {100} planes plot at $\tan^{-1}\frac{l}{k}$, 90°, $\tan^{-1}\frac{k}{l}$, 90° from the centre along radii $90 - \theta$, $180 - \theta$, $270 - \theta$, and $360 - \theta$, respectively.

3. SQUARE ETCH-PITS

These may be conveniently dealt with as a special case of rectangular etch-pits when two crystallographic axes lie parallel to the sectioning plane. Thus one of the cube planes is parallel to the sectioning plane. If, in the calculations for rectangular etch-pits, $k = 0$, then the sectioning plane may be represented as (001), and the reference direction as [*uvw*], where

$$\left.\begin{aligned}\cos\theta &= \frac{+\,u}{\sqrt{u^2 + v^2}}\\ 0 &= w\end{aligned}\right\} \quad . \quad . \quad . \quad (10)$$

TABLE I.—*Angles (α, β, γ) Between Cube Planes and*

Smallest Etch-Pit Angle (*A*)	Largest Etch-Pit Angle (*C*)														
	60	*61*	*62*	*63*	*64*	*65*	*66*	*67*	*68*	*69*	*70*	*71*	*72*	*73*	*74*
1															
2															
3															
4															
5															
6															
7															
8															
9															
10															
11															
12															
13															
14															
15															
16															
17															
18															
19															
20															
21															
22															
23															
24															
25															
26															
27															
28															
29															
30															
31															
32															73 47 47
33															73 48 47
34														72 48 48	72 49 46
35														72 49 47	72 50 46
36													71 48 48	71 50 47	71 51 45
37													70 49 47	71 50 46	71 52 44
38												70 48 48	70 50 47	70 51 46	70 53 44
39												69 49 48	69 51 46	69 52 45	70 54 44
40											69 49 49	69 50 47	69 52 46	69 53 45	69 54 43
41											68 50 48	68 51 47	68 52 46	68 54 44	69 55 43
42										67 49 49	67 51 48	68 52 47	68 53 45	68 54 44	68 56 43
43										67 50 49	67 51 48	67 53 46	67 54 45	67 55 44	68 56 42
44									66 50 50	66 51 48	66 52 47	66 53 46	67 55 45	67 56 43	67 57 42
45									66 51 49	66 52 48	66 53 47	66 54 46	66 55 44	66 56 43	67 58 42
46								65 50 50	65 51 49	65 53 48	65 54 47	65 55 45	65 56 44	66 57 43	66 58 42
47								64 51 50	64 52 49	64 53 48	65 54 46	65 56 45	65 57 44	65 58 43	66 59 42
48							64 51 51	64 52 50	64 53 49	64 54 47	64 55 46	64 56 45	64 57 44	65 58 43	65 59 42
49							63 52 50	63 53 49	63 54 49	63 55 47	63 56 46	64 57 45	64 58 44	64 59 43	64 60 41
50						62 51 51	62 52 50	62 53 49	62 54 48	63 55 47	63 56 46	63 57 45	63 59 44	64 60 42	64 61 41
51						63 52 51	62 53 50	62 54 49	62 55 48	62 56 47	62 57 46	62 58 45	63 59 44	63 60 42	63 61 41
52					61 52 52	61 53 51	61 54 50	61 55 49	61 56 48	61 57 47	62 58 46	62 59 45	63 60 43	62 61 42	63 62 41
53					60 53 52	60 54 51	60 55 50	60 56 49	60 57 48	61 57 47	61 58 46	61 59 45	62 60 43	62 61 42	62 62 41
54				59 53 53	59 53 52	59 54 51	60 55 50	60 56 49	60 57 48	60 58 47	60 59 46	61 60 45	61 61 43		
55				59 53 52	59 54 51	59 55 51	59 56 50	59 57 49	59 58 48	59 59 47	60 60 46				
56			58 53 53	58 54 52	58 55 51	58 56 50	58 57 50	58 58 49	59 59 48						
57			57 54 53	57 55 52	57 56 51	57 57 50	57 57 49								
58		56 54 54	56 55 53	56 56 52	56 56 51										
59		56 55 54	56 56 53												
60	55 55 55														

and

$$\sin\theta = \frac{-v}{\sqrt{u^2+v^2}} \quad . \quad . \quad . \quad (11)$$

enabling the sign of v/u to be determined. For the purpose of plotting the orientation on a pole figure, k may again be put equal to 0, in which case $UT/QU = 0$, i.e. no bottom edge is visible. The angles $U\widehat{T}P$ and $U\widehat{Q}P$ become 90° and 0°, respectively. The poles in this case then plot at the centre and at 90° from the centre on each of the radii $90 - \theta$, $180 - \theta$, $270 - \theta$, and $360 - \theta$.

V.—CALCULATION OF ANGULAR CO-ORDINATES OF {100} POLES

As shown in the previous section, the angular coordinates of the {100} poles may be calculated directly from the measurements made of the original etch-pit angles. The first operation is to obtain the positions of the radii of the pole figure on which the poles lie by simple subtractions from multiples of 90°. Thus the radii for triangular pits lie at angles $450 - (\theta + A + B)$, $90 - \theta$, and $270 - (\theta + A)$ to the reference direction. In rectangular and square pits the radii lie at angles $90 - \theta$, $180 - \theta$, $270 - \theta$, and $360 - \theta$.

Sectioning Plane for Various Etch-Pit Angles.

	Largest Etch-Pit Angle (*C*)														
Smallest Etch-Pit Angle (*A*)	*75*	*76*	*77*	*78*	*79*	*80*	*81*	*82*	*83*	*84*	*85*	*86*	*87*	*88*	*89*
1															
2															89 45 45
3															89 55 35
4														88 45 45	88 60 30
5														88 51 39	88 63 27
6													87 45 45	87 55 35	88 66 24
7													87 49 41	87 58 32	88 68 22
8												86 45 45	86 52 38	87 60 30	87 69 21
9												86 48 42	86 55 35	86 62 28	87 71 20
10											85 45 45	85 50 39	85 57 33	86 64 27	87 72 19
11											85 48 43	85 53 37	85 59 32	86 65 26	87 73 18
12										84 45 45	84 50 41	84 55 36	85 60 30	86 66 24	87 73 17
13										84 48 43	84 52 39	84 57 34	84 62 29	85 67 23	87 74 16
14									83 45 45	83 50 41	83 54 37	84 58 33	84 63 28	85 68 23	86 75 16
15									82 47 44	83 51 40	83 55 36	83 59 32	84 64 27	85 69 22	86 75 15
16								82 46 46	82 49 42	82 53 39	83 56 35	83 60 31	84 65 26	85 70 21	86 76 15
17								81 47 44	82 51 41	82 54 37	82 58 34	83 61 30	83 66 25	84 70 21	86 76 14
18							81 46 46	81 49 43	81 52 39	81 55 36	82 59 33	82 62 29	83 66 25	84 71 20	86 77 14
19							80 47 44	81 50 41	81 53 38	81 56 35	82 60 32	82 63 28	83 67 24	84 71 20	86 77 14
20						80 46 46	80 49 43	80 52 40	80 55 37	81 57 34	81 61 31	82 64 27	83 68 24	84 72 19	86 77 13
21						79 47 45	79 50 42	80 53 39	80 56 37	80 58 33	81 62 30	82 65 27	83 68 23	84 72 19	85 78 13
22					79 46 46	79 49 44	79 51 41	79 54 38	80 57 35	80 59 33	81 62 30	81 65 26	82 69 23	84 73 18	85 78 13
23					78 47 45	78 50 43	79 52 40	79 55 37	79 57 35	80 60 32	80 63 29	81 66 26	82 69 22	83 73 18	85 78 13
24				78 46 46	78 49 44	78 51 42	78 53 39	78 56 37	79 58 34	79 61 31	80 64 28	81 67 25	82 70 22	83 74 18	85 79 12
25				77 48 45	77 50 43	77 52 41	78 54 38	78 57 36	79 59 33	79 62 31	80 64 28	81 67 25	82 70 21	83 74 17	85 79 12
26			77 47 47	77 49 44	77 51 42	77 53 40	77 55 38	78 58 35	78 60 33	79 62 30	79 65 28	80 68 24	81 71 21	83 75 17	85 79 12
27			76 48 46	76 50 43	76 52 41	77 54 39	77 56 37	77 58 35	78 61 32	78 63 30	79 66 27	80 69 24	81 71 21	83 75 17	85 79 12
28		76 47 47	76 49 45	76 51 43	76 53 41	76 55 39	77 57 36	77 59 34	77 61 32	78 64 29	79 66 27	80 69 24	81 72 21	83 75 17	85 80 12
29		75 48 46	75 50 44	75 52 42	75 54 40	76 56 38	76 58 36	77 60 34	77 62 31	78 64 29	79 67 26	80 69 24	81 72 20	82 75 17	84 80 12
30	74 47 47	74 49 45	75 51 43	75 53 42	75 55 40	75 57 38	76 59 36	76 61 34	77 63 31	78 65 29	78 67 26	79 70 23	81 72 20	82 76 16	84 80 11
31	74 48 46	74 50 44	74 52 43	74 54 41	75 55 39	75 57 37	75 59 35	76 61 33	76 63 31	77 65 28	78 68 26	79 70 23	80 73 20	82 76 16	84 80 11
32	73 49 46	73 51 43	74 53 42	74 54 40	74 56 38	75 58 36	75 60 34	76 62 32	76 64 30	77 66 28	78 68 25	79 70 23	80 73 20	82 76 16	84 80 11
33	73 50 45	73 52 43	73 53 42	73 55 40	74 57 38	74 59 36	75 60 34	75 62 32	76 64 30	77 66 28	78 68 25	79 71 23	80 74 19	82 77 16	84 81 11
34	72 51 44	72 53 43	73 54 41	73 56 39	73 58 38	74 59 36	74 61 34	75 63 32	76 65 30	76 67 27	77 69 25	78 71 22	80 74 19	82 77 16	84 81 11
35	72 52 44	72 53 42	72 55 41	73 57 39	73 58 37	73 60 35	74 62 33	75 63 31	75 65 29	76 67 27	77 69 25	78 72 22	80 74 19	81 77 16	84 81 11
36	71 53 43	72 54 42	72 56 40	72 57 38	73 59 37	73 61 35	73 62 33	74 64 31	75 66 29	76 68 27	77 70 25	78 72 22	79 74 19	81 77 15	84 81 11
37	71 53 43	71 55 41	71 56 40	72 58 38	72 59 36	73 61 35	73 63 33	74 64 31	75 66 29	75 68 27	76 70 24	78 72 22	79 75 19	81 78 15	83 81 11
38	70 54 42	71 56 41	71 57 39	71 59 38	72 60 36	72 62 34	73 63 33	73 65 31	74 67 29	75 68 27	76 70 24	77 73 22	79 75 19	81 78 15	83 81 11
39	70 55 42	70 56 41	70 58 39	71 59 37	71 61 36	72 62 34	72 64 32	73 65 31	74 67 29	75 69 26	76 71 24	77 73 22	79 75 19	81 78 15	83 82 11
40	69 56 42	70 57 40	70 58 39	70 60 37	71 61 36	71 63 34	72 64 32	73 66 30	74 68 28	75 69 26	76 71 24	77 73 21	79 76 19	81 78 15	83 82 11
41	69 56 41	69 58 40	70 59 39	70 60 37	70 62 35	71 63 34	72 65 32	72 66 30	73 68 28	74 70 26	75 72 24	77 74 21	78 76 19	80 78 15	83 82 11
42	68 57 41	69 58 40	69 60 38	70 61 37	70 62 35	71 64 34	71 65 32	72 67 30	73 68 28	74 70 26	75 72 24	77 74 21	78 76 19	80 79 15	83 82 11
43	68 58 41	68 59 40	69 60 38	69 62 37	70 63 35	70 64 34	71 66 32	72 67 30	73 69 28	74 70 26	75 72 24	76 74 21	78 77 18	80 79 15	83 82 11
44	67 58 41	68 59 39	68 61 38	69 62 36	69 63 35	70 65 33	71 66 32	71 68 30	72 69 28	73 71 26	75 73 24	76 74 21	78 77 18	80 79 15	83 82 11
45	67 59 41	67 60 39	68 61 38	68 63 36	69 64 35	69 65 33	70 67 32	71 68 30	72 70 28	73 71 26	74 73 24	76 75 21	77 77 18	80 79 15	83 82 11
46	66 59 40	67 61 39	67 62 38	68 63 36	68 64 35	69 66 33	70 67 31	71 68 30	72 70 28	73 71 26	74 73 24	75 75 21	77 77 18	79 79 15	
47	66 60 40	66 61 39	67 62 37	67 64 36	68 65 35	69 66 33	69 67 31	70 69 30	71 70 28	72 72 26	74 73 24	75 75 21			
48	65 61 40	66 62 39	66 63 37	67 64 36	68 65 35	68 67 33	69 68 31	70 69 30	71 71 28	72 72 26					
49	65 61 40	65 62 39	66 63 37	66 65 36	67 66 35	68 67 33	69 68 31	70 70 30							
50	64 62 40	65 63 39	65 64 37	66 65 36	67 66 34	67 67 33									
51	64 62 40	64 63 39	65 64 37	65 65 36											
52	63 63 40	64 64 39													
53															
54															
55															
56															
57															
58															
59															
60															

The other co-ordinate of each pole (i.e. the angle between the appropriate etch-pit face and the sectioning plane) is given for triangular pits by the relations (5). Calculation of the angles α, β, and γ for each pit has been obviated by the construction of a table, in which the values of α, β, and γ are given for all values of angles A and C at one-degree intervals. Table I gives that part of the table corresponding to the restriction that A is the smallest etch-pit angle and C the largest; α, β, and γ are given in that order. The range of values covered by Table I is sufficient for practical application of the method, since it is obvious that the nomenclature of the etch-pit angles is purely arbitrary. The use of Table I is simplified considerably by the fact that the angles α, β, and γ are in reverse order of magnitude from angles A, B, and C.

This will be illustrated by a numerical example. Suppose that the measured angles, A, B, and C are 73°, 39°, and 68°, respectively. If the angles had been in the order 39°, 68°, 73°, direct reference to Table I would have given $\alpha = 69°$, $\beta = 52°$, $\gamma = 45°$. To obtain α, β, and γ for the actual measurements $A = 73°$, $B = 39°$, $C = 68°$, the values obtained from the table must be re-arranged in the same way as the set 39°, 68°, and 73° must be re-arranged to give the set 73°, 39°, and 68°. This gives $\alpha = 45°$, $\beta = 69°$, and $\gamma = 52°$. It can be seen that α, β, and γ are then in reverse order of magnitude from A, B, and C, the direct use of this property being a quicker way of obtaining the above result. The poles are then plotted at angles α, β, and γ along the appropriate radii.

For rectangular pits the angles between the etch-pit faces and the sectioning plane are 90° for two faces, $\tan^{-1}\frac{l}{k}$ for the third face, and $\tan^{-1}\frac{k}{l}$ for the fourth face. Table II gives values of $\tan^{-1}\frac{l}{k}$ for various values of l^2/k^2; it will be noted that $\tan^{-1}\frac{k}{l} = 90° - \tan^{-1}\frac{l}{k}$.

Table II.—*Angle Between* (0*kl*) *Plane and* (010) *Plane for Various Values of* l^2/k^2.

l^2/k^2	$\tan^{-1}\frac{l}{k}$	l^2/k^2	$\tan^{-1}\frac{l}{k}$
1/1	45°	11/2	66° 54′
5/4	48° 28′	6/1	67° 48′
4/3	49° 6′	7/1	69° 18′
3/2	50° 46′	8/1	70° 32′
2/1	54° 44′	9/1	71° 34′
5/2	57° 41′	10/1	72° 27′
3/1	60°	15/1	75° 31′
7/2	61° 53′	20/1	77° 24′
4/1	63° 26′	25/1	78° 42′
9/2	64° 46′	30/1	79° 39′
5/1	65° 54′	∞/1	90°

These angles are again plotted along the appropriate radii.

In the case of square pits the angles between the etch-pit faces and the sectioning plane are 90° for four of the faces and 0° for the fifth. Thus the poles are plotted at the centre and at 90° along the four radii.

VI.—CONSTRUCTION OF POLE FIGURES FOR POLYCRYSTALLINE AGGREGATES

In Section IV it was shown that data obtained from an etch-pit enabled the orientation of the grain in which it was formed to be plotted on a stereographic projection. A simple extension of this procedure enables pole figures to be plotted for polycrystalline aggregates, provided that the reference direction is maintained constant throughout all the readings. In applying the method for this purpose the problem is essentially one of sampling.

If a large number of etch-pits distributed over the sectioning plane is measured and the {100} poles for each one plotted on a stereographic projection, the resulting array of points is a true pole figure and may be simplified in the conventional manner by drawing contour lines to enclose areas of equal pole density. It is merely necessary to estimate how many etch-pits must be measured and how they should be distributed in order to obtain a reliable pole figure. With regard to distribution, the most satisfactory way of distributing the points at which readings are taken is probably a triangular array such that each point has six equidistant nearest neighbours at the corners of a regular hexagon. However, a square array is probably almost as good, and when the practical details of ease of location of the specimen are taken into consideration the square array has much to commend it.

It is obviously desirable that the array shall cover as much of the area of the sectioning plane as possible, and so, for any particular specimen, the total number of readings to be taken depends on the shortest distance between two points in the array. It can be seen that an almost perfect sample should be obtained when this distance is less than the average apparent grain diameter (on the sectioning plane), but for small grain-sizes this would require a very large number of readings.

The number of pits which must be measured is determined in practice so as to achieve an economic balance between the desirability of obtaining at least one measurement on each grain of the specimen and the time and labour spent in obtaining the measurements and plotting the poles. It is obvious that the fewer measurements are required the more highly preferred the texture of the sample.

Care must be taken in drawing contour lines on the pole figure, particularly when the orientation of the sample is almost random. This arises from geometrical considerations of the stereographic projection, since equal areas on the ideal projection sphere do not plot as equal areas on the pole figures, unless they are geometrically similar and lie at the same latitude. The effect of this inequality is to make the pole density at the centre of the pole figure apparently four times as great as that at the circumference for a perfectly random orientation. A convenient means of overcoming this difficulty is by plotting the poles initially on a specially prepared chart instead of a stereographic projection, the scales of the co-ordinates being such that equal areas on the chart correspond to equal areas on the projection sphere. Contour lines may then be drawn on the chart and subsequently transferred to a normal pole figure without replotting individual poles. The chart may conveniently be in the form of a rectangle with a linear scale (from 0° to 360°) in one direction. (This scale corresponds to the radius co-ordinate of the pole.) The other axis may then be divided into n equal ranges and the values of ω (the co-ordinate of the pole representing the angle between itself and the centre of the pole figure) at the origin, and at the end point of each of the n ranges taken to be :

$$0, \cos^{-1}\left(1-\frac{1}{n}\right), \quad \cos^{-1}\left(1-\frac{2}{n}\right), \; . \; . \; . \; . \; \cos^{-1}\left(1-\frac{n-1}{n}\right), 90^\circ$$

It can be proved that the areas formed by intersection of perpendiculars at the ends of these ranges and a set of equally spaced lines parallel to the axis of ω will plot as equal areas on the projection sphere.

It has been found convenient to take n as 10, and in this case the angles, 0°, 25° 50′, 36° 52′, 45° 34′, 53° 8′, 60°, 66° 25′, 72° 32′, 78° 28′, 84° 16′, and 90° are marked at equal intervals along the ω axis, the radius co-ordinate axis being divided into 36 equal intervals of 10° each. Thus the pole figure is effectively divided into 360 areas which appear as equal areas on the chart and correspond to equal areas on the projection sphere. Each pole may then be plotted on the chart using its two co-ordinates at their face values. Pole densities on the chart are proportional to true pole densities on the projection sphere. Contour lines drawn on the chart may then be transferred directly to the normal pole figure using a polar net.

It will be noted that since this method determines pole density directly, rather than by indirect measurements of physical quantities such as darkening of a photographic emulsion and its numerous corrections and uncontrollable variations, it is possible to give reliable quantitative estimates of the proportions of any actual orientations present in any specimen, within the limits of sampling errors.

VII.—EXAMPLES OF USE OF THE METHOD

Two examples of the use of the method described in this paper are given.

The first of these is taken from Fig. 10 (Plate III) of Lacombe and Beaujard's paper,[2] in which etch-pits and slip lines are visible on a single grain. Angular measurements were made on this illustration using the vertical as reference direction and the method described in the present paper applied to plot the {100} poles, these being shown in Fig. 6. The {111} poles were then obtained from the {100} poles by geometrical construction, and the observed positions of the normals to the slip lines compared with the calculated positions of the {111} poles. (There is, of course, no means of determining from the photograph at what angle the slip planes are situated to the sectioning plane.) It can be seen from Fig. 6 that the agreement is good, particularly when the difficulty

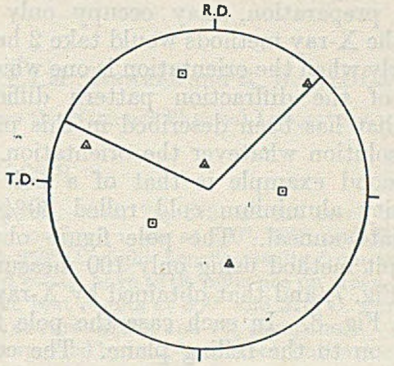


Fig. 6.—Stereographic Projection of Grain Showing Etch-Pits and Slip Lines. (Data taken from Fig. 10 (Plate III) of Lacombe and Beaujard's paper.[2]) Normals to slip lines shown as radii.

⊡ {100} poles plotted from measurement of etch-pits.

△ {111} poles obtained by geometrical construction from {100} poles.

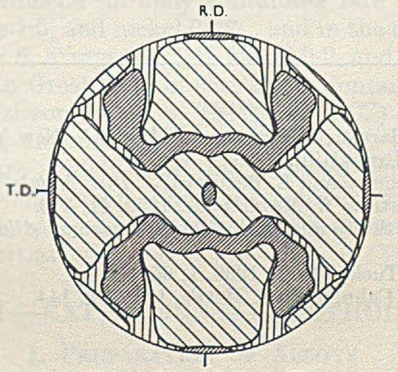


Fig. 7.—(100) Pole Figure Plotted from Measurements of Etch-Pits on Super-Purity Aluminium Cold Rolled 60% After Annealing.

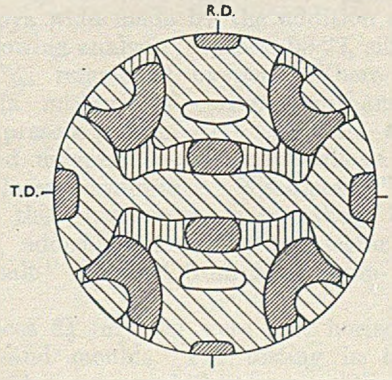


Fig. 8.—As Fig. 7, but Plotted from X-Ray Diffraction Photographs.

of making accurate measurements on a photograph is taken into account.

The plotting of the orientation of a single crystal in this way offers considerable advantages over any X-ray technique, since the whole process, including specimen preparation, may occupy only 15 min., whereas the X-ray methods would take 2 hr. or more, particularly when the orientation is one which renders solution of the diffraction pattern difficult. The method that has been described in this paper gives a direct solution whatever the orientation.

The second example is that of a pole figure of super-purity aluminium cold rolled 60% after an intermediate anneal. The pole figure obtained by the etch-pit method using only 100 measurements is given in Fig. 7, and that obtained by X-ray means is shown in Fig. 8. In each case the pole figure was projected on to the rolling plane. The contours in each case were fixed independently by dividing the total pole density range, from zero to the maximum observed, into four approximately equally spaced ranges. The agreement is good and, furthermore, the etch-pit pole figure was produced in approximately half the time required for producing the X-ray pole figure.

VIII.—CONCLUSIONS

A method of determining orientation of individual crystals from etch-pits has been developed which is suitable for plotting pole figures for polycrystalline aggregates and does not require any apparatus additional to that normally found in a metallurgical laboratory. It yields results that agree well with X-ray means of determining orientations, and is usually faster than X-ray methods, particularly for single crystals or samples possessing highly preferred orientations. The method is particularly useful for determining the pole figures of materials whose grain-sizes are too large for X-ray-scanning methods, and it may be used to give quantitative estimates of proportions of various orientations present in any sample.

ACKNOWLEDGEMENTS

The authors wish to thank Aluminium Laboratories Limited for permission to publish this paper, and their colleagues for helpful discussions and assistance with the development and practical application of the method.

REFERENCES

1. C. S. Barrett, "Structure of Metals", New York: **1943** (McGraw-Hill).
2. P. Lacombe and L. Beaujard, *J. Inst. Metals*, 1948, **74**, 1.
3. H. Kostron, *Z. Metallkunde*, 1950, **41**, 370.
4. D. W. Smith and R. F. Mehl, *Metals and Alloys*, 1933, **4**, 31.
5. C. M. Tucker, *ibid.*, 1930, **1**, 655.
6. R. W. Cahn, *J. Inst. Metals*, 1951, **79**, 144.
7. H. Mahl and I. N. Stranski, *Z. physikal. Chem.*, 1942, [B], **51**, 319.
8. R. Jacquesson and J. Manenc, *Compt. rend.*, 1950, **230**, 959.
9. H. Mahl and I. N. Stranski, *Z. physikal. Chem.*, 1942, [B], **52**, 257.
10. A. Hone and E. C. Pearson, *Metal Progress*, 1948, **53**, 363.

CREEP AT 250° AND 300° C. OF SOME MAGNESIUM ALLOYS CONTAINING CERIUM [*]

1438

By G. A. MELLOR,[†] M.Sc., F.I.M., MEMBER and R. W. RIDLEY,[‡] B.Sc.

(Communication from the National Physical Laboratory.)

SYNOPSIS

Following on an investigation (*J. Inst. Metals*, 1949, 75, 679) of the creep strength of magnesium–cerium alloys at 200° C., tests have now been made at 250°, 300°, and 316° C., with and without addition of a third element. At 250° C. slight benefit results from additions of silicon and zirconium; silver, lithium, zinc, and cobalt have little influence on creep behaviour; and cadmium and aluminium are harmful. The best results at 300° and 316° C. were obtained with as-cast alloys containing about 2% cerium or rare earths, and 1% manganese. Both rolled and cast alloys of low manganese content can be improved by heat-treatment, but cast magnesium–cerium–manganese alloys remain the most creep-resistant.

Examination with the optical and electron microscopes showed improvement in creep-resistance to be due mainly to the presence of fine precipitates, manganese apparently restraining the coarsening of the Mg_9Ce particles.

I.—INTRODUCTION

In previous work [1] on the creep strength of magnesium–cerium alloys, it had been established that creep-resistance both in cast and in rolled bars at 200° C. was due principally to the cerium taken into solution and precipitated in a fine form at the temperature of test. Solution-treatment and ageing at 350° C. resulted in a coarser precipitate and in poor creep behaviour at 200° C. It was apparent therefore that industrial requirements for alloys maintaining their strength at 250°–300° C. were unlikely to be met by binary magnesium–cerium alloys. It was accordingly decided to explore the creep behaviour at temperatures higher than 200° C. and to see what improvement could be obtained by the addition of a third element.

The test period was extended to 1000 hr., or to 1% strain, in order to establish the long-time behaviour of the alloys. In one or two cases of particular interest tests were continued for 3000–4000 hr., and improvements due to metallurgical changes taking place during test became clearly apparent.

Reference to other published work on the effect of additions to magnesium–cerium alloys did not reveal any outstanding improvement in creep-resistance. Leontis [2] tested at 149° and 204° C. some 50 magnesium alloys containing zinc, together with various third elements, but no materials better in creep than binary magnesium–cerium alloys were discovered. The work of Leontis and Murphy,[3] of Grube and Eastwood,[4] and of Grube, Davis, and Eastwood [5] in America has shown the effectiveness of manganese additions to magnesium–cerium alloys. For high creep-resistance in the wrought state Grube and Eastwood recommend an alloy containing rare earths 2, manganese 1·5, and nickel 0·2%, and in the cast state rare earths 6, manganese 0·8, nickel 0·2, and tungsten 0·02%. In Great Britain cast alloys containing rare earths 3, zirconium 0·6% (Elektron MCZ) and rare earths 2·5, zirconium 0·6, and zinc 2·5% (Elektron ZRE 1) are widely used for elevated-temperature service.[6, 7] Zirconium and zinc contribute to grain refinement, ease of casting, and good room-temperature properties.

II.—EXPERIMENTAL METHOD

1. PREPARATION OF ALLOYS

The preparation of the alloys and metallographic examination were carried out in the Metallurgy Division of the National Physical Laboratory. Most of the alloys were made by the addition of cerium to the following analysis: cerium 98·22, silicon 0·04, iron 0·77%, remainder oxygen. Where "Mischmetall" (a mixture of cerium, lanthanum, neodymium, praseodymium, &c.) was used, the alloys are termed magnesium–rare earth alloys. As will be shown later, cerium and "Mischmetall" are very similar in their behaviour. Some alloys contained zirconium, added by means of a master alloy or "master salt" obtained from Magnesium Elektron, Ltd.

Test-pieces 8½ in. long were cast horizontally in pairs in sand moulds. For testing in the rolled condition 2¼-in.-square billets were chill cast and preheated to 500° C. for rolling to $\frac{15}{16}$ in. dia. Alloys containing lithium were rolled from 450° C.

---

[*] Manuscript received 23 May 1952.

[†] Metallurgy Division, National Physical Laboratory, Teddington, Middlesex.

[‡] High-Temperature Mechanical Properties Section, National Physical Laboratory, Teddington, Middlesex.

2. Creep-Test Procedure

The creep tests were carried out in the High-Temperature Mechanical Properties Section of the National Physical Laboratory.

The majority of the tests were made in a 5-ton-capacity creep machine, using specimens with a gauge-length of 5 in. and a dia. of 0·564 in. The creep strain was measured by means of a modified Martens-type mirror extensometer capable of measuring strain of the order of 1×10^{-6}. For certain tests on the rolled alloys 2-ton-capacity creep machines were used, with a smaller specimen of 2 in. gauge-length and 0·357 in. dia. These machines were equipped with a mirror-type extensometer sensitive to 1×10^{-5} strain.

The period of soaking at the test temperature before the load was applied was approximately 16 hr., the test temperature being reached in about 5 hr. The temperature of the furnace was electronically controlled and did not depart at any part of the gauge-length by more than ±1° C. from the required temperature throughout the test.

III.—EXPERIMENTAL RESULTS

1. Effect of Additions to Magnesium–Cerium Alloys

In choosing the third elements to be added the desirability was considered of:

(*a*) Increasing the amount of cerium in solution at the solution-treatment temperature.

(*b*) Preventing the formation of a coarse precipitate of Mg_9Ce at 300° C.

(*c*) Producing alloys with a high solidus temperature.

Additions of manganese, aluminium, zirconium, and zinc were selected, because these elements were already employed in commercial alloys. Silver, cadmium, or lithium forms an extensive solid solution in magnesium, and might therefore modify the behaviour of dissolved cerium. Cobalt had given promising results in some previous work by Haughton,[8] and silicon was added in order to produce fine particles in the structure.

Table I.—*Solidus Temperatures of Ternary Alloys.*

Alloy No.	Composition: Ce, %	Composition: Addition Element, %	No Indication of Liquid *	Indication of Liquid *
*X*67	3·8	Nil	585° C.	590° C.
*X*88	0·81	0·38 Si	620° C.	...
*X*93	1·05	0·16 ,,	,,	...
*X*80	1·4	0·19 ,,	,,	...
*X*99	2·74	0·11 ,,	580° C.	600° C.
*X*78	0·58	1·52 Mn	610° C.	620° C.
*X*71	1·82	1·28 ,,	580° C.	600° C.
*X*70	4·35	1·19 ,,	585° C.	590° C.
*X*83	1·86	0·60 Zr	600° C.	610° C.
*X*115	2·05	0·54 ,,	590° C.	603° C.
*X*85	2·12	1·51 Ag	580° C.	600° C.
*X*81	2·01	1·47 Li	550° C.	580° C.
*X*84	2·09	1·68 Al	620° C.	...
*X*82	1·91	2·15 Cd	580° C.	600° C.

* Specimens water-quenched after 15 min. at temperature.

The tests at 200° C. had shown that higher creep-resistance was often obtained by increasing the

Table II—*Effect of Additions to Rolled and Solution-Treated Alloys Tested at 250° C. and 1 Ton/In.^{2}.*

Alloy No.	Composition		Solution-Treatment Temp., °C.	At 120 Hr.		At 300 Hr.		At 1000 Hr.		Time to Reach 1% Strain, hr.
	Ce, %	Addition Element, %		Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	
*X*61*D*1	1·45	Nil	580	0·0245	0·86	0·037	0·65	...	...	...
*X*80*B*2	1·50	0·13 Si	600	0·020	0·35	0·023	0·18	...	...	...
*X*80*B*3			600	0·0153	0·825	0·0274	0·55	0·0616	0·481	...
*X*72*C*7	1·71	0·25 Mn	580	0·0098	0·414	...	...	...	...	...
*X*71*B*1	1·82	1·28 ,,	580	0·0280	0·48	0·0375	0·296	0·048	0·083	...
*X*83*B*1	1·86	0·60 Zr	550	0·0260	1·05	...	...	...	...	...
*X*83*B*2			550	0·0276	0·72	0·0367	0·336	...	...	...
*X*83*B*4			600	0·0162	0·744	0·0257	0·406	0·039	0·115	...
*X*85*B*1	2·12	1·51 Ag	580	0·017	0·55	0·027	0·595	...	...	...
*X*85*B*4			550	0·0202	1·04	0·0371	0·89	...	...	...
*X*81*B*1	2·01	1·47 Li	550	0·022	1·00	0·041	1·2	...	...	...
*X*81*B*2			550	0·0234	1·235	...	...	...	...	...
*X*84*B*1	2·09	1·68 Al	550	1·2	117	...	...	...	...	100
*X*82*B*1	1·91	2·15 Cd	550	0·0268	2·3 *	...	...	...	...	...

* Minimum rate at 50 hr.

amount of cerium in solution as a result of heat-treatment at a higher temperature. Specimens of a number of alloys were therefore heat-treated at temperatures from 550° to 620° C., with subsequent microscopical examination to establish the highest possible temperature which could be used for solution-treatment without formation of liquid. The results are shown in Table I. Silicon seems to raise the sloping solidus of the magnesium–cerium diagram but not the eutectic temperature of 590° C. Zirconium appears to raise slightly the temperature of the magnesium–Mg_9Ce eutectic. Manganese additions up to 1·5% make little difference to the magnesium–cerium solidus. Leontis and Murphy[3] have stated that manganese increases the solid solubility of cerium in magnesium, but this point was not investigated in the present work.

TABLE III.—*Effect of Additions to Rolled and Solution-Treated Alloys Tested at 300° C. and 1 Ton/In.*^{2}.

Alloy No.	Composition			Solution-Treatment Temp., °C.	At 120 Hr.		At 300 Hr.		At 1000 Hr.		Time to Reach 1% Strain, hr.
	Ce, %	R.E.,* %	Addition Element, %		Strain, %	Creep Rate, Strain/hr. $\times 10^4$	Strain, %	Creep Rate, Strain/hr. $\times 10^4$	Strain, %	Creep Rate, Strain/hr. $\times 10^4$	
*X*61*D*3	1·45	...	Nil	580	0·20	15·0	0·66	33·2	...	...	...
*X*83*B*3	1·86	...	0·6 Zr	600	0·102	6·9	0·24	8·5	0·89	10·6	1100
*X*83*C*1				600	0·083	5·52	0·185	5·52	0·60	5·52	>1000
*X*74*B*1	...	1·76	0·46 Mn	580	0·067	4·57	0·182	9·44	...	...	...
*X*101*C*1	2·7	...	1·4 ,,	580	0·113	5·3	0·268	14·3	2·02	27·0	620
*X*80*B*5	1·5	...	0·13 Si	610	0·70	158	...	...	...	...	136
*X*80*C*1				600	0·38	60·6	...	...	...	...	181
*X*85*B*2	2·12	...	1·51 Ag	580	0·31	31·5	...	...	...	...	270
*X*100*B*1	1·83	...	1·66 Li	580	Fractured in 73 hr.			...	...	...	28
*X*100*B*2				580	,, 17½ ,,			...	...	...	...

* R.E. = rare earths.

TABLE IV.—*Effect of Additions to Cast Alloys Tested at 300° C. and 1 Ton/In.*^{2}.

Alloy No.	Composition			At 120 Hr.		At 300 Hr.		At 1000 Hr.		Time to Reach 1% Strain, hr.
	Ce, %	R.E., %	Addition Element, %	Strain, %	Creep Rate, Strain/hr. $\times 10^4$	Strain, %	Creep Rate, Strain/hr. $\times 10^4$	Strain, %	Creep Rate, Strain/hr. $\times 10^4$	
*X*33*A*1	5·52	...	Nil	0·44	34	...	...	...	...	230
*X*109*A*1	...	3·07	0·43 Zr	1·39	128·5	...	...	...	...	90
*X*115*A*2	2·05	...	0·54 ,,	...	...	...	...	...	...	53
*X*102*A*1	2·28	...	0·6 Mn	0·055	3·05	0·11	3·07	0·48	9·7	1400
*X*101*A*1	2·7	...	1·4 ,,	0·0426	2·60	0·071	1·57	0·204	2·33	>1000
*X*101*B*1				0·0450	1·76	0·078	1·78	0·242	2·16	>1000
*X*12*A*1	...	3·19	2·86 Zn	0·58	75·7	...	...	...	...	170
*X*104*A*1	3·0	...	1·57 Co	0·76	75·0	...	...	...	...	153
*X*99*D*1	2·7	...	0·11 Si	1·13	171	...	...	...	...	110
*X*111*A*1	...	2·93	2·83 Zn 0·48 Zr	0·33	25	0·78	25	...	...	380
*X*113*A*1	...	3·37	2·89 Zn 0·55 Mn	0·082	3·6	0·134	2·5	0·298	2·55	>1000

The compositions of the alloys tested in the rolled, solution-treated, and water-quenched condition, and the results of creep tests, under a stress of 1 ton/in.^{2}, at 250° and 300° C. respectively, are given in Tables II and III. The solution-treatment temperature was in many cases close to the maximum, but in other cases a rather lower temperature of 550° C. was used. Table IV contains the results of creep tests at 1 ton/in.^{2} and 300° C. on a similar range of as-cast alloys. Creep curves showing some 250° C. tests extended to 1000 hr. are reproduced in Fig. 1.

Whereas silicon, zirconium, and manganese proved beneficial to rolled and solution-treated alloys at

250° C., silver and lithium made little difference to creep behaviour, and cadmium and aluminium were harmful. It is interesting to note that Mäder and Laves [9] found that Al_2Ce is formed in preference to Mg_9Ce when aluminium is added to magnesium–cerium alloys, so that no cerium would remain in solution and the creep-resistance would be diminished. The increased resistance to creep at 250° C. resulting from the addition of silicon (see Table II) may be due to the higher solution temperature that can be used and not to the presence of silicon particles. Silicon was found to have no useful effect in the absence of cerium.

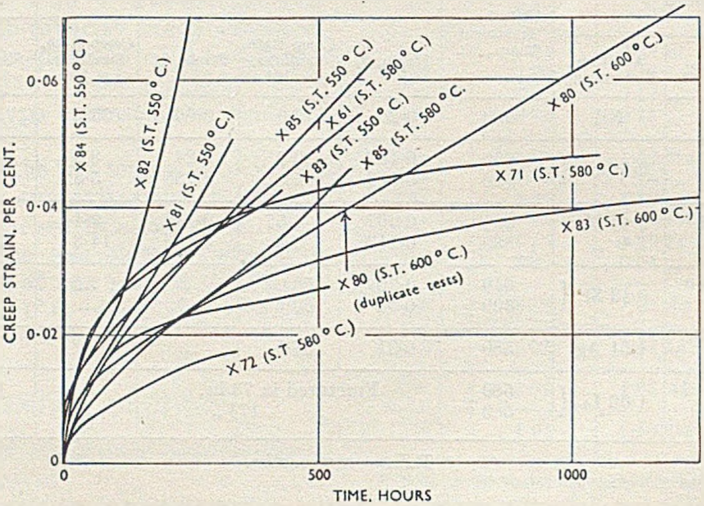


Fig. 1.—Creep Tests on Magnesium–Cerium Alloys at 250° C. and 1 Ton/In.², Showing Effect of Additions. All alloys in the rolled and solution-treated condition. For compositions see Table II.

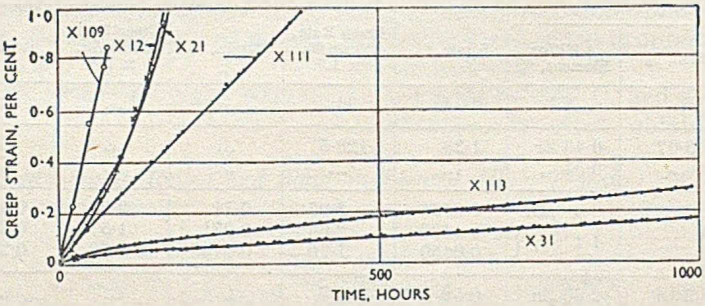


Fig. 2.—Comparison of Creep Properties of Cast Alloys Containing Rare Earths, Zinc, Manganese, and Zirconium, at 300° C. and 1 Ton/In.².

Key.

X12	3·19% rare earths, 2·86% Zn.
X21	2·67% rare earths.
X31	4·83% rare earths, 1·22% Mn.
X109	3·07% rare earths, 0·43% Zr.
X111	2·93% rare earths, 2·83% Zn, 0·48% Zr.
X113	3·37% rare earths, 0·55% Mn, 2·89% Zn.

In view of the poor results obtained at 250° C. with alloys containing cadmium or aluminium, they were not tested at 300° C. In the rolled and solution-treated state at 300° C. additions of manganese or zirconium were found to be advantageous, but lithium, silver, and silicon were harmful to the creep properties.

With alloys in the cast condition, manganese gave greater creep-resistance than zirconium. The use of zinc in the commercial cast alloy ZRE 1 (rare earths 2·5, zirconium 0·6, zinc 2·5%) suggested tests on alloys containing zinc, and Fig. 2 gives creep curves of cast alloys with additions of zinc plus zirconium, of manganese, and of zinc plus manganese. These all showed improvement on the binary magnesium–cerium alloys, but the general trend of the curves emphasizes the importance of manganese. Zinc or zirconium alone brought about little improvement, nor did cobalt or silicon (Table IV).

Further experiments were therefore confined to the effect of manganese on alloys containing cerium or rare earths.

2. Creep Behaviour of Magnesium–Cerium and Magnesium–Cerium–Manganese Alloys

(a) *Tests at 250° and 300° C.*

The results of a few tests of binary magnesium–cerium alloys at 250° C., with a stress of 1 ton/in.²,

Table V.—*Magnesium–Cerium Alloys Tested at 250° C. and 1 Ton/In.^{2}.*

Alloy No.	Ce, %	Treatment	At 120 Hr.		At 300 Hr.	
			Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$
*X*61*D*1	1·45	Rolled, 24 hr. at 580° C.; water-quenched.	0·0245	0·86	0·037	0·65
*X*69*E*5	2·55	Rolled, 24 hr. at 550° C.; water-quenched.	0·0232	1·35	0·046	1·375 *
*X*46*A*1	3·5	As cast.	0·0345	1·67	...	...
*X*46*C*1			0·082	3·85	0·136	2·71
*X*67*B*4	3·8	Rolled, 24 hr. at 550° C.; water-quenched.	0·029	1·50 *	...	...
*X*67*C*4	3·8	Rolled, 24 hr. at 580° C.; water-quenched.	0·0266	0·845	0·0368	0·39
*X*67*C*3			0·0184	0·56	...	...
*X*59*B*3	4·1	As cast.	0·0577	2·67	0·113	2·43
*X*59*B*5			0·08	3·62	...	...
*X*55*B*3	6·0	„	0·059	1·20	...	...

* Minimum rate.

are set out in Table V. They indicate: (i) that cast material is somewhat inferior to that rolled and solution-treated, (ii) that a higher solution-treatment temperature confers some small benefit, (iii) that cerium in excess of about 1·6% gives no further increase in strength to the rolled and solution-treated alloys, and (iv) that additions of cerium beyond 3·5% are of no advantage in cast material. The behaviour is thus very similar to that at 200° C. with a stress of 2 tons/in.^{2}.

Testing at 250° C. was not carried any further, as it was considered of greater practical importance to investigate the behaviour of magnesium–cerium alloys at 300° C. A number of tests were therefore made at that temperature, at a stress of 1 ton/in.^{2}, on alloys of varying cerium or rare-earth contents and containing additions of manganese from 0 to 1·54%.

Whereas at 200° C. such additions seemed of little value, at 300° C. the ternary magnesium–cerium–manganese alloys proved markedly superior to corresponding binary magnesium–cerium alloys. From the results given in Table VI it is possible to compare cerium with the rare earths and to assess the comparative values of cerium and manganese additions. The differences between alloys made with cerium and with rare earths are small, and although the best results were obtained with the latter, it cannot be said that the superiority was significant. The poor results on alloy *X*70 probably represent a failure of this particular material to respond to heat-treatment. In this connection it may be noted that Leontis [10] found little difference between alloys made with cerium and with "Mischmetall" in tests at 316° C.

Table VI.—*Creep Tests at 300° C. and 1 Ton/In.^{2} on Magnesium–Cerium and Magnesium–Rare Earths Alloys, with and without Manganese.*

		Alloy No.	Composition			At 120 Hr.		At 300 Hr.		At 1000 Hr.		Time to Reach 1% Strain, hr.
			Ce, %	R.E., %	Mn, %	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	
Rolled and solution-treated at 580° C.	Without manganese	*X*61*D*2	1·45	...	...	0·26	33·5	...	...	...	...	260
		*X*61*D*3	1·45	...	...	0·20	15	0·66	33·2	...	...	...
		*X*62*B*3	4·33	...	...	...	...	...	...	...	...	48
		*X*63*D*1	5·95	...	...	0·82	112·6	...	...	...	...	135
		*X*73*C*6	...	1·76	...	0·24	23·8	0·96	53·5	...	...	310
		*X*103*C*1	...	5·4	...	0·92	120	...	...	...	...	130
	With manganese	*X*78*A*3	0·58	...	1·52	0·114	5·08	0·242	9·5	1·72	31·5	770
		*X*102*C*1	2·28	...	0·6	0·0735	5·95	0·198	7·6	1·23	10·0	880
		*X*101*C*1	2·7	...	1·4	0·113	5·3	0·268	14·3	2·02	27·0	650
		*X*70*B*1	4·35	...	1·19	1·12	330	...	...	...	...	117
		*X*70*B*2	4·35	...	1·19	0·70	148	...	...	...	...	139
		*X*74*B*1	...	1·76	0·46	0·067	4·57	0·182	9·44	...	...	...
		*X*64*C*5	...	6·14	1·54	0·055	2·7	0·104	3·42	0·376	2·13	>1000
Cast alloys	Without manganese	*X*45*B*2	1·3	...	...	1·56	285	...	...	...	...	100
		*X*46*C*2	3·5	...	...	0·90	85·5	...	...	...	...	134
		*X*33*A*1	5·52	...	...	0·44	34	1·64	116	...	...	230
		*X*21*A*1	...	2·7	...	0·57	60	...	...	...	...	176
		*X*103*A*1	...	5·4	...	0·50	42	1·74	94	...	...	210
	With manganese	*X*102*A*2	2·28	...	0·6	0·041	1·77	...	...	...	...	...
		*X*102*B*1	2·28	...	0·6	0·0326	1·37	0·054	1·41	0·27	6·5	>1000
		*X*101*A*1	2·7	...	1·4	0·0426	2·61	0·071	1·57	0·204	2·33	>1000
		*X*101*B*1	2·7	...	1·4	0·045	1·76	0·078	1·78	0·242	2·16	>1000
		*X*116*A*1	...	1·99	0·95	0·022	0·648	0·0310	0·69	0·065	0·69	>1000
		*X*31*A*2	...	4·8	1·2	0·052	1·68	0·075	1·02	0·14	0·985	>1000
		*X*31*B*1	...	4·8	1·2	0·054	1·46	0·08	1·45	0·178	1·49	>1000
		*X*64*F*	...	6·14	1·54	0·098	2·32	0·16	3·0	0·35	3·55	>1000
		*X*64*G*	...	6·14	1·54	0·099	2·32	0·127	1·10	...	...	...

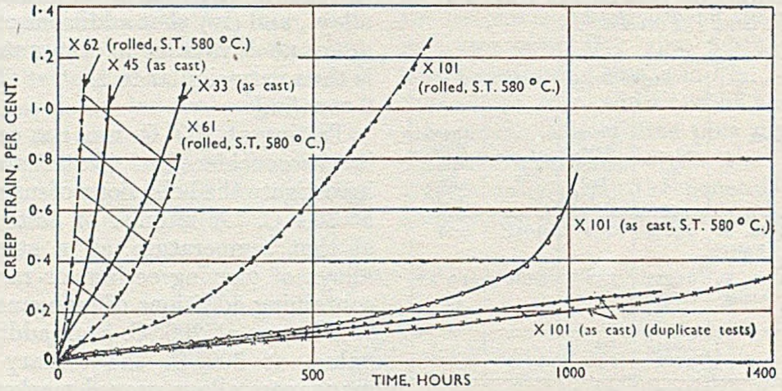


Fig. 3.—Improvement in Creep Properties of Magnesium–Cerium Alloys at 300° C. and 1 Ton/In.², Due to Addition of Manganese. The shaded area covers tests on alloys containing cerium only, in varying amounts. For compositions see Table VI.

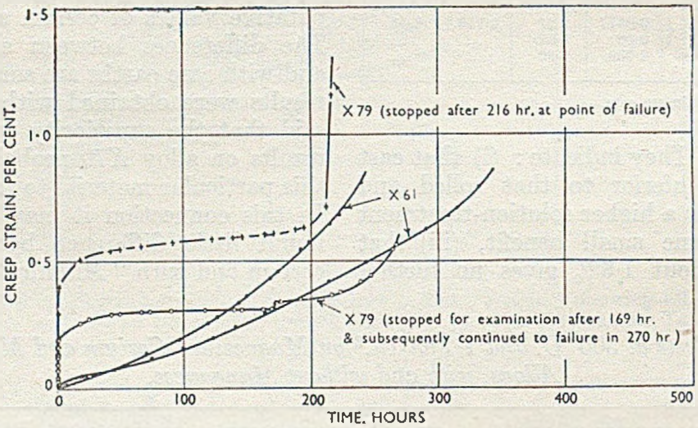


Fig. 4.—Comparison of Creep Properties of Magnesium–Manganese (*X*79) and Magnesium–Cerium (*X*61) Alloys, Rolled and Solution-Treated at 580° C. and Tested at 300° C. and 1 Ton/In.².

Table VII.—*Effect of Heat-Treatment on Cast Alloys as Shown by Creep Tests at 300° C. and 1 Ton/In.².*

Alloy No.	Composition			Treatment	At 120 Hr.		At 300 Hr.		At 1000 Hr.		Time to Reach 1% Strain, hr.
	Ce, %	R.E., %	Mn, %		Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	Strain, %	Creep Rate, Strain/hr. $\times 10^6$	
*X*116*A*1	...	1·99	0·95	As cast.	0·022	0·648	0·031	0·69	0·064	0·69	...
*X*116*B*1				Cast, 24 hr. at 580° C., water-quenched.	0·0352	0·96	0·048	0·615	0·113	1·53	...
*X*116*C*1				Cast, 24 hr. at 580° C., water-quenched, aged 16 hr. at 350° C.	0·0083	0·456	0·0164	0·48	0·063	0·79	...
*X*116*C*2				,, ,, ,,	0·0118	0·665	0·019	0·281	...	...	...
*X*102*B*1	2·28	...	0·6	As cast.	0·0326	1·37	0·054	1·41	0·27	6·5	...
*X*102*B*2				Cast, 24 hr. at 580° C., water-quenched.	0·0142	0·88	0·036	0·99	0·128	2·1	...
*X*101*B*1	2·7	...	1·4	As cast.	0·045	1·76	0·078	1·78	0·242	2·16	...
*X*101*B*2				Cast, 24 hr. at 580° C., water-quenched.	0·051	2·32	0·086	2·2	0·60	66·4	...
*X*31*A*2	...	4·8	1·2	As cast.	0·052	1·68	0·075	1·02	0·14	0·985	...
*X*31*C*2				Cast, 24 hr. at 580° C., water-quenched.	0·042	1·125	0·058	0·82	0·118	0·82	...
*X*33*A*1	5·52	...	...	As cast.	0·44	34	1·64	116	...	...	230
*X*33*A*2				Cast, 24 hr. at 580° C., water-quenched.	0·094	5·4	0·208	7·22	...	...	780

Both at 200° and 250° C. it had been found that no advantage was derived from cerium in excess of 1·6% in rolled and solution-treated binary magnesium–cerium alloys, and in excess of about 3% in cast alloys. There is no indication in Table VI that this conclusion cannot also be applied to the binary alloys tested at 300° C., although the number of results is limited. When manganese is present, less cerium is needed, e.g. the manganese-containing alloy *X*78*A*3 with only 0·58% cerium is stronger than any of the rolled and solution-treated alloys without manganese, and alloy *X*116*A*1 with only 1·99% rare earths is much stronger than any of the cast alloys without manganese addition.

Fig. 3 illustrates the typical improvement due to manganese. Creep curves for alloy *X*101 (cerium 2·7, manganese 1·4%) in the as-cast, cast and solution-treated, and rolled and solution-treated conditions are compared with curves for alloys containing cerium only (shaded area). The improvement shown by the as-cast alloy compared with those rolled and solution-treated, is typical of alloys of similar composition, and the results for duplicate creep tests give a good idea of the reproducibility obtained with cast alloys. The advantage of the longer-time tests is also well illustrated here, the effect of solution-treatment of the cast alloys becoming apparent only after 500 hr.

Grube, Davis, and Eastwood,[5] who tested wrought magnesium–rare earths alloys with various additions, concluded that cerium (or rare earths) was responsible for the tensile properties at 316° C., while manganese was responsible for the creep properties. They found the highest resistance to creep in a 1·8% manganese alloy containing no rare earths, extruded and quenched from 560° C. and tested at 316° C. under a stress of 2000 lb./in.^{2}. The total strain at 120 hr. was 0·09% and at 607 hr. 0·281%. While the importance of manganese is confirmed in the present work, it is not considered as effective in conferring creep-resistance as cerium or rare earths. For example, an alloy containing no cerium and 1·48% manganese, rolled and solution-treated at 580° C., was tested at 300° C. and 1 ton/in.^{2}. Fig. 4 gives the results compared with those of a magnesium–cerium alloy tested under the same conditions. Although the creep rate of the manganese-containing alloy was low at 120 hr., there was early failure with intercrystalline fracture.

A limited number of cast alloys were solution-treated and tested at 300° C., and the results are set out in Table VII. Solution-treatment seems to improve the cast material where there is a high cerium or low manganese content, as in alloys *X*102, *X*33, and *X*31. In the case of alloys *X*101 and *X*116 solution-treatment resulted in a deterioration in creep-resistance which set in at 1000–1100 hr.

The effect of ageing at 350° C. had been found harmful both to cast and rolled binary alloys tested at 200° C. The properties of a rolled and solution-treated binary alloy (*X*61), tested at 300° C., were rendered much worse by ageing at 350° C., whereas a cast and solution-treated alloy containing manganese (*X*116) was slightly improved by the same treatment.

It appears, then, that for the best creep-resistance at 300° C., under a stress of 1 ton/in.^{2}, cast magnesium–cerium–manganese or magnesium–rare earths–manganese alloys are to be preferred. Heat-treatment effects little improvement unless the manganese content is low, and cases exist in which the heat-treatment has resulted in a sharp deterioration.

(b) *Tests at* 316° *C.* (600° *F.*)

Creep tests in other laboratories are often made at 316° C. (600° F.), and it was thought desirable to determine what difference in creep strain could be expected from the rise of 16° C. Three cast alloys, *X*116 (rare earths 1·99, manganese 0·95%), *X*120 (rare earths 4·69, manganese 1·50%), and *X*121 (cerium 4·21, manganese 1·61%) were therefore tested at 1 ton/in.^{2} at both 300° and 316° C. The creep curves reproduced in Fig. 5 show that the strain at 1000 hr. of the best alloy (*X*116) was 0·065% at 300° C., and 0·195% at 316° C.

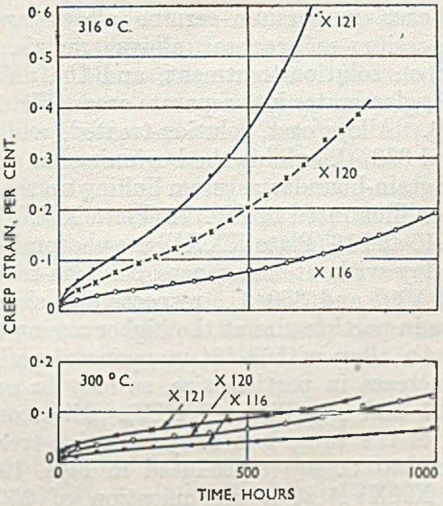


Fig. 5.—Comparison of Creep Properties of Cast Alloys Containing Manganese, at 300° and 316° C. and 1 Ton/In.^{2}.

Similarly, stresses applied in creep tests reported by other workers are often in multiples of 1000 lb./in.^{2} and a direct comparison cannot be made. By extrapolation from design curves of Grube and Eastwood,[4] it is estimated that two experimental alloys of optimum composition stressed at 2240 lb./in.^{2} at 316° C. would extend 0·1% in 100 hr. and 130 hr., respectively. Reference to Fig. 5 shows that the alloys here included reached 0·1% strain in 130, 235, and 660 hr., respectively.

3. Microstructure

Cast magnesium–cerium alloys are coarse-grained, as shown in Fig. 7 (Plate XXXIV). With large amounts of cerium a continuous network of eutectic is found within the larger grain-boundary system (see Fig. 8, Plate XXXIV), and when "Mischmetall" is added instead a eutectic of the divorced type is found, as shown in Fig. 9 (Plate XXXIV). In an alloy containing zirconium (Fig. 10, Plate XXXIV), there is some coring due to the zirconium, and divorce of the eutectic is prevented. There is nothing in the creep results obtained to suggest that the form of the eutectic has any significance.

In the binary magnesium–cerium (or magnesium–rare earths) alloys no precipitate is visible in the magnesium-rich matrix (Fig. 11, Plate XXXV), and the fact that the cast alloys show as good behaviour in creep at 300° C. as the rolled and solution-treated ones suggests that the rate of cooling of the sand-cast bar is sufficiently rapid to retain a considerable amount of cerium in solution. Addition of about 1% of manganese, however, results in a characteristic precipitate in the cast alloy, as shown in Fig. 12 (Plate XXXV).

Sand-cast magnesium–cerium alloys and magnesium–cerium–manganese alloys remain coarse-grained on solution-treatment, and the undissolved Mg_9Ce agglomerates into separate areas (Fig. 13, Plate XXXV). Alloys cast, solution-treated, and aged at 300° and 350° C. exhibit the beginnings of a new and smaller grain-boundary system linking up the areas of Mg_9Ce as illustrated in Fig. 14 (Plate XXXV).

Figs. 15 and 16 (Plate XXXV) are photomicrographs of an alloy without manganese, solution-treated and aged at 300° and 350° C., respectively, showing the increase in particle size at the higher ageing temperature. An alloy with 1·4% manganese exhibits very little increase in particle size, as may be seen from Figs. 17 and 18 (Plate XXXV). Electron micrographs of the same two alloys solution-treated and aged at 350° C. are reproduced in Figs. 19 and 20 (Plate XXXVI) at a magnification of 2500. The finer particles in the manganese-containing alloy appear as white dots in Fig. 20.

4. Relation of Microstructure to Creep Behaviour

In magnesium–cerium and magnesium–cerium–manganese alloys resistance to creep might be due to: (*a*) the eutectic network, (*b*) a strengthening of the matrix by dissolved cerium or manganese, or (*c*) precipitation before or during creep.

As a result of the earlier tests at 200° C. it was concluded that removal of the eutectic network by solution-treatment had little effect on creep behaviour. When testing at 300° C. it was found that high-cerium alloys (see alloys *X*33 and *X*102 in Table VII) were improved by solution-treatment, suggesting that a complete network of eutectic is undesirable. When the network is complete, the alloy may behave like fine-grained material which would be expected to be weaker in creep at the higher testing temperature, or the eutectic itself may be weaker than the matrix. The slight weakening of the manganese-containing alloys *X*101 and *X*116 (Table VII) by solution-treatment is probably connected not with the agglomeration of the eutectic, but with the re-solution of the manganese precipitate. The specimen as cast begins its test with a precipitate already present, as shown in Fig. 12.

Haughton and Schofield[11] reported the solid solubility of cerium in magnesium to be about 1·3% at 580° C., falling to about 0·1% at 350° C. Previous work of the present authors has shown that there is coarse precipitation of Mg_9Ce on ageing at 350° C. with low resistance to creep; while, when a fine precipitate was formed at 200° C., the creep-resistance was high. If creep-resistance were due to the cerium in solution at the temperature of test, it would be expected that: (i) alloys with 0·1–0·2% cerium would be as creep-resistant as those with 1·6%, and (ii) an alloy with 1·6% cerium, solution-treated and aged at 350° C., would be more creep-resistant than one aged at 200° C. In fact, neither is the case, and creep-resistance must be due mainly to a fine precipitation.

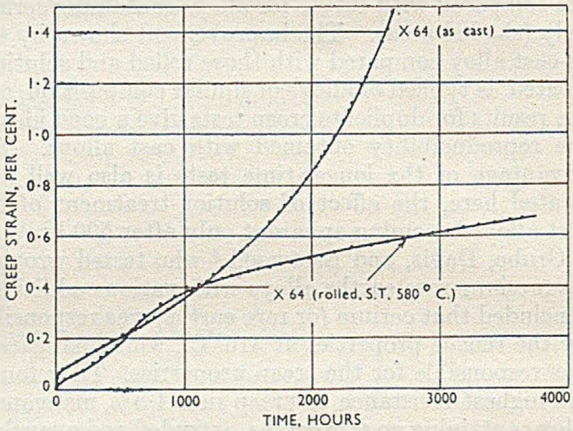


Fig. 6.—Long-Time Creep Behaviour of Alloy *X*64 (Rare Earths 6·14, Manganese 1·54%) in the As-Cast and the Rolled and Solution-Treated Condition, Tested at 300° C. and 1 Ton/In.^{2}

A sloping solid-solubility line is also found in the magnesium–manganese system according to Schmid and Siebel,[12] the solubility being 3·4% manganese at 600° C. and less than 0·1% at 200° C. In solution-treated and aged magnesium–cerium–manganese alloys, both Mg_9Ce and, presumably, α-manganese are precipitated.

The superiority of cast magnesium–cerium–manganese alloys in creep seems to be the result of: (*a*) the presence of a fine precipitate in the cast material, so that the creep specimen is strengthened before test, (*b*) the influence of manganese in restraining the growth of Mg_9Ce particles on ageing (see Figs. 16 and 18), and (*c*) possibly, of a slight increase in the solid solubility of cerium in magnesium when manganese is present. The behaviour of alloy *X*64 (see Table VI and Fig. 6) is of interest in that as rolled and solution-treated it is more resistant to creep than as cast. The structure of the cast alloy contains a large amount of eutectic, and may therefore be expected to be weak in creep, while in the rolled and solution-treated condition the eutectic network has been removed. The improvement, however, is not sufficient to make the alloy as good as other cast alloys. The creep curve (Fig. 6) for the rolled and solution-treated alloy shows a double inflection after about 700 hr., followed by a decreasing creep rate, indicating a delayed precipitation during the test. Its microstructure (Fig. 21, Plate XXXVI) exhibits some grain-boundary precipitation, which is probably responsible for the decrease in the creep rate. When delayed precipitation of this kind takes place, it is important to relate the testing time to the service life of the component. The conclusions reached in Section IV relate to a service life of about 1000 hr., and may have to be modified for materials intended to last appreciably longer.

In the absence of manganese, solution-treatment and ageing at 350° C. of both cast and rolled alloys result in poor creep properties by reason of the coarse precipitate of Mg_9Ce. When sufficient manganese is present, finer particles are precipitated, as shown in Fig. 18, and the alloy resists creep.

The inferiority of rolled and solution-treated alloys in comparison with cast material, when tested at 300° C., may be due to their small grain-size and to the absence of a fine manganese precipitate at the beginning of the test.

IV.—CONCLUSIONS

Various third elements have been added to magnesium–cerium alloys, and their effect on the creep-resistance at 250° and 300° C. examined. Aluminium and cadmium decreased the creep-resistance. Little or no improvement resulted from the addition of silver, lithium, or silicon to wrought alloys, or of cobalt, silicon, or zinc to cast alloys. Zirconium and combined additions of zinc and zirconium had a favourable effect under certain conditions, but the greatest improvement was obtained with manganese. This was marked only when the test temperature was raised to 250°–300° C.; in previous work at 200° C. little effect had been observed.

The good creep-resistance of magnesium–cerium–manganese alloys is well known,[3–5] but the high cerium contents often recommended appear unnecessary. In the present work the best creep-resistance was found in a cast alloy containing 2% rare earths and 1% manganese. The composition range has not been sufficiently explored to enable the limits of cerium and manganese contents for optimum creep-resistance to be defined, but it is thought unlikely that they could be appreciably reduced. The addition of " Mischmetall " has much the same effect as that of cerium.

The effect of heat-treatment varied with different compositions, but as a rule the heat-treatment of cast alloys is not justified. Creep tests of short duration may be misleading, and should be followed up by tests covering the full service life to which the alloys are to be subjected.

Cerium is effective in promoting creep-resistance by reason of a fine precipitate which forms during test at 200°–300° C. Manganese appears to modify the precipitation phenomenon, delaying the coarsening of the Mg_9Ce precipitate and possibly altering the quantity and nature of the precipitated particles. Leontis and Murphy [3] noted that manganese delayed the changes in electrical conductivity on ageing these alloys. Manganese probably enters into the constitution of the Mg_9Ce precipitate in some way, since if manganese and cerium in the alloy were without action upon each other, it is unlikely that the rate of coarsening of the Mg_9Ce precipitate would be affected.

ACKNOWLEDGEMENTS

The authors desire to acknowledge the assistance rendered by Mr. L. J. V. Mackay in making up the alloys and by Miss L. V. Game in carrying out the creep tests.

The work described forms part of the research programme of the National Physical Laboratory, and the paper is published by permission of the Director of the Laboratory.

REFERENCES

1. G. A. Mellor and R. W. Ridley, *J. Inst. Metals*, 1949, **75**, 679.
2. T. E. Leontis, *Trans. Amer. Inst. Min. Met. Eng.*, 1949, **180**, 287.
3. T. E. Leontis and J. P. Murphy, *ibid.*, 1946, **166**, 295.
4. K. Grube and L. W. Eastwood, *Proc. Amer. Soc. Test. Mat.*, 1950, **50**, 989.
5. K. Grube, J. A. Davis, and L. W. Eastwood, *ibid.*, 1950, **50**, 965.
6. A. J. Murphy and R. J. M. Payne, *J. Inst. Metals*, 1947, **73**, 105.
7. R. G. Wilkinson, *Metallurgia*, 1949, **41**, 91.
8. J. L. Haughton, " Magnesium and Its Alloys ". **1940**: London (H.M. Stationery Office).
9. H. Mäder and F. Laves, *Aluminium*, 1943, **25**, 157.
10. T. E. Leontis, *Trans. Amer. Inst. Min. Met. Eng.*, 1949, **185**, 968.
11. J. L. Haughton and T. H. Schofield, *J. Inst. Metals*, 1937, **60**, 339.
12. E. Schmid and G. Siebel, *Metallwirtschaft*, 1931, **10**, 923.

1439 New Values of the Coefficients of Equivalence for Manganese, Iron, Cobalt, and Nickel in Copper–Zinc Alloys *

By J. B. HAWORTH,† D.Phil., B.Met., JUNIOR MEMBER

SYNOPSIS

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 manganese, iron, and nickel are in error, and has enabled a hitherto undetermined coefficient to be stated for cobalt.

THE industrial alloy manufacturer has frequently made use of a system of " coefficients of equivalence " when making complex brasses. These coefficients were determined empirically by Guillet [1] for a number of metals which form solid solutions with copper–zinc alloys, and exert effects on the properties similar to changes in the zinc content. They enable proportions of these metals to be expressed in terms of zinc or copper, so that in complex alloys the nature and amounts of the additions subject to the retention of the α, $\alpha + \beta$, or β structure desired may be ascertained. Work at Oxford on the constitution of ternary alloys of copper–zinc and copper–aluminium with manganese, iron, cobalt, and nickel produced results which proved to be of particular interest with regard to the theory of the transition metals, and it was from the theoretical rather than the practical aspect that the paper was published recently.[2] However, the experimental work which was carried out on the copper–zinc ternary alloys is of practical significance also, as new values for several coefficients of equivalence may be derived from the results.

In the above work the effect on the solubility relationships of the α and β phases in copper–zinc and copper–aluminium alloys of small additions of manganese, iron, cobalt, and nickel was studied at a standard temperature of 672° C. by the method of annealing and microscopical examination. In this way isothermal sections were plotted in the α/β regions for the eight ternary systems. In the copper–zinc series the $\alpha/\alpha + \beta$ and $\alpha + \beta/\beta$ boundaries proved to be straight and parallel lines for small additions of the third metal, but the directions of the boundaries relative to the copper–zinc side of the ternary diagram were found to be different in each system. Furthermore, the angle between this axis and each boundary was found to increase progressively, in congruence with the series Mn–Fe–Co–Ni in the Periodic Table.

It is well known that in alloy systems such as these, where the atomic-size factors are favourable and the electrochemical factors are small, the α/β-brass type of equilibrium is determined mainly by electron concentration. If it is assumed, therefore, that the $\alpha/\alpha + \beta$ and $\alpha + \beta/\beta$ solubility curves in the isothermal sections are lines of constant electron concentration, the slopes of the boundaries determined experimentally show that manganese, iron, cobalt, and nickel act in copper–zinc alloys as though they possessed valencies of 1·83, 1·0, 0·8, and 0·61 respectively. Thus, if one atom of copper (1 valency electron) in a copper–zinc alloy is replaced by one atom of manganese (1·83 valency electrons), there is a net increase of 0·83 electrons. In other words, one atom of manganese added to a copper–zinc alloy produces a change equivalent to the addition of 0·83 atoms of zinc. Similarly, if one atom of zinc (2 valency electrons) is replaced in a copper–zinc alloy by one atom of iron (1 valency electron), one atom of cobalt (0·8 valency electron), or one atom of nickel (0·61 valency electron), there are net changes in the number of electrons in each case of —1·0, —1·2, and —1·39, respectively. We thus arrive at the following values for the coefficients of equivalence :

Mn + 0·83 (i.e. 1 at.-% Mn is equivalent to 0·83 at.-% Zn)
Fe — 1·0 („ „ „ Fe „ „ „ 1·0 at.-% Cu)
Co — 1·2 („ „ „ Co „ „ „ 1·2 at.-% Cu)
Ni — 1·39 („ „ „ Ni „ „ „ 1·39 at.-% Cu)

The values for manganese, iron, and nickel may be compared with Guillet's values of +0·5, +0·9, and —1·2, respectively.[1] These latter values would be in weight per cent., but conversion to atoms per cent. would make only a slight difference in the figures.

REFERENCES

1. See, e.g., H. C. H. Carpenter and J. M. Robertson, " Metals ". Vol. II, p. 1305. Oxford: **1939** (Oxford University Press).
2. J. B. Haworth and W. Hume-Rothery, *Phil. Mag.*, 1952, [vii], 43, (6), 613.

---

* Manuscript received 12 September 1952.

† Late Inorganic Chemistry Laboratory, Oxford; now with Sheepbridge Engineering, Ltd., Chesterfield.

THE EFFECT OF MINOR ADDITIONS ON THE AGE-HARDENING PROPERTIES OF A HIGH-PURITY LEAD–ANTIMONY ALLOY *

1440

By L. M. T. HOPKIN,† B.Sc., A.R.S.M., A.I.M., Member, and
C. J. THWAITES,‡ B.Sc., A.R.S.M., Junior Member

(Communication from the British Non-Ferrous Metals Research Association.)

SYNOPSIS

It was found that little or no age-hardening occurred, after suitable treatments, in a lead–0·85% antimony alloy of extreme purity, although the alloy of commercial purity is known to age-harden extensively. Minor additions of various elements whose solid solubility in lead varied with temperature caused marked age-hardening of the alloy and increased the dispersion of the precipitated antimony. As little as 0·001% arsenic promoted marked hardening. Addition elements either completely or negligibly soluble at all temperatures had no effect. A possible mechanism is suggested which is similar to that offered by Hardy (*J. Inst. Metals*, 1950–51, 78, 169) to explain the effect of minor additions on the age-hardening characteristics of an aluminium–4% copper–0·15% titanium alloy.

I.—INTRODUCTION

The British Non-Ferrous Metals Research Association's current investigation of the creep and fatigue properties of lead and its alloys includes tests on a specially refined lead of the highest purity and on alloys of this material in which the addition element (i) is in solid solution, (ii) is insoluble, and (iii) causes age-hardening. Antimony was selected as the alloying element causing age-hardening, but it was found that a high-purity alloy containing 0·85% antimony would not age-harden in the manner characteristic of the commercial alloy. The present paper shows that the addition of small amounts of other elements to the high-purity alloy causes marked age-hardening.

II.—PREVIOUS WORK

Examination of the literature revealed that Hofmann, Schrader, and Hanemann [1] obtained no age-hardening with a 1% antimony alloy based on a lead of 99·994% purity, whereas the same alloy based on a lead of 99·960% purity age-hardened rapidly under similar conditions. Several papers were found which described how minor additions increase the rate of hardening of lead–antimony alloys. Seljesater [2] showed that the addition of 0·01% arsenic to a 1% antimony alloy increased the hardness, after ageing at room temperature for 28 days, by about 70%. According to Bluth and Hanemann,[3] small additions of both copper and arsenic increased the rate of age-hardening of a 1% antimony alloy. Schumacher, Bouton, and Ferguson [4] found that the resistivity of a 1% antimony alloy decreased more rapidly with time, indicating an increased rate of precipitation, when small additions of a third element were made to the alloy. Additions of arsenic, copper, silver, nickel, and manganese were investigated by these workers, and arsenic was found to be the element with the greatest effect. Considerable effects were also obtained with 0·002% copper or with 0·005% silver. Schumacher [5] observed that the tensile strength of a lead + 1% antimony + 0·005% arsenic alloy increased more rapidly, during ageing, than that of a binary lead–1% antimony alloy. Hofmann and Hanemann [6] found that the resistivity of a 2% antimony alloy decreased more rapidly with time when 0·05% arsenic was present in the alloy.

From the above it is clear that minor additions of various third elements markedly increase the rate of age-hardening and precipitation in lead–antimony alloys. The present paper describes tests on the effects of small additions of arsenic, tin, copper, and silver on the age-hardening properties, after heat-treatment, of a lead–0·85% antimony alloy of high purity.

III.—EXPERIMENTAL DETAILS

It was originally expected that the high-purity binary lead–0·85% antimony alloy would age-harden, and it was extruded into rods suitable for creep and

---

* Manuscript received 28 August 1952. The work described in this paper was made available to members of the B.N.F.M.R.A. in a confidential research report issued in October 1951.

† Formerly Investigator, B.N.F.M.R.A.; now at Metallurgy Division, National Physical Laboratory, Teddington.

‡ Formerly Investigator, B.N.F.M.R.A.; now with the Tin Research Institute, Greenford, Middlesex.

fatigue testing. When it was found that this material did not age-harden, trial experiments were carried out on cast alloys with minor additions of the elements already mentioned.

1. Materials

The analyses of the specially refined Tadanac lead and the high-purity antimony were:

Specially Refined Tadanac Lead

Element	%	Element	%	Element	%
Ag .	0·00003	Tl .	0·0002	Au .	trace
Cu .	0·0001	Sb .	0·0001	In .	„
Zn .	0·0001	Cd .	<0·0001	Ni .	nil
Fe .	0·0001	As .	trace	Co .	„
Bi .	0·0001	Se + Te	„	Sn .	„

High-Purity Antimony

Element	%	Element	%
Pb . . .	0·006	As . . .	0·006
S . . .	<0·001	Na . . .	0·005
Cu . . .	<0·001	Fe . . .	0·005
Ni . . .	<0·001	Sb (by difference).	99·975

Chemical analysis of the binary lead–0·85% antimony alloy showed that it conformed to its nominal composition.

The ternary alloys investigated contained 0·85% antimony with separate additions of 0·01% silver, copper, and tin, and 0·001, 0·002, 0·005, 0·01, and 0·02% arsenic. Tests were also carried out on binary lead–arsenic alloys containing the same amounts of arsenic as in the ternary lead–antimony–arsenic alloys. Spectrographic analysis for these minor additions showed that the nominal and actual amounts present were almost identical.

2. Preparation of the Alloys

(a) *The Binary Lead–0·85% Antimony Alloy, Extruded as Rod*

The basis lead was melted in a steel pot held at 400° C. in a thermostatically controlled furnace. The required amount of antimony was dissolved in a small quantity of the lead at 450° C., which was then added to the main melt at 400° C. The metal was chill cast from a bottom-pouring crucible. The cast billets were homogenized for one week at 300° C., machined, and then extruded in a laboratory extrusion press as straight lengths of rod, $\frac{7}{16}$ in. in dia.

(b) *The Binary and Ternary Alloys Cast as Sticks*

Small charges of the extruded binary lead–0·85% antimony alloy rods were remelted in Salamander crucibles, coated on the inside with a wash of alumina. The minor additions of tin and silver were made directly to the melts, and copper and arsenic were added as 1% and 1·5% master alloys, respectively. The ternary alloys were chill cast into cylindrical sticks, which were homogenized for 4 days at 300° C. and then quenched in water. The binary lead-arsenic alloys were made up in a similar way.

3. Hardness Determinations

Hardness determinations were made on metallographically polished specimens with a Vickers pyramid hardness testing machine, a load of 1 kg. being applied for 30 sec. The hardness of a material was taken as the average of the values obtained from six impressions.

IV.—RESULTS OF HARDNESS DETERMINATIONS

1. The Binary Lead–0·85% Antimony Alloy

The lead–0·85% antimony alloy was extruded at each of the temperatures 160°, 200°, 250°, and 300° C. and the product air-cooled. Metallographic examination showed that the alloy was single-phase at 150° C. and higher temperatures, in accordance with the results of Obinata and Schmid.[7]

The hardness of the four rods determined immediately after extrusion was independent of grain-size, which differed as a result of the various temperatures of extrusion. The average hardness of the materials was 5·1. The four rods were aged at 20° C., but did not age-harden even after 12,000 hr. The rod extruded at 300° C. was solution-treated for 24 hr. at 175° C. and water-quenched. This material was aged at 50° and 100° C., but did not age-harden even after 2800 and 770 hr., respectively, when the tests were discontinued.

Since prestraining is known to increase the rate of age-hardening, the effect of this variable was examined. The rod extruded at 300° C. was solution-treated for 24 hr. at 175° C. and then prestrained 5 and 10% in tension; the initial hardnesses increased from 5·1 to 6·6 and 7·2, respectively. Only slight further hardening occurred in 5800 hours' ageing at 20° C., hardnesses of 7·5 and 8·5, respectively, being reached.

To examine the effect of a different temperature of solution-treatment, the as-extruded rod was heated at 300° C. for 2 hr. and water-quenched. Slow hardening took place during ageing at 50° C., the hardness increasing from 5·4 to 9·0 in about 1800 hr.

2. Effect of Various Minor Additions on Cast Lead–0·85% Antimony Alloy

Ternary alloys containing separate 0·01% additions of arsenic, copper, tin, and silver were homogenized for four days at 300° C., water-quenched, and aged at 20° C. The ternary alloys containing tin and copper did not harden during 1400 hr. at 20° C., whereas after a similar time the hardness of the arsenic-bearing material had increased from 7·5 to 13·8. . The silver-bearing alloy aged more slowly than that containing arsenic, the hardness increasing from 6·7 to 7·7 in 1400 hr. at 20° C.

3. Effect of Various Additions of Arsenic on Cast Basis Lead and Lead-0·85% Antimony Alloy

Since arsenic had the greatest effect of the addition elements examined, a more extensive investigation was made of ternary alloys with a range of arsenic contents. It was also desirable that the effect of various arsenic additions to the basis lead should be examined. Ternary alloys containing 0·001, 0·002, 0·005, 0·01, and 0·02% arsenic, and binary lead–arsenic alloys containing the same amounts of arsenic were made up and homogenized for four days at 300° C.

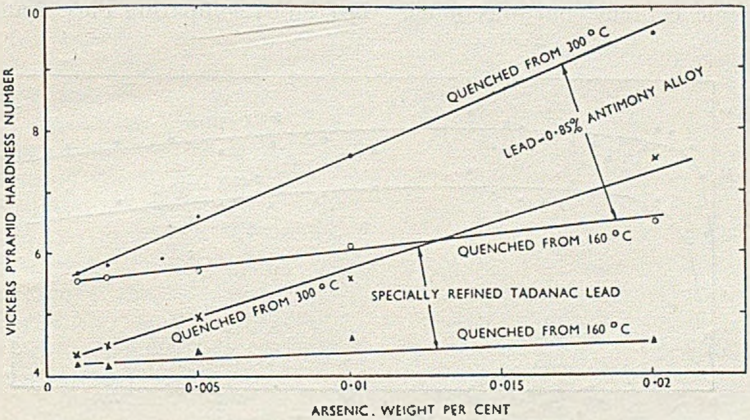


Fig. 1.—Effect of Arsenic Content and Quenching Temperature on the Initial Hardness of the Basis Lead and the Lead-0·85% Antimony Alloy.

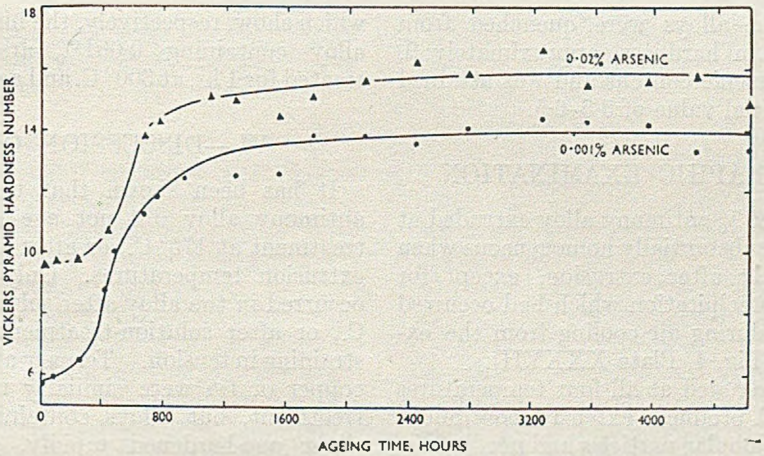


Fig. 2.—Age-Hardening Curves for Lead–0·85% Antimony Alloy Containing Various Amounts of Arsenic. Water-quenched from 300° C. and aged at 20° C.

(a) *The Solubility of Arsenic in the Basis Lead and the Lead–0·85% Antimony Alloy*

The hardnesses of the lead–arsenic and lead–0·85% antimony–arsenic alloys immediately after water-quenching from 300° and 160° C. are shown in Fig. 1. These hardness values are characteristic of the respective solution-treated alloys, because no significant change in hardness can have occurred between quenching and hardness testing (see later paragraphs and Fig. 2). The hardness values are linearly related to the arsenic contents of the alloys, although the lines corresponding to the two heat-treatment temperatures differ markedly in slope. This indicates that the arsenic was in solid solution and mainly out of solid solution, respectively, after heat-treatment at 300° and 160° C. On this interpretation, the intersections of the straight lines for the materials heat-treated at 160° and 300° C. indicate the approximate solid solubility of arsenic at 160° C. It appears that in both the binary and ternary alloys the solid solubility of arsenic at 160° C. is less, possibly much less, than 0·001%.

(b) *The Effect of Ageing*

Each ternary alloy, water-quenched from 300° C., was aged at 20° and 50° C., and the age-hardening curves obtained are shown in Figs. 2 and 3. For the sake of clarity the curves of the alloys containing 0·001 and 0·02% arsenic, only, are plotted for each temperature. The curves for all the other materials lay between these two extremes. Although the initial hardness increased markedly with arsenic content, the rate of hardening at both temperatures was independent of the arsenic content, as shown by the parallelism of the curves.

Ageing tests at 20° C. on the binary lead–arsenic alloys after water-quenching from 300° C. showed that these materials did not age-harden.

The rate of hardening, at 50° C., of the ternary alloys water-quenched after 2 hr. at 160° C. was again independent of the arsenic content, but was much slower than when the alloys were quenched from 300° C. The maximum hardness, approximately 9, was independent of arsenic content and was attained in 380 hr. from an initial value of 5·5–6·5.

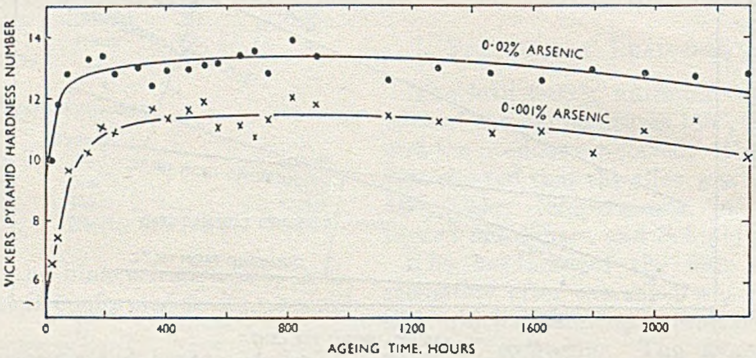


Fig. 3.—Age-Hardening Curves for Lead–0·85% Antimony Alloy Containing Various Amounts of Arsenic. Water-quenched from 300° C. and aged at 50° C.

V.—METALLOGRAPHIC EXAMINATION

The binary lead–0·85% antimony alloy extruded at any temperature was substantially homogeneous when examined immediately after extrusion, except for slight discontinuous precipitation which had occurred at grain boundaries during air cooling from the extrusion temperature (Fig. 4, Plate XXXVII).

Ageing the alloy extruded at all four temperatures for 12,000 hr. at 20° C. produced extensive continuous precipitation as fine globular particles and needles distributed throughout the grains (Fig. 5, Plate XXXVII). The width of the band of discontinuous precipitation had only slightly increased.

The alloy extruded at 300° C. and solution-treated for 24 hr. at 175° C. showed extensive continuous precipitation after subsequent ageing at both 50° and 100° C. A slight increase in the width of the discontinuous precipitation at grain boundaries occurred during the first few hours of ageing, but there was no further change. When the tests were abandoned, continuous precipitation was extensive. Discontinuous precipitation was no longer distinguishable at the grain boundaries, since the amount of antimony in solid solution within the grains had become similar to that in the new solid solution formed during the discontinuous precipitation.

Extensive continuous precipitation occurred, during ageing at 20° C., along the slip planes in the alloy extruded at 300° C., solution-treated, and overstrained by both 5 and 10%. Only a slight increase in discontinuous precipitation occurred after the first few hours, and there was no further change.

The presence of arsenic in the lead–0·85% antimony alloy increased the dispersion, and possibly the rate, of continuous precipitation, although there was no effect on discontinuous precipitation. This fact can be seen by comparing Figs. 6 and 7 (Plate XXXVII), which show, respectively, the binary alloy and ternary alloy containing 0·001% arsenic, both solution-treated for 1 hr. at 300° C. and aged for 120 hr. at 50° C.

VI.—DISCUSSION OF RESULTS

It has been shown that the binary lead–0·85% antimony alloy did not age-harden after solution-treatment at 175° C., or after air-cooling from higher extrusion temperatures. Only slight age-hardening occurred in the alloy after solution-treatment at 300° C., or after solution-treatment at 175° C. and pre-straining in tension. Ternary alloys containing 0·01% copper or tin were similarly unresponsive to heat-treatment, but alloys containing 0·01% arsenic or silver age-hardened rapidly, the arsenic addition having the most marked effect. The rate of age-hardening of the ternary alloys containing arsenic was independent of the arsenic content, and 0·001% arsenic had almost as much effect as 0·02%.

The rate of age-hardening of the ternary alloys containing arsenic has been shown to be greater after heat-treatment at 300° C. than at 160° C. This difference can be attributed to the arsenic being in solid solution after solution-treatment at 300° C., but being probably nearly all rejected from solid solution during the treatment at 160° C. It is therefore probable that the slight age-hardening in the binary lead–0·85% antimony alloy, after solution-treatment at 300° C., was due to some minute quantity of impurity being in solid solution. When the impurity was rejected from solid solution, by heat-treatment at 175° C., or by air-cooling from higher extrusion temperatures, no age-hardening took place.

It might be argued that the difference in the rate of hardening between the arsenic-bearing ternary alloys heat-treated at, and water-quenched from, 160° and 300° C. is due entirely to the difference between the resulting quenching strains. For example, Phillips and Brick[8] have shown that, for aluminium alloys, the rate of age-hardening is markedly increased by the presence of quenching strains. Although quenching strains may have some effect in the present work, the observed results were not entirely due to this cause for two reasons. Firstly, hardening of the cast sticks occurred uniformly across their diameters; had there been a major effect of quenching strains the most rapid increase of hardness would have occurred near the outside circumference of the sticks, which must have been the most severely strained region during quenching. Secondly, a specimen of the ternary alloy, containing 0·001% arsenic, was cooled freely in air after solution-treatment at 300° C. Any cooling strains in this material must have been less than those in the quenched alloys, but the specimen age-hardened at 50° C. at the same rate as the material quenched from 160° C.

The literature records that a minor addition of copper accentuates the age-hardening of lead–antimony alloys, but the present work does not confirm this observation.

Any theory which is put forward to explain the observed phenomena recorded above must account for the following facts:

(i) The minor addition must be in solid solution in the lead–0·85% antimony alloy to be effective in causing age-hardening.

(ii) The minor addition of arsenic increases the dispersion, and probably the rate, of precipitation of the antimony-rich phase.

(iii) A small amount of tin, which is in solid solution, does not cause age-hardening.

(iv) The rate of age-hardening is increased markedly by 0·001% arsenic, which is in solid solution. Above 0·001% and up to 0·02% arsenic the rate of hardening is independent of the arsenic content.

A simple hypothesis in which the antimony-rich precipitate is nucleated by the prior precipitation of the minor addition is not tenable, since the insoluble arsenic rejected, at 160° C., from solid solution has only a small effect on the rate of age-hardening.

The theory put forward by Hardy[9] to account for the effect of soluble minor additions on the age-hardening characteristics of an aluminium–4% copper–0·15% titanium alloy cannot be applied directly to explain the present phenomena. In the alloys investigated by Hardy, the chief alloying element, copper, and the minor additions had atomic diameters smaller and larger, respectively, than that of aluminium. Hardy postulated that the small atoms of copper in solid solution form an "atmosphere" around the large atoms of the minor additions in such a way as to reduce the strain energy of the system to a minimum, such an arrangement being favourably placed for precipitation under suitable conditions. In the lead–0·85% antimony–arsenic alloys, however, both antimony and arsenic have atomic diameters less than that of lead, so that, providing that substitutional solid solutions are formed, the strain energy of the system will be a minimum when the atoms of antimony and arsenic tend to avoid each other and do not form an atmosphere of one type of atom around the other.

Although there seems to be no completely adequate theory to explain all the observed phenomena, the authors are indebted to Dr. W. Hume-Rothery for his suggestion as to the possible mechanism of the effect of the arsenic addition. He suggests that, although the diameter of the interstitial octahedral holes in the lead lattice are 0·41 times the diameter of the lead atom, assuming the atoms to be spheres, the high compressibility of lead atoms might permit arsenic atoms, which have an atomic diameter 0·7 times that of lead, to take up interstitial positions. In the interstitial compounds of the transition elements, the limiting radius ratio for filling octahedral holes is about 0·6, so that an extension to 0·7 for lead would not seem unreasonable in view of the much higher compressibility. The arsenic atoms would, therefore, produce a local expansion of the lead lattice, so that under these conditions the strain energy of the system would be reduced when the substitutional antimony atoms, which cause local contraction of the lead lattice, form an atmosphere around the arsenic atoms. This arrangement, according to Hardy's theory, would be favourably placed for precipitation. Dr. Hume-Rothery points out that an arsenic atom in an interstitial hole would have six close neighbours, and that if some of these positions were occupied by antimony atoms, the bonds would be at right angles and might then provide a nucleus whose shape was favourable for the formation of the antimony structure which, although an (8-*N*) rule structure, is to a rough approximation nearly a simple cube. This effect might be expected for arsenic, which itself forms approximately right-angled bonds, but not for normal metals. Dr. Hume-Rothery, however, considers it unlikely that this theory can explain the effect of the silver addition, which is similar to that of the arsenic addition though less pronounced, since the atomic diameter of silver makes it unlikely that this element forms interstitial solid solutions in lead.

ACKNOWLEDGEMENTS

The authors wish to thank the Director and Council of the British Non-Ferrous Metals Research Association for permission to publish this paper.

REFERENCES

1. W. Hofmann, A. Schrader, and H. Hanemann, *Z. Metallkunde*, 1937, **29**, 39.
2. K. S. Seljesater, *Trans. Amer. Inst. Min. Met. Eng., Inst. Metals Div.*, 1929, 573.
3. M. Bluth and H. Hanemann, *Z. Metallkunde*, 1937, **29**, 48.
4. E. E. Schumacher, G. M. Bouton, and L. Ferguson, *Indust. and Eng. Chem.*, 1929, **21**, 1042.
5. E. E. Schumacher, *Trans. Amer. Inst. Min. Met. Eng.*, 1950, **188**, 1097.
6. W. Hofmann and H. Hanemann, *Z. Metallkunde*, 1938, **30**, 416.
7. I. Obinata and E. Schmid, *Metallwirtschaft*, 1933, **12**, 101.
8. A. Phillips and R. M. Brick, *Trans. Amer. Inst. Min. Met. Eng.*, 1934, **111**, 94.
9. H. K. Hardy, *J. Inst. Metals*, 1950–51, **78**, 169.

THE EFFECT OF COLD WORK ON THE MICROSTRUCTURE AND CORROSION-RESISTANCE OF ALUMINIUM–5% MAGNESIUM ALLOYS CONTAINING 0–1% ZINC*

By P. BRENNER,† Dr.Ing., Member, and G. J. METCALFE,‡ M.Sc.Tech., Member

SYNOPSIS

The effects of cold work, of prolonged ageing at 70° C., and of additions of up to 1% zinc on the microstructure, corrosion-resistance, and stress-corrosion-resistance of an aluminium alloy containing 5% magnesium, have been investigated.

The presence of zinc has little effect on the mechanical properties, but the alloy containing 1% zinc shows pronounced precipitation after ageing at 70° C., and corrosion attack is greater than in the case of the 0–0·5% zinc 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 maximum corrosion attack, at about 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 maximum 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.

I.—INTRODUCTION

The corrosion and stress-corrosion properties of aluminium–magnesium alloys have usually been investigated on alloys containing more than 5% magnesium. Wrought alloys containing 7–10% magnesium have been found particularly susceptible to intercrystalline corrosion and stress-corrosion in certain conditions of heat-treatment and cold working, and it has been shown that there is a correlation between microstructure and intercrystalline attack.[1, 2, 3] Few published results relate to alloys with lower magnesium contents. Service failures in aluminium–5% magnesium alloys have been attributed to intercrystalline corrosion, probably accentuated by stress, by one of the present authors,[4] who found that aluminium–5% magnesium alloy rivets failed as a result of severe intercrystalline corrosion, particularly at the junction of the shank and head. In this region the rivets had been very severely cold worked during driving, and heavy precipitation had occurred owing to exposure of the material to tropical temperatures. The corrosion attack was most severe in the regions of heaviest precipitation.

Since the aluminium–magnesium alloys are of great practical importance, it was considered advisable to obtain more information, particularly on their behaviour under tropical conditions. Siebel and Vosskühler [5] have suggested that small additions of zinc have a pronounced influence on the corrosion-resistance of aluminium–magnesium alloys, and the investigation was therefore extended to cover aluminium–5% magnesium alloys with and without additions of zinc up to 1%. Subsequent work by Vosskühler [6] has dealt with alloys of higher zinc content. In the present work the alloys were prepared in sheet form and were solution-treated and quenched from various temperatures. Subsequently the material was given up to 50% reduction in thickness by cold rolling and was tested both in the solution-treated condition and after ageing at 70° C. This temperature was chosen as being one likely to be reached in direct tropical sunshine.

II.—EXPERIMENTAL PROCEDURE

1. Preparation of Materials

The approximate chemical composition, treatment, and method of identification of the specimens were as follows :

(i) *Nominal Compositions* :—(1) Al–5% Mg; (2) Al–5% Mg–0·2% Zn; (3) Al–5% Mg–0·5% Zn; (4) Al–5% Mg–1% Zn.

---

* Manuscript received 16 July 1952.

† Director of Research, Vereinigte Aluminiumwerke A.G. and Vereinigte Leichtmetallwerke G.m.b.H., Bonn (Rhein), Germany; formerly at the Royal Aircraft Establishment, Farnborough.

‡ Senior Research Metallurgist, Precious Metals Division, The Mond Nickel Co., Ltd., Acton; formerly at the Royal Aircraft Establishment, Farnborough.

(ii) *Treatments*:—*T*—2 hr. at 500° C., cooled to 300° C., held 1 hr., and cold-water quenched. *U*—2 hr. at 500° C., cooled to 280° C., held 1 hr., and cold-water quenched. *V*—2 hr. at 500° C., cooled to 260° C., held 1 hr., and cold-water quenched.

(iii) *Cold Work*:—0, 5, 10, 20, 30, 50% reduction by cold rolling.

(iv) *Ageing Treatments*:—*C*—14 days at 70° C.; *D*—28 days at 70° C.

Four alloys of the above nominal compositions were prepared, using metals of the following purity: aluminium 99·8%, magnesium 99·9%, and zinc 99·9%. Sufficient material to make two ingots, 5 in. in dia. and 16 in. long, in each alloy was melted and degassed with aluminium chloride at 680° C. The alloy was cast in water-cooled steel moulds, the water flow for each cast being maintained at 3 gal./min. The top 4 in. of each ingot was discarded, and the remainder machined to two portions each 4½ in. in dia. and 5¾ in. long.

The machined ingots were homogenized for 48 hr. at 450° C. before forging. The initial forging temperature was 380° C., but some cracking occurred, and subsequent operations were begun at 410° C. The cast alloy was forged to slabs approximately 5 in. wide and ⅞ in. thick, and was machined after forging and before hot rolling.

Reduction of the material to 0·6 in. thick was carried out at 410° C., but subsequently the temperature was reduced to 390° C. and the material hot rolled to 0·072 in. thick at this temperature. Some of the material was further hot rolled at the same temperature to thicknesses of 0·050, 0·045, 0·040, and 0·038 in. All the four alloys rolled satisfactorily.

After hot rolling, all the material was annealed for 20 hr. at 380° C. and cooled in the furnace. The sheets were then cut into pieces which were given 1 hour's treatment at 390° C. to relieve stresses due to cutting.

The specimens *T*, *U*, and *V* were solution-treated for 2 hr. at 500° C. and transferred directly to ovens at 300°, 280°, and 260° C., respectively, held for 1 hr. and then cold-water quenched. These temperatures were chosen to cover both sides of the limit of solid solubility of 5% magnesium in aluminium.

Specimens of each alloy were given 50, 30, 20, 10, and 5% cold reductions in accordance with the following schedule, so that the finished material was all 0·036 in. thick.

Redn.		Passes
50% :	0·072 in. in 0·005 in. stages	(8)
30% :	0·050 in. in 0·005 in. stages	(3)
20% :	0·045 in. in 0·003 in. stages	(3)
10% :	0·040 in. in 0·002 in. stages	(2)
5% :	0·038 in. in 0·002 in. stage	(1)

The actual chemical compositions of the alloys used, as determined by chemical analysis, were:

Alloy	Mg, %	Zn, %	Fe, %	Si, %	Al
No. 1	4·98	Nil	0·1	0·07	Rest
No. 2	4·81	0·25	0·1	0·07	,,
No. 3	5·02	0·45	0·1	0·09	,,
No. 4	5·01	1·10	0·1	0·08	,,

2. MICRO-EXAMINATION BEFORE CORROSION

All the specimens were electrolytically polished on sections parallel to the rolling direction and perpendicular to the rolled surface, etched after polishing, and then examined microscopically. The electropolishing and etching techniques are described in the Appendix (p. 268).

The etching characteristics of the alloys change with their chemical composition and structure, the attack becoming greater with increasing zinc content. There was therefore a tendency for specimens of alloy 4 to be slightly over-etched, whereas alloy 1 was slightly under-etched.

The variations in zinc content, heat-treatment, and cold reduction had quite different effects on the microstructures. Some of these variables, such as small amounts of cold work or additions of zinc less than 0·5%, had only a slight effect on the microstructure, whereas other conditions, such as ageing at 70° C., the addition of 1% zinc, and large amounts of cold work, affected the microstructure considerably. The photomicrographs described are examples of typical structures produced by the different treatments.

3. MICROSTRUCTURES OF UNAGED SPECIMENS

The microstructures of alloys 1, 2, and 3 in the unaged condition were very similar, the grain boundaries usually being only very faintly revealed. One of the chief observations was the occurrence of etch-pits at the junctions of three grain boundaries, which tended to increase in number with decreasing zinc content. Occasionally, randomly distributed pits were also observed within the grains. These tended to increase in number with increasing zinc content, there being many more in alloy 4 than in the other three alloys. There was some indication that in alloy 4 the number of pits within the grains increased as the quenching temperature was raised. This was probably due to the reduced solubility of magnesium in aluminium in the presence of zinc and to the increased precipitation at temperatures below the solid-solubility line.[7]

After the specimens had been rolled to 5, 10, and 20% reduction in thickness, it was found that those in the worked but not aged condition had similar structures to those of the unworked specimens, except that there was probably a slight increase in the number of pits as the amount of rolling was increased. Heavier cold reductions resulted in an increase in the number of pits within the grains.

4. MICROSTRUCTURES OF SPECIMENS HEAT-TREATED AND AGED AT 70° C. (NOT COLD WORKED)

Ageing at 70° C. for 14 days affected the microstructures of the four alloys to different extents. Alloys 1, 2, and 3 were very similar, in that in specimens that had not been cold worked, an incomplete grain-boundary network was formed, as a result of precipitation occurring only on certain boundaries between grains, as shown in Fig. 6 (Plate XXXVIII). In addition, pits were observed at grain-boundary junctions. There was no indication of any variations in microstructure due to the different heat-treatments *V*, *U*, and *T*. In alloy 4 a fine precipitate was observed at some parts of the grain boundaries, together with a few coarse plate-like particles.

In order to determine whether the incomplete network in the aged specimens was due to microsegregations originating from the cast structure which may not have been completely removed by homogenization, specimens of alloy 1 were re-heat-treated at 460° C. for 16, 24, 40, and 64 hr., and water-quenched. No influence of the annealing temperature on the microstructure could be detected after subsequent ageing for 14 days at 70° C. The incompleteness of the grain-boundary network was very marked in all the specimens. Prolonging the time of etching caused stronger attack on the etched portions of the grain boundaries, which became more pronounced, but there was no tendency to form a complete network.

Examination of aged specimens that had been electrolytically polished and etched so as to produce etch pits,[8, 9] indicated that there was no preferred orientation of the grains, as judged by the changes in shape of the rectangular and triangular pits from grain to grain. It was also shown that quenching stresses played no part in the formation of the incomplete network, since structures identical with those previously described were observed after examination of specimens that had been aged at 70° C. after (*a*) cooling in air from 460° C. or (*b*) quenching in boiling water after soaking for 2 hr. at 460° C.

It is considered that the incomplete nature of the grain-boundary network is related to the difference in orientation between adjacent grains and possibly to the direction of the grain boundary.[10] The distortion of the lattice in the grain-boundary zone, which varies appreciably with the difference in orientation of adjacent grains, influences the occurrence of precipitation.

5. MICROSTRUCTURES OF SPECIMENS COLD ROLLED AND AGED AT 70° C.

The effect of ageing the cold-worked material at an elevated temperature (i.e. 14 days at 70° C.) was very marked; 5, 10, and 20% cold reduction led to an increasing amount of precipitation in the grain boundaries and the gradual appearance of a more complete network with increased amounts of cold work. Specimens of alloy 1 that had been given 20% cold reduction had still an incomplete grain-boundary network of precipitation, as shown in Fig. 7 (Plate XXXVIII), and in a few grains there was very slight indication of precipitation on certain crystallographic planes (the slip planes). After 30% cold reduction the amounts of precipitation in the slip planes occurring after ageing gradually increased, but the distribution and density of the precipitate was not uniform from grain to grain. Ageing after 50% cold reduction resulted in heavy precipitation, but again it was obvious that the degree of precipitation varied considerably from grain to grain.

In the zinc-containing alloys (2, 3, and 4) precipitation along slip lines was even more marked. Areas of heavy precipitation occurred in the middle of the grains as well as at the grain boundaries, at which the slip lines were frequently heavily distorted.

As might be expected, 50% cold reduction followed by ageing resulted in pronounced slip-plane precipitation in all four alloys. The grains in all the specimens were markedly elongated in the direction of rolling, and precipitation was heavier in some regions than in others. In fact, in some grains very little precipitation occurred. Fig. 8 (Plate XXXVIII) shows a typical example of the microstructures observed. Frequently a band of heavy precipitation in a series of slip lines crossed more than one grain without its direction being affected by the grain boundaries. Where this occurred there was severe distortion of the grain boundary, and marked steps were visible where the precipitation bands crossed the boundaries.

In specimens aged for 28 days at 70° C. structures were observed similar to those in specimens aged for 14 days at 70° C., but precipitation at the grain boundaries and along the slip lines was slightly heavier. Some specimens showed areas with steps in the grain boundaries caused by slip-plane movement (Fig. 9, Plate XXXVIII). Heavy precipitation also occurred on slip planes in the region of the grain boundary where the slip lines are bent at the boundary (Fig. 10, Plate XXXVIII). The neighbouring grain in Fig. 10 is free from any slip-plane precipitation.

In order to examine the development of precipitation and to determine the critical ageing time at which grain-boundary precipitation and precipitation along the crystallographic planes starts, specimens of alloys 1 and 4 were aged after treatments *V*, *U*, and *T* and 30% cold reduction for 2, 5, 8, 24, 80, 200, and 960 hr. at 70° C. No evidence of precipitation could be detected microscopically after periods up to 80 hr., but after 200 hr. precipitation started at the grain boundaries in all specimens, and faint slip lines appeared in some grains. The network, however, was incomplete. After 960 hr., alloy 1 showed an almost complete network and alloy 4 a complete network, and in both alloys many slip lines and bands were visible.

6. CORROSION TESTS

Specimens in the form of strips 4 in. long × 0·75 in. wide × 0·036 in. thick were suspended in a shed with one side (west) open and were sprayed with natural sea-water three times a day for five days a week. The tensile properties were determined before exposure and after 28 days' and 12 months' exposure, the test-pieces being cut from the corroded strips and tested in triplicate.

The results of tensile tests on unexposed materials in the aged and unaged conditions are shown graphically in Fig. 1, plotted against the amount of cold reduction. Since no influence of the different heat-treatments *T*, *U*, and *V* was detected, each point on these graphs is the average figure for all three treatments. In the unaged condition, the tensile strength and, in particular, the proof stress are considerably increased with increasing cold reduction, whilst the elongation decreases. After ageing for 28 days at 70° C., the ultimate tensile strength is not much affected, but in the cold-rolled condition the proof stress is clearly reduced and the elongation increased, as a consequence of recovery. The differences between the proof stress and elongation in the unaged and aged conditions appear to be less in alloys 3 and 4 than in alloys 1 and 2. Comparison of the properties in the unaged and aged conditions shows that age-hardening effects, if any, must be very small. The addition of zinc up to 1% has apparently no marked effect on the tensile properties of an aluminium–5% magnesium alloy under the conditions used.

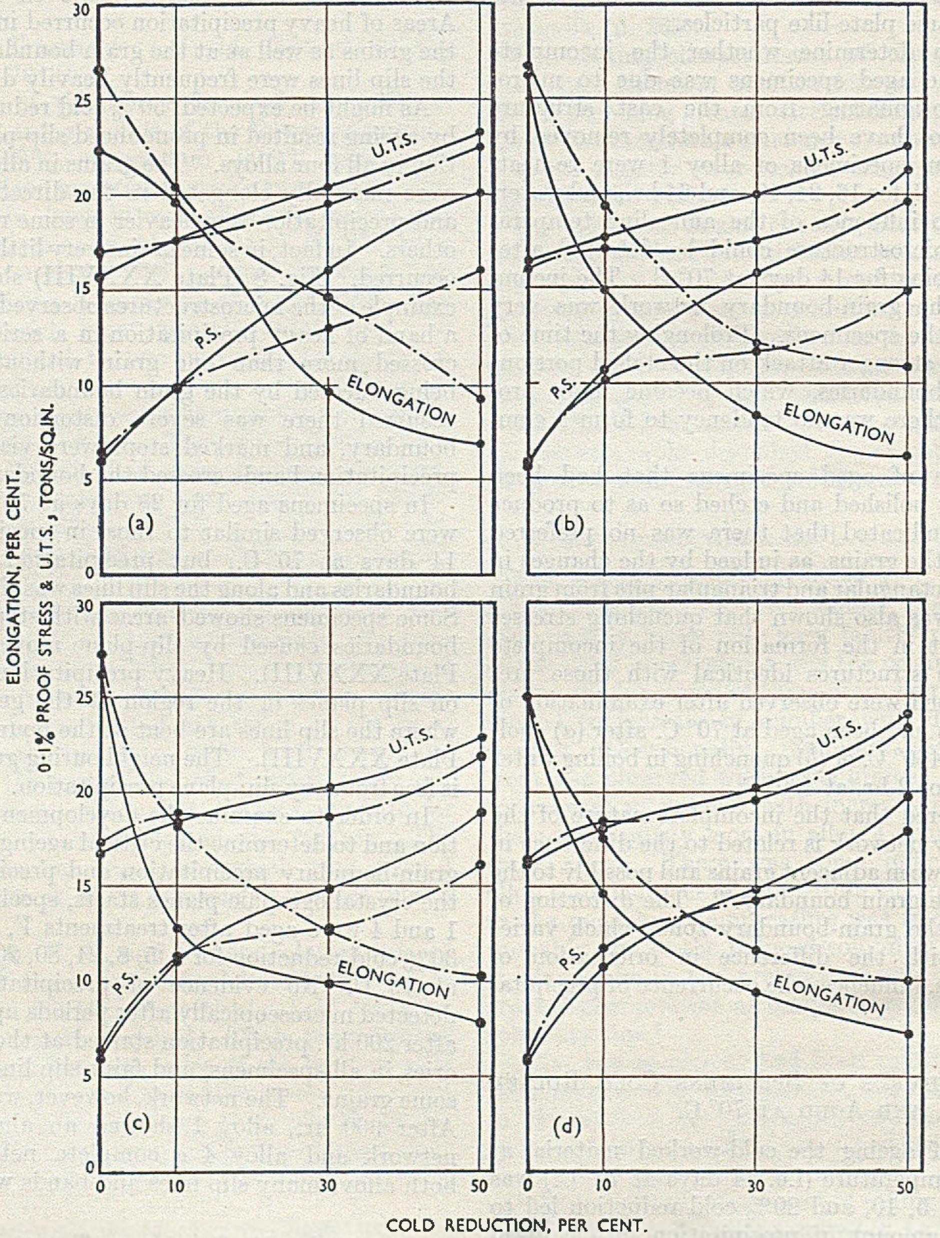


Fig. 1.—Mechanical Properties of (*a*) Alloy 1, (*b*) Alloy 2, (*c*) Alloy 3, and (*d*) Alloy 4.

——————— Before ageing. —·—·—·— After ageing.

After 28 days' exposure, the specimens were only slightly attacked. On the surface of a number of specimens a few small pits were observed, which did not appear to be associated with a particular alloy or condition, but rather with inhomogeneities or local variations in the surface of the specimens (e.g.

rolled-in impurities) or local differences due to the cold work.

After 12 months' exposure corrosion was greater, but the attack was not uniform over the surface, and there were areas in most of the specimens that appeared almost unattacked. Numerous specimens showed local severe pits, which apparently increased with the zinc content. The pits sometimes formed striations in the rolling direction, particularly in alloy 4. The differences in the degree of attack of specimens of the same alloy and condition were frequently as great as the differences in the attack of different alloys and conditions. No difference between *V*-, *U*-, and *T*-treated specimens was detected, but the aged specimens appeared to be slightly more attacked than the unaged specimens. The 50% cold-rolled specimens, especially alloy 1 in the unaged condition, appeared to be less attacked than the other specimens.

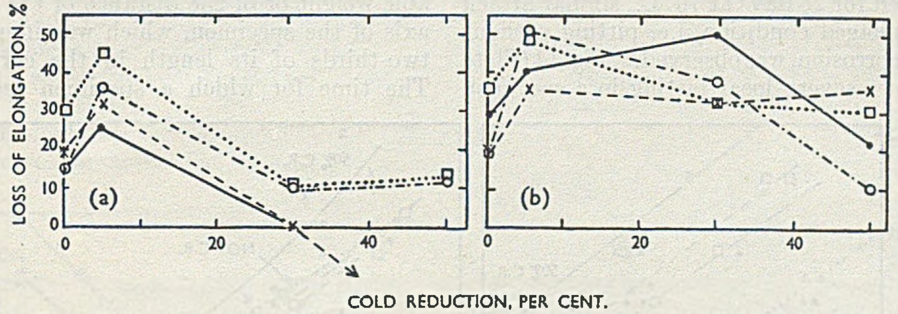


Fig. 2.—Percentage Loss of Elongation After 12 Months' Corrosion (*a*) Unaged and (*b*) Aged at 70° C.

Key.

●———● Alloy 1.	×— — —× Alloy 2.	o—·—·—o Alloy 3.	□· · · · □ Alloy 4.

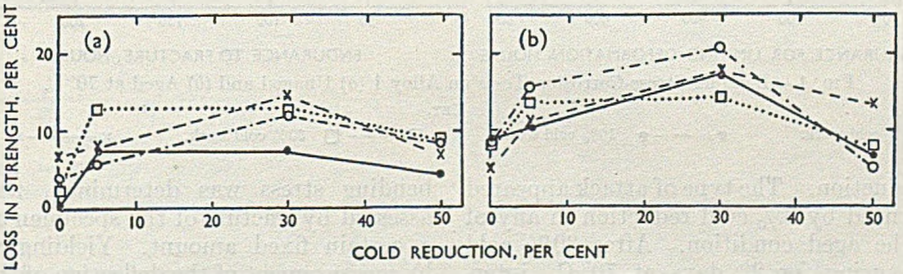


Fig. 3.—Percentage Loss of U.T.S. After 12 Months' Corrosion (*a*) Unaged and (*b*) Aged at 70° C.

Key as for Fig. 2.

The tensile strength and elongation were scarcely affected after 28 days' exposure, and only a few aged specimens of alloys 1, 2, and 3 with 5% and 30% cold reduction showed losses in elongation. The scatter in tensile strength and elongation was greater than before corrosion, however.

As would be expected, after 12 months' exposure the specimens had suffered greater corrosion attack, and the scatter range for the tensile strength was ±20% and for the elongation ±40%. This high scatter may be due to the non-uniformity of the attack and possibly due also to segregation in some specimens causing severe local corrosion. The results are expressed graphically in Figs. 2 and 3. The superiority in tensile strength produced by high degrees of cold reduction was maintained after 12 months' attack in both the unaged and aged conditions. It is clear, however, that there was a general tendency for loss of strength to increase with increasing cold reduction up to about 30%.

With 50% cold reduction the loss in strength due to corrosion was generally less than with smaller reductions and was comparable with that of material that had not been cold worked. It seems possible therefore that there is a critical amount of cold reduction in the range between 5 and 50% which produces the maximum reduction of tensile strength due to corrosion.

In both the aged and unaged condition increasing zinc content produced no improvement in corrosion-resistance.

7. Micro-Examination of Corroded Specimens

The specimens for micro-examination were cut from the corroded samples before making the tensile test, in such a way that the polished surface was in the direction of rolling and perpendicular to the rolled surface. Though the surface corrosion was very slight after 28 days' exposure, in some specimens a little intercrystalline corrosion had occurred. After 12 months' exposure the attack had proceeded sufficiently for comparison to be made of the corrosion of the material in different conditions. In the unaged condition slight superficial corrosion with a tendency to intercrystalline attack was observed in the *V*-, *U*-, and *T*-treated specimens of alloy 1 (see Fig. 11, Plate XXXVIII). As a consequence of the elongation of the grains by cold reduction, the direction of the intercrystalline cracks formed small angles with the surface in specimens of high cold reduction. No intercrystalline corrosion was detected in alloy 4, but local heavy pits were observed (Fig. 12, Plate XXXVIII). The extent of pitting in alloys 2 and 3 was intermediate between that of alloys 1 and 4. The attack in alloy 2 resembled the intercrystalline attack of alloy 1, while alloy 3 showed a slight tendency to intercrystalline attack and local pits as in alloy 4.

In the aged condition, intercrystalline corrosion was very pronounced in alloy 1 without cold reduction, though only localized, as shown in Fig. 13 (Plate XXXVIII). Similar attack occurred in alloy 2 and to some extent in alloy 3, without cold reduction.

In alloy 4, aged for 28 days at 70° C., similar attack to that in the unaged condition, i.e. pitting without intercrystalline corrosion, was observed. Fig. 14 (Plate XXXVIII) shows severe local pitting in a specimen with 5% cold reduction. The type of attack appeared not to be influenced by 5% cold reduction in any of the alloys in the aged condition. After 30% cold reduction and ageing for 28 days at 70° C., intercrystalline corrosion was very pronounced in alloy 1 (Fig. 15, Plate XXXVIII), whereas alloy 4 in the same condition, showed only small pits. After 50% cold reduction and ageing, in addition to intercrystalline corrosion, attack along slip planes was detected in alloys 1, 2, and 3 (Fig. 16, Plate XXXVIII). Alloy 4, after 50% cold reduction and ageing, generally showed local pitting, with a slight tendency to intercrystalline attack in a few areas (Fig. 17, Plate XXXVIII), but no slip-plane attack was detected.

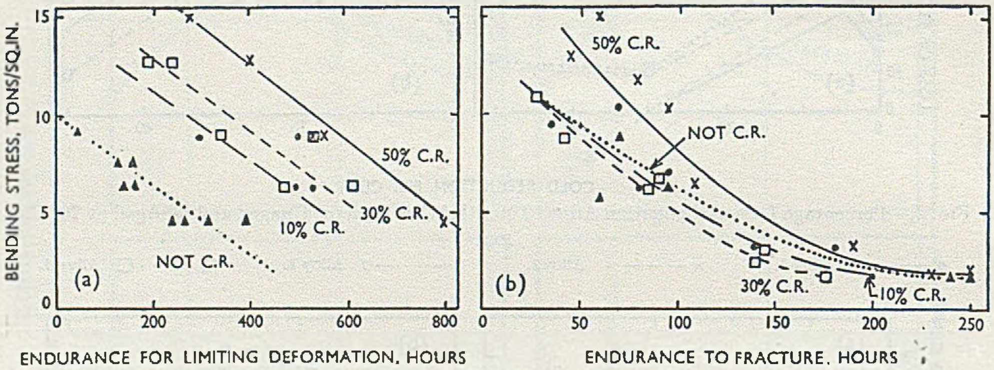


Fig. 4.—Bending Stress-Corrosion Tests on Alloy 1 (*a*) Unaged and (*b*) Aged at 70° C.

Key.

▲ · · · · · ▲ 0% cold work. ● — — — ● 10% cold work. □ - - - - □ 30% cold work. × ——— × 50% cold work.

There was no indication of changes in the type or extent of corrosion attack that could be attributed to differences in the V, U, and T heat-treatments of any of the alloys in either the aged or unaged condition, either cold worked or as heat-treated.

8. Stress-Corrosion Test Results

Some stress-corrosion tests were made in an accelerated corrosion-testing solution containing 3% sodium chloride and 1·2% hydrochloric acid by weight. The specimens were immersed in the solution and stressed in bending and also in direct tension. The solution was not renewed during the test, but distilled water was added daily so that the volume remained constant. During the testing of alloys 1, 2, and 3 little change in the pH of the solution occurred, but during tests of alloy 4 there was a rise of pH. In the bending stress-corrosion test, the specimen, which was $\frac{3}{4}$ in. wide and about 4 in. long, was attached at the lower end to the bottom of a glass tank containing the corroding solution. The upper end of the specimen was attached to a lever at right-angles to the axis of the specimen. This lever was approximately 12 in. long and was so loaded that the desired bending stress could be produced by alteration of an adjustable weight or of the distance of the weight from the axis of the specimen, which was immersed for about two-thirds of its length in the corroding solution. The time for which a specimen resisted a specific bending stress was determined, the failure being assessed by fracture of the specimen or by its yielding a certain fixed amount. Yielding was determined by measurement of the deflection of the lever arm at a distance of 12 in. from the top end of the specimen.

For direct tension stress-corrosion tests, a special apparatus was designed in which the specimen was totally immersed in the corroding medium with the exception of the upper end, which was connected to a 20 : 1 lever. The lower end of the specimen was fixed in plastic wedge grips (Perspex) to a support in similar material, which was in turn fixed to the frame of the apparatus. The normal 4-in. strip tensile test-piece was used for the tests.

The results of the tests on specimens stressed in bending showed that the unaged specimens were very resistant to stress-corrosion failure, with the exception of those made from alloy 4, which failed after less than 100 hr. by yielding. Most of the specimens of alloys 1, 2, and 3 did not fail after more than 500 hr., and it was clear that the stress-corrosion properties of the cold-worked material were appreciably superior to those of the material that had been given no cold work in the case of all three alloys. The results of the tests on these three alloys were very similar, and Fig. 4 (*a*), which shows graphically the results obtained on alloy 1, is typical. It should be noted that the points on the graph indicate the times at which a certain limit of deformation of the specimens had occurred corresponding to a drop of 100 mm. of the lever ends.

After ageing for 28 days at 70° C., the stress-corrosion properties of alloys 1, 2, and 3 decreased appreciably. In Fig. 4 (*b*) the results of tests on alloy 1 are shown, and these are typical of those obtained on alloys 2 and 3. The stress-corrosion properties of alloy 4 were scarcely affected by ageing at 70° C.

For alloys 1, 2, and 3 the stress-corrosion curves show that the lowest stress-corrosion-resistance occurred in aged specimens that had been given 30% cold reduction, and the highest stress-corrosion-resistance in specimens that had received 50% cold reduction. This was particularly noticeable in specimens stressed at high stresses.

The stress-corrosion tests in tension confirmed the bending stress-corrosion tests, but failure occurred more slowly in the direct tension test. This is to be expected, since, as the test-piece cracks, the increase in stress is proportional to the square of the reduction in thickness in the bend test and to the reduction of thickness in the direct tension test. It may be expected, therefore, that bending stresses in service are much more dangerous than tension stresses if they are associated with corrosion attack.

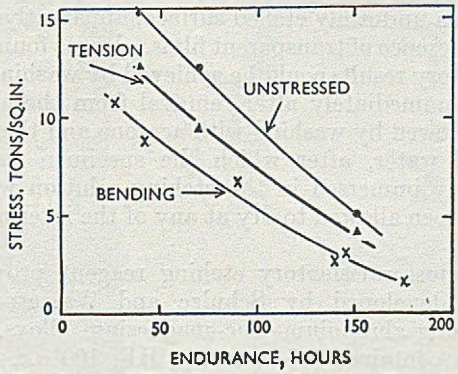


Fig. 5.—Bending and Tension Stress-Corrosion Tests on Alloy 1, Cold Rolled 30% and Aged at 70° C.

The results of bending and tension stress-corrosion tests on alloy 1 in the aged condition after 30% cold reduction are shown graphically in Fig. 5, together with the results of residual strength tests on unstressed specimens exposed to the same corroding conditions and for the same time as the tension stress-corrosion test-pieces. The same relative positions of the three curves were obtained for all degrees of cold work of alloys 1 to 4 in both the aged and unaged conditions. Micro-examination confirmed that in both bending and tension stress-corrosion tests, specimens that had been given 30% cold reduction suffered much more severe intercrystalline attack in alloys 1, 2, and 3 than specimens that had been less severely cold worked. After 50% cold reduction the attack occurred along slip planes as well as along grain boundaries, and since the boundaries were orientated parallel to the surface of the sheet, penetration across the section occurred less rapidly.

As regards the corrosion attack on alloy 4, microscopical examination showed that corrosion was in general of the pitting type and usually occurred in bands. This banded attack became less pronounced as the amount of cold work was increased, but even material that had been given 50% cold work was attacked much more rapidly than the alloys 1, 2, and 3.

III.—COMMENTS AND CONCLUSIONS

The results of the investigation have shown that additions of 0–1% zinc to aluminium alloys containing 5% magnesium have very little effect on the mechanical properties. The addition of up to 0·5% zinc has little effect on the corrosion properties, and the microstructure is only slightly affected, although after ageing at 70° C. there is a very slight increase in the amount of precipitate with increasing zinc content. Precipitation on ageing occurs at the grain boundaries and takes the form of an incomplete network. The alloy containing 1% zinc, however, shows pronounced precipitation after ageing at 70° C., and thus corrosion attack of the heat-treated material is increased.

The effect of cold work in general is to promote precipitation during ageing at 70° C. Two effects are evident in the microstructure as the amount of cold work is increased. There is an increased tendency to form a continuous network of precipitation at the grain boundaries with increasing amounts of cold work up to 30%, whilst precipitation along slip bands occurs at higher degrees of cold work.

The most interesting observation made in this investigation is, however, the effect of cold work on the corrosion-resistance of the alloys. Small amounts of cold work reduce the corrosion-resistance appreciably, but heavy cold work, both of aged and unaged material, results in a corrosion-resistance as high as that of material that has not been cold worked. The maximum corrosion attack at about 30% cold reduction can be correlated with continuous grain-boundary films formed on ageing. The increased corrosion-resistance at heavier reductions can probably be associated with a less localized attack due to the heavier and more general precipitation.

The reason for the corrosion behaviour of the unaged alloys is not very clear. It is known that cold work produces a surface in which there are couples at different electrolytic potentials, and it seems that smaller amounts of cold work, up to about 30%, cause increased local attack owing to non-uniformity of cold work. The effect of increasing the amount of cold work still further to 50%, however, results in a more uniformly cold-worked structure. The corrosion attack is therefore much more evenly distributed over the surface and the loss in tensile strength is smaller.

If this variation in degree of cold work were also associated with incipient precipitation of a second phase, the susceptibility to corrosion attack would be still further increased. No evidence for this assumption was found, however.

Comparable behaviour of the material has been found in stress-corrosion tests both in bending and in direct tension. The corroding medium used was very aggressive and bears no relation to the possible behaviour of the material in service conditions, though the results obtained indicate that ageing after 30% cold reduction produces a structure that is more susceptible to stress-corrosion attack than material that has had greater and smaller amounts of cold work.

ACKNOWLEDGEMENTS

This paper is published by kind permission of the Chief Scientist, Ministry of Supply. The authors would like to take this opportunity of thanking their colleagues of the Metallurgy Department of the Royal Aircraft Establishment, Farnborough, for their valuable assistance during the years 1945–47, when the laboratory work was completed. They are also indebted to Mr. E. A. G. Liddiard, M.A., of the Fulmer Research Institute, for helpful suggestions during preparation of the paper and for providing facilities for reproduction.

APPENDIX

Electropolishing and Etching Procedures

The method of electrolytic polishing adopted was a modification of that described by de Sy and Haemers[11] in which a perchloric acid–ethyl alcohol solution was used.

The electrolyte, which was placed in a 400-c.c. glass container, consisted of 200 c.c. of absolute alcohol and 40 c.c. of perchloric acid (75%). The cathode was a piece of aluminium sheet about 10 cm. long, bent to a radius of 5 cm., and immersed in the electrolyte to a depth of approximately 5 cm. The specimen formed the anode of the cell, and was placed about 5 cm. from the surface of the cathode. Current was supplied from the 230-V. D.C. mains, and the required voltage (12 V.) was obtained by means of a potentiometer.

Specimens of sheet, approximately 5 cm. long × 0·5 cm. wide × 0·1 cm. thick, were polished on the cross-sectional surface with emery papers, the last paper used being 00, and the final scratches running longitudinally.

The specimen was immersed in the electrolyte to a depth of about 4 cm. in a vertical position, with the polished surface facing the cathode. The portion of the specimen that projected above the surface of the liquid acted as a contact. Stirring of the electrolyte was not satisfactory, as the edges of the specimen tended to become rounded owing to removal of material at a greater rate in this region. It was found, however, that excellent results were obtained if the specimen was subjected to slight vibration, which was produced by lightly tapping the specimen holder. As a result of the vibration a high and steady current passed through the cell and the potential was maintained at 12 V. by suitable adjustment of the potentiometer. It was necessary to keep the temperature of the electrolyte below 30° C. to ensure satisfactory results.

Normal methods of washing and etching failed to produce a uniformly etched surface, apparently owing to the presence of transparent films. It was found that satisfactory results could be achieved by washing with alcohol immediately after removal from the electrolyte, followed by washing with acetone and then with distilled water, after which the specimen was immediately immersed in the etching solution without having been allowed to dry at any of the intermediate stages.

The most satisfactory etching reagent proved to be one developed by Schulze and Wassermann[12] for etching aluminium–zinc–magnesium alloys. The solution contained 0·5 c.c. 60% HF, 100 c.c. water, 1 c.c. conc. HCl, 10 c.c. conc. HNO_3, 50 c.c. of a 10% solution of $K_2Cr_2O_7$. This solution deteriorated fairly rapidly when stored in a glass bottle. In order to overcome this difficulty two solutions were made up as follows:

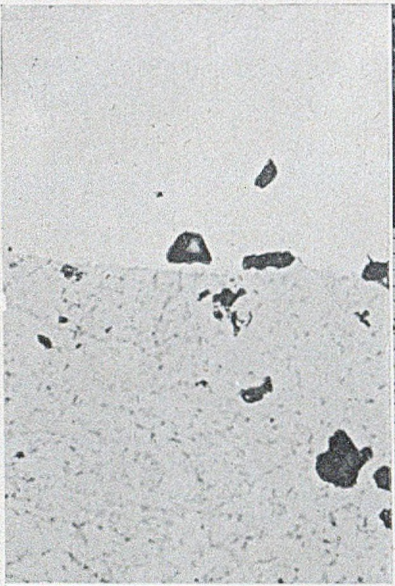
Solution A	Solution B
500 c.c. Water	250 c.c. of 10% solution of $K_2Cr_2O_7$
100 c.c. Conc. HNO_3	10 c.c. Conc. HCl
2·5 c.c. 60% HF	

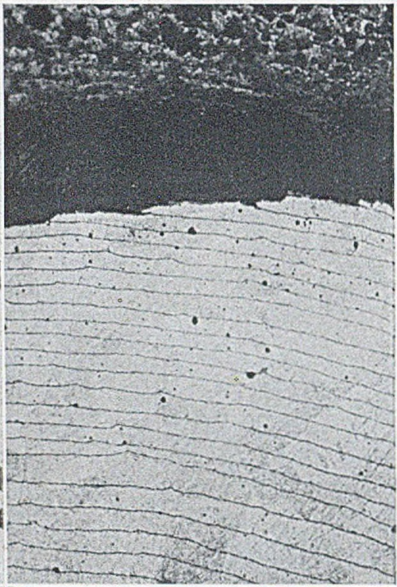
Solution A was stored in a suitable waxed bottle and was mixed with an equal volume of solution B as required. All the specimens examined were etched for 60 sec. in this reagent.

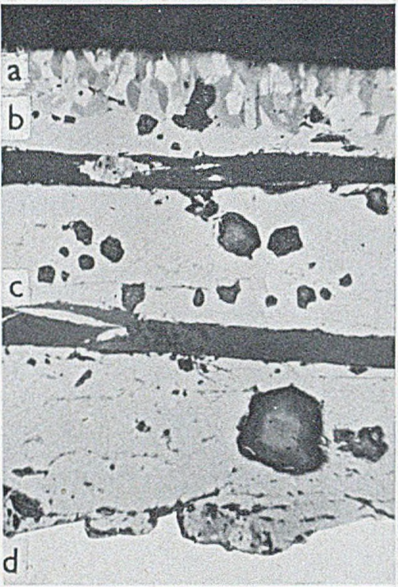
REFERENCES

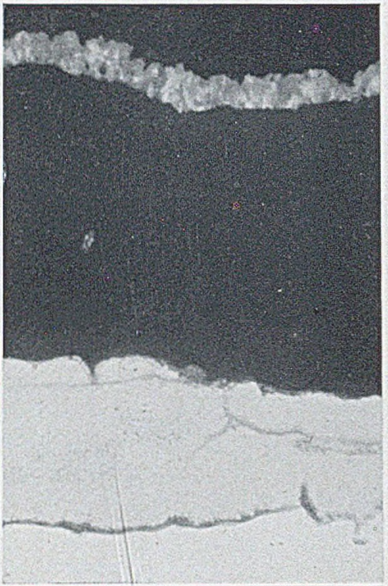
1. P. Brenner and W. Roth, *J. Inst. Metals*, 1948, **74**, 159.
2. E. C. W. Perryman and G. B. Brook, *ibid.*, 1951, **79**, 19.
3. E. C. W. Perryman and S. E. Hadden, *ibid.*, 1950, **77**, 207.
4. G. J. Metcalfe, *ibid.*, 1946, **72**, 487.
5. G. Siebel and H. Vosskühler, *Z. Metallkunde*, 1940, **32**, 298.
6. H. Vosskühler, *Werkstoffe u. Korrosion*, 1950, **1**, 143, 179, 310, 357.
7. W. L. Fink and L. A. Willey, *Trans. Amer. Inst. Min. Met. Eng.*, 1937, **124**, 78.
8. C. S. Barrett and L. H. Levenson, *ibid.*, 1940, **137**, 76.
9. P. Lacombe and L. Beaujard, *Compt. rend.*, 1944, **219**, 66.
10. P. J. E. Forsyth, R. King, G. J. Metcalfe, and B. Chalmers, *Nature*, 1946, **158**, 875.
11. A. de Sy and H. Haemers, *Rev. Mét.*, 1941, **38**, 122.
12. E. Schulze and G. Wassermann, *Z. Metallkunde*, 1940, **32**, 415.

11, 12



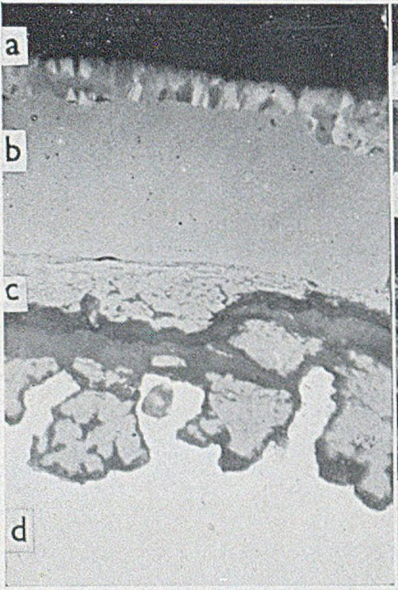


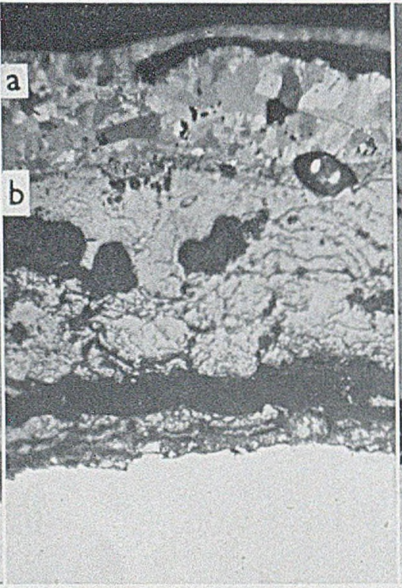


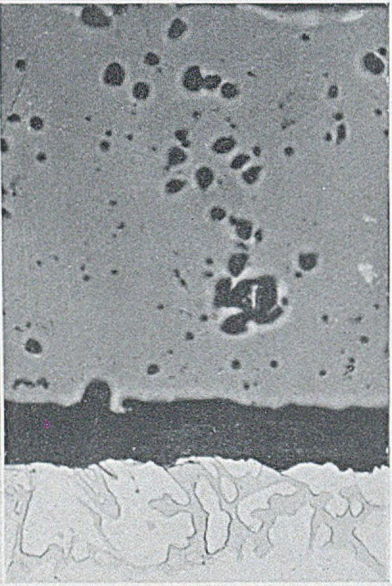


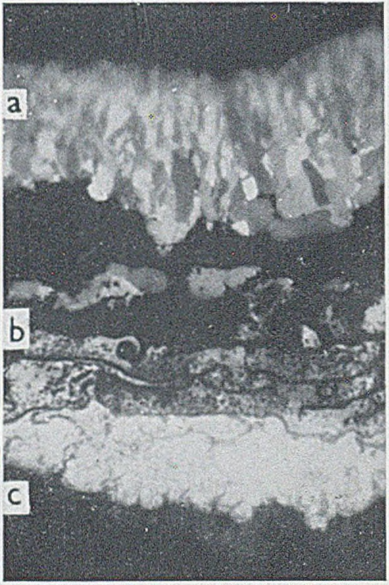
13, 14

15, 16









17, 18

Fig. 11.—Junction Between Duplex Scale and Cuprous Oxide on a 2% Al Alloy Oxidized at 1000° C. Unetched. × 500.

Fig. 12.—"Nessler Ring" Subscale on 1·2% Cr Alloy Oxidized at 1000° C. Etched in ammonia + hydrogen peroxide. × 30.

Fig. 15.—Scale on 2% Si Alloy Oxidized at 750° C. *a*—cupric oxide; *b*—cuprous oxide; *c*—cuprous oxide + silica; *d*—metal. Unetched. × 300.

Fig. 16.—Scale on 2% Si Alloy Oxidized at 800° C. *a*—cupric oxide; *b*—cuprous oxide + silica. Unetched. × 300.

Fig. 13.—Scale on 1% Mg Alloy Oxidized at 800° C. *a*—cupric oxide; *b*—cuprous oxide; *c*—cuprous oxide + magnesia; *d*—metal. Unetched. × 300.

Fig. 14.—Detached Cupric Oxide and Subscale on 2% Si Alloy Oxidized at 550° C. Unetched. × 500.

Fig. 17.—Cuprous Oxide and Appearance of Second Phase in Underlying Metal on 2% Si Alloy Oxidized at 950° C. Unetched. × 200.

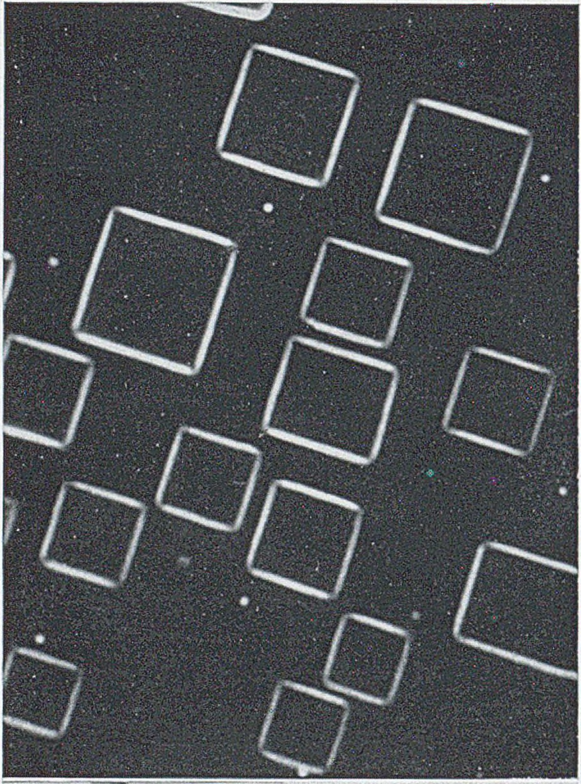
Fig. 18.—Detached Scale on 3·5% Si Alloy Oxidized at 650° C. *a*—cupric oxide; *b*—cupric oxide + silica; *c*—cuprous oxide + silica. Unetched. × 1000.

[To face p. 208

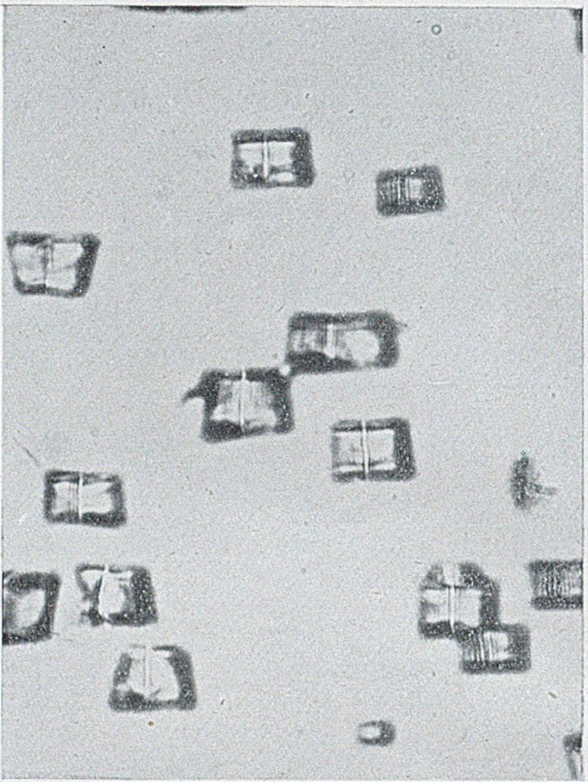
Plate XXXII.

ETCH-PITS IN SUPER-PURITY ALUMINIUM. × 750.

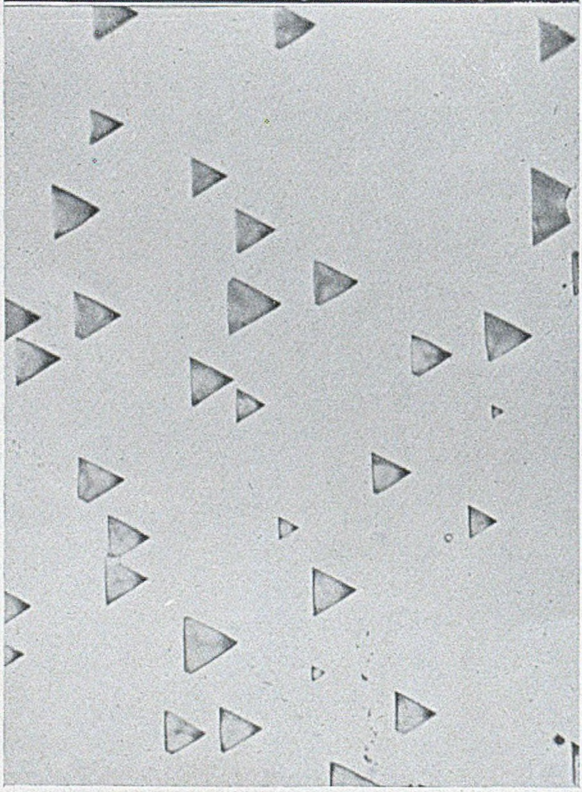
9



10



11



12

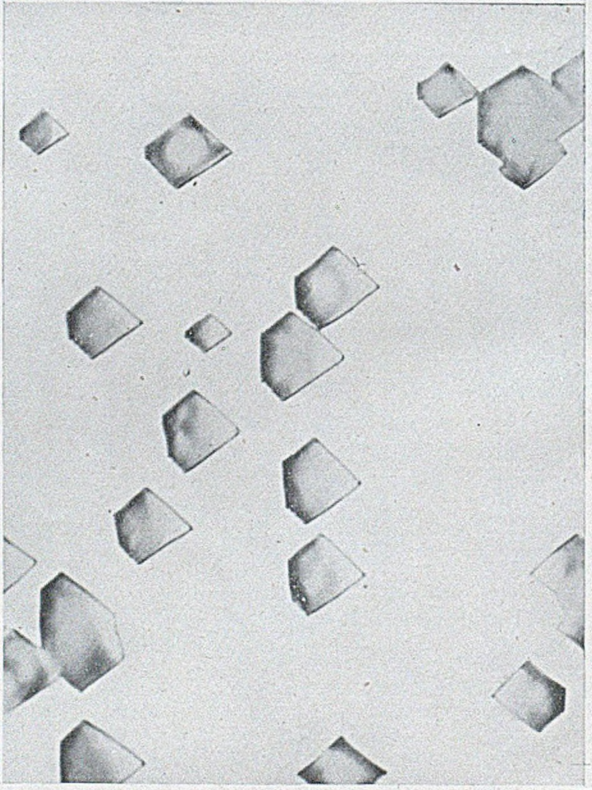
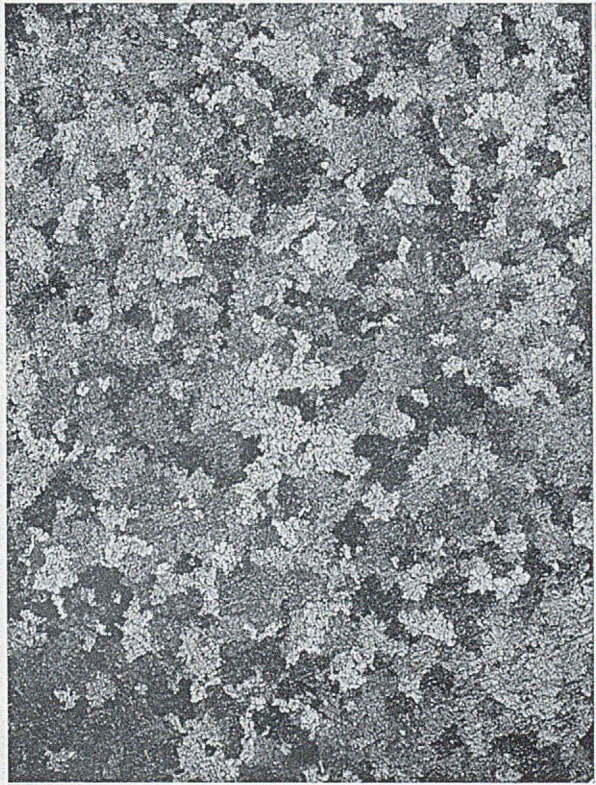
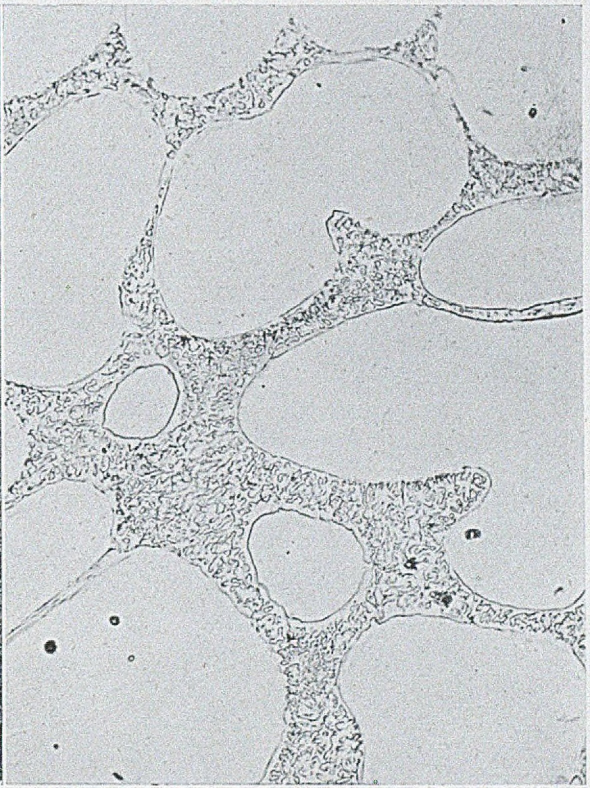


FIG. 9.—(001) Type. Dark-field illumination.
FIG. 11.—(111) Type.

FIG. 10.—(011) Type. Focused on bottom edges of pits.
FIG. 12.—Truncated Triangular Type.



7



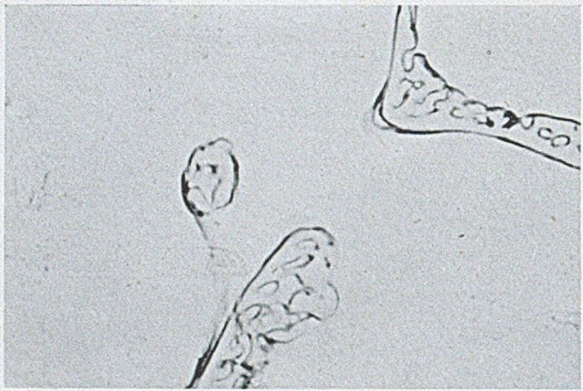
8

9

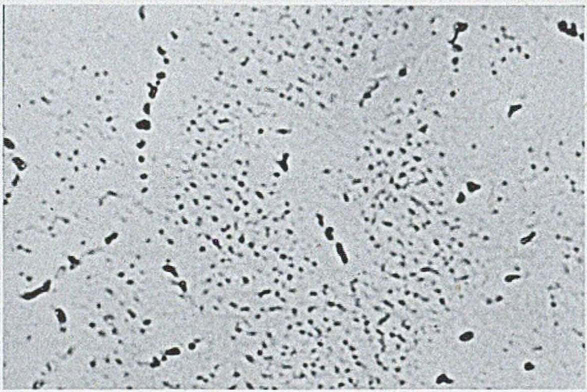
10

FIGS. 7 and 8.—Alloy *X*38 (5·95% Cerium), As Cast. × 10 (Fig. 7) and × 500 (Fig. 8).
FIG. 9.—Alloy *X*32 (5·44% Rare Earths), As Cast. × 500.
FIG. 10.—Alloy *X*109 (3·07% Rare Earths, 0·43% Zirconium), As Cast. × 750.

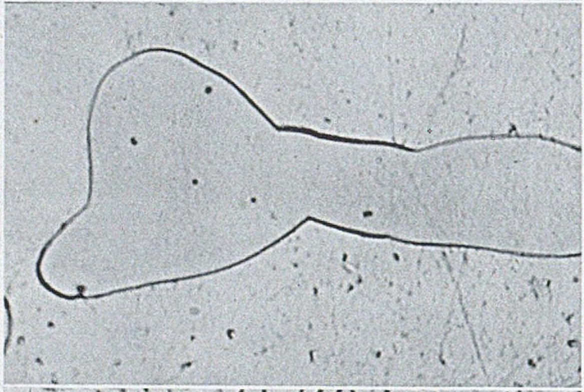
11



12



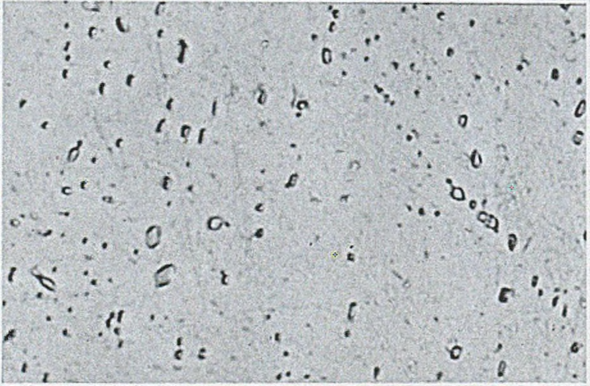
13



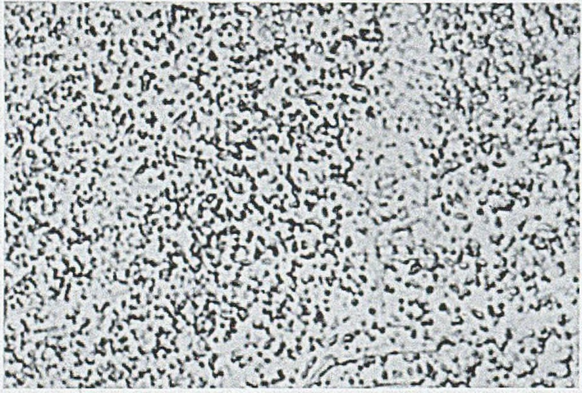
14

15

16



17



18

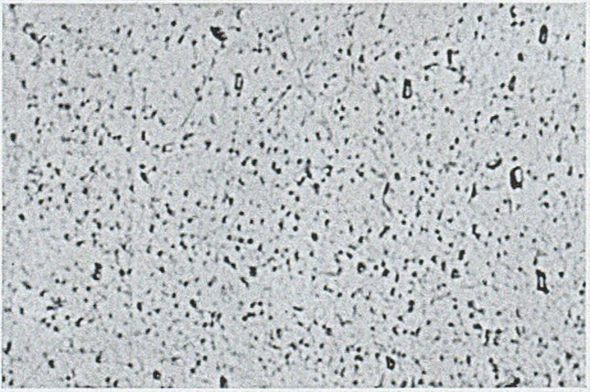


FIG. 11.—Alloy *X*91 (2·73% Cerium), As Cast. × 2000.
FIGS. 12–14.—Alloy *X*101 (2·7% Cerium, 1·4% Manganese).

Fig. 12.—As Cast. × 2000.
Fig. 13.—Cast, 24 Hr. at 580° C., Water-Quenched. × 2000.
Fig. 14.—As Fig. 13 and Aged 7 Days at 350° C. × 500.

FIGS. 15 and 16.—Alloy *X*91 (2·73% Cerium), Cast, 24 Hr. at 580° C., Water-Quenched. Aged 7 Days at 300° C. (Fig. 15) or at 350° C. (Fig. 16). × 2000.

FIGS. 17 and 18.—Alloy *X*101 (2·7% Cerium, 1·4% Manganese), Cast, 24 Hr. at 580° C., Water-Quenched. Aged 7 Days at 300° C. (Fig. 17) or at 350° C. (Fig. 18). × 2000.

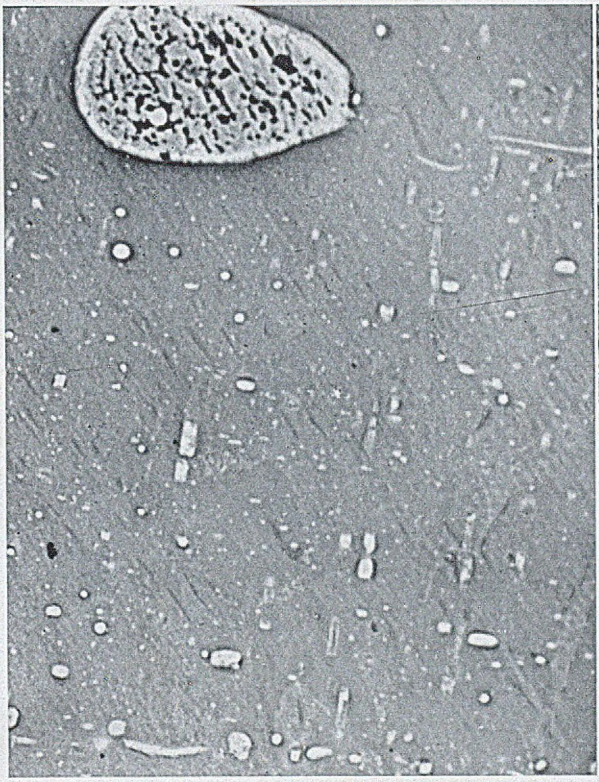


FIG. 19.—Electron Micrograph of Same Alloy as Fig. 16. × 2500.

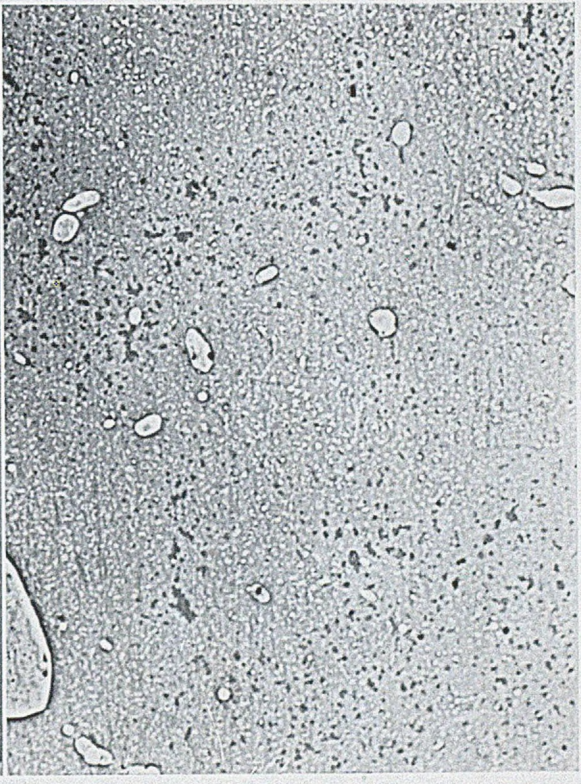


FIG. 20.—Electron Micrograph of Same Alloy as Fig. 18. × 2500.

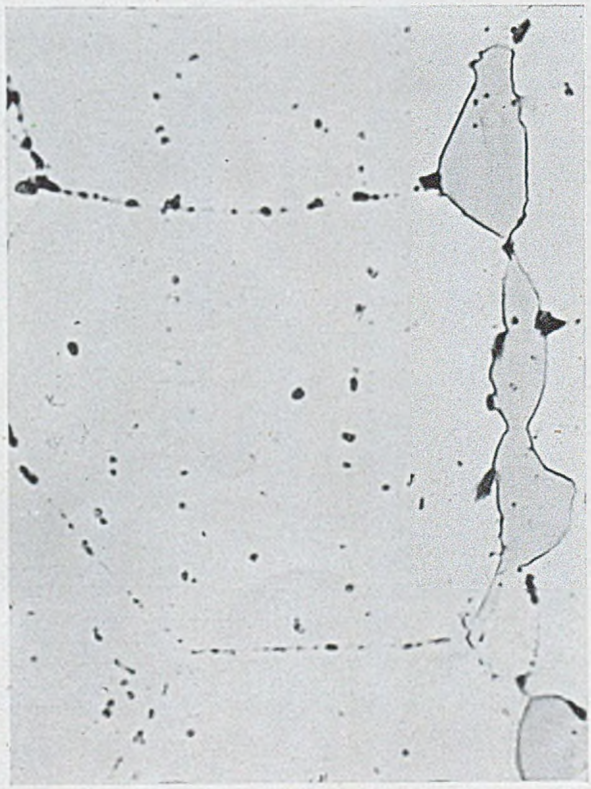


FIG. 21.—Alloy *X*64 (6·14% Rare Earths, 1·54% Manganese), Rolled, 24 Hr. at 580° C., Water-Quenched. Tested at 300° C., 1 ton/in.^{2} for 3660 hr. Centre of parallel portion. × 2000.

FIG. 4.—Pb–0·85% Sb Alloy Air-Cooled after Extrusion at 300° C. Slight discontinuous precipitation at grain boundaries, although no continuous precipitation within the grains. × 300.

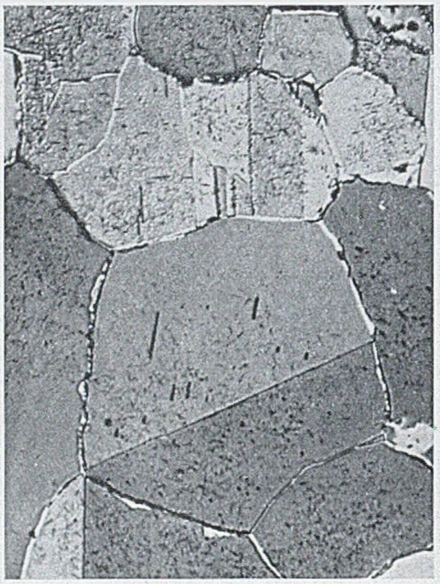


FIG. 5.—Pb–0·85% Sb Alloy Air-Cooled after Extrusion at 160° C. and Aged at 20° C. for 12,000 Hr. Continuous precipitation within grains. × 300.

FIG. 6.—Pb–0·85% Sb Alloy Solution-Treated at 300° C., Water-Quenched and Aged 120 Hr. at 50° C. Slight continuous precipitation within grains; mainly in form of widely distributed needles. × 300.

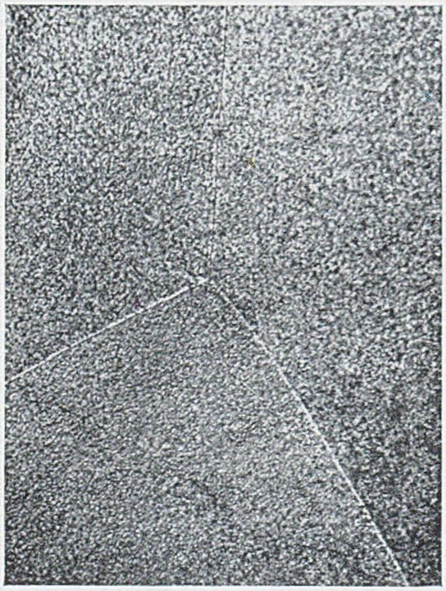
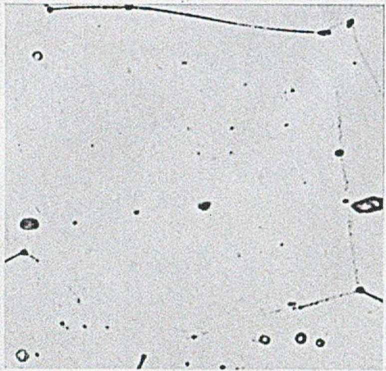
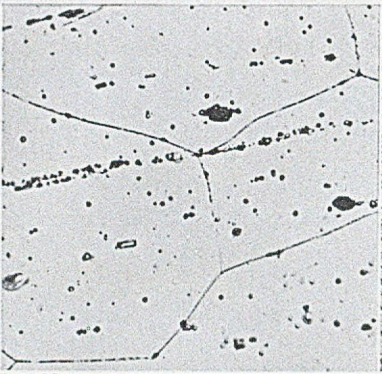


FIG. 7.—Pb–0·85% Sb–0·001% As Ternary Alloy Solution-Treated at 300° C., Water-Quenched and Aged 120 Hr. at 50° C. Extensive continuous precipitation within grains; precipitate is globular. × 300.

6

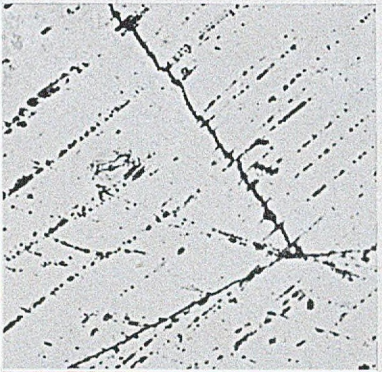


7

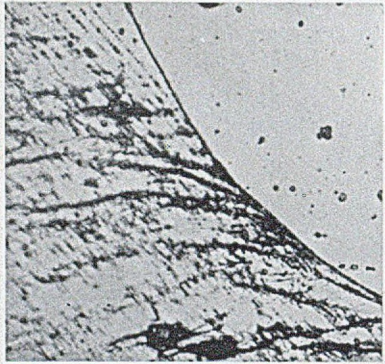


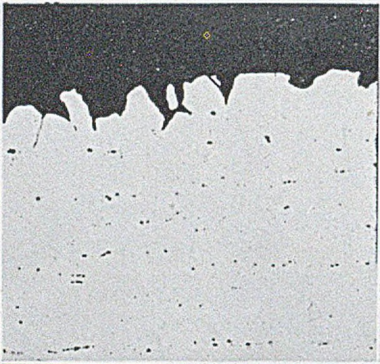
8

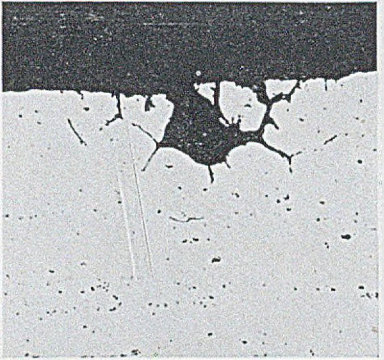
9



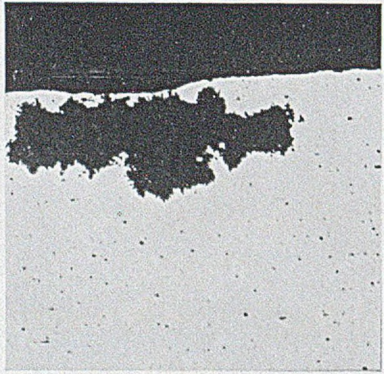
10, 11



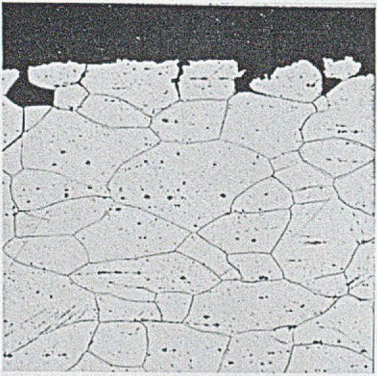




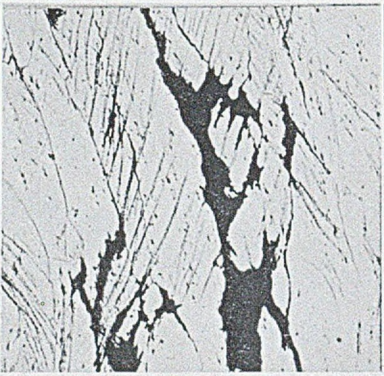
12, 13



14



15



16

17

FIG. 6.—Alloy 1, Aged at 70° C., Showing Incomplete Grain-Boundary Network of Precipitate. × 1000.

FIG. 7.—Alloy 1, Cold Rolled 20% and Aged at 70° C., Showing an Almost Complete Grain-Boundary Network of Precipitate. × 500.

FIG. 10.—Alloy 4, Cold Rolled 30% and Aged at 70° C., Showing Precipitation and Distortion of Slip Planes near Grain Boundary. × 1000.

FIG. 11.—Alloy 1, Not Aged, Exposed 12 Months, Showing Tendency to Intercrystalline Corrosion. × 100.

FIG. 14.—Alloy 4, Cold Rolled 5% and Aged at 70° C., Showing Severe Local Pitting. × 100.

FIG. 15.—Alloy 1, Cold Rolled 30% and Aged at 70° C., Etched, Showing Intercrystalline Corrosion. × 100.

FIG. 8.—Alloy 4, Cold Rolled 50% and Aged at 70° C., Showing Heavy Precipitation Within Grains. × 100.

FIG. 9.—Alloy 4, Cold Rolled 30% and Aged at 70° C., Showing Distortion of Grain Boundary Where Slip Planes Cross It. × 1000.

FIG. 12.—Alloy 4, Not Aged, Exposed 12 Months, Showing Heavy Pitting. × 100.

FIG. 13.—Alloy 1, Aged at 70° C., Exposed 12 Months, Showing Intercrystalline Corrosion. × 100.

FIG. 16.—Alloy 1, Cold Rolled 50% and Aged at 70° C., Showing Corrosion Attack Along Slip Planes. × 100.

FIG. 17.—Alloy 4, Cold Rolled 50% and Aged at 70° C., Showing Local Pitting With Slight Tendency to Intercrystalline Corrosion. × 100.

PLATE XXXVIII.

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 20 JANUARY 1953 PART 5

1 — PROPERTIES OF METALS

***Thermal Expansion of Aluminium and Some Aluminium Alloys.** (Hidnert and Krider). See col. 331.

***The Absolute Determination of the Condensation Factors of Molecular Jets of Antimony on a Surface.** Marcel Devienne (*Compt. rend.*, 1952, 234, (1), 80–81).—By the use of radioactive Sb, it has been found that the condensation factor for Sb deposited on a Cu surface is 0·74 for a layer 40 Å. thick and 0·55 for a layer 14 Å. thick.—N. B. V.

***Thermal Dependence of Elastic Constants of Electrodeposited Chromium.** H. Pursey (*Nature*, 1952, 169, (4291), 150).—An investigation has been made of the transition near 37° C. in the thermal dependence of the elastic const. of electrodeposited Cr recently reported by Fine, Greiner, and Ellis (*Trans. Amer. Inst. Min. Met. Eng.*, 1951, **191**, 56 ; *M.A.*, 19, 2). Variations of Young's modulus (E) and the modulus of rigidity (G) with temp. between 10° and 80° C. were studied, using a resonance method, on a tube of electroformed Cr annealed at 544° C. At the lower frequencies in the range 15–160 kc./s. the results are similar to those of F., G., and E., who used a tube annealed at 1000° C. ; at higher frequencies in this range the magnitude of the fluctuation is reduced, although the min. in the E/temp. curve remains const. at 36·5° C. The G/temp. curve shows a stationary portion between 36·5° and 42·5° C. and is independent of frequency. The results correspond to a process which varies the bulk modulus as a single-valued function of temp. with a time-const. of ~10 microsec.—R. S. B.

***A Study of the Annealing Kinetics in Cold-Worked Copper.** Dwain Bowen, R. R. Eggleston, and R. N. Kropschot (*J. Appl. Physics*, 1952, 23, (6), 630–635).—Samples of pure Cu wire cold-drawn to 97% reduction in area were pulse-annealed for successive periods at each of a number of temp. in the range 100°–250° C. Measurements of resistance were made at temp. in the range 4°–30° K. Matthiesen's rule was obeyed in that the abs. change of resistance as measured in a given annealed specimen was independent of temp. of measurement up to ~20° K. This means that lattice distortions due to cold work change the scattering cross-section for electron waves but not the number of carriers, and hence the residual resistance due to cold work ∝ concentration of lattice distortions. On this argument, and applying rate theory, the order of the annealing reaction was determined by finding which order was most nearly consistent with all the results. A fourth-order reaction is strongly indicated, with an activation energy of 28·3 kg.cal./mole. This implies that four lattice distortions co-operate in each unit process of recovery. No mechanism for this is suggested. It is assumed throughout that recovery of resistance is a single-stage process.—R. W. C.

***The Thermal and Electrical Conductivity of Copper at Low Temperatures.** R. Berman and D. K. C. MacDonald (*Proc. Roy. Soc.*, 1952, [A], **211**, (1104), 122–128).—The thermal and elect. conductivities—κ and σ, resp.—of high-purity Cu (Ag ~0·0005, Ni <0·0002, and Pb <0·0004%), annealed in He for 6 hr. at 450° C., have been determined continuously from 90° to 2° K. The results show a pronounced min. in the elect. resistance at ~10° K., and a similar but smaller anomaly in the thermal resistivity at about the same temp., with a corresponding small deviation in the Wiedemann–Franz law at the lowest temp. At temp. (T) between ~12° and 30° K., κ (in W./cm./°K.) follows the law $1/\kappa = 0{\cdot}212/T + 2{\cdot}5_5 \times 10^{-5}T^2$, while σ (in Ω-cm.$\times 10^{-6}$) is well represented by the equation $1/\sigma = (1/189{\cdot}6) + 2{\cdot}64 \times 10^{-10}\,T^5$. As in the case of Na (*ibid.*, 1951, [A], **209**, 368 ; *M.A.*, **20**, 137) marked disagreement with theory was found in the temp. variation of: (1) κ—no evidence being found of the predicted min. at ~80° K. (T/Θ ~0·25) and (2) the Lorenz number, $\kappa/\sigma T$. 12 ref.—E. N.

Temperature/Resistance Chart for Electrical Copper. A. E. Maine (*Electronic Eng.*, 1952, 24, (289), 111).—A chart is constructed for determining rapidly the change of resistance with temp. of a unit wound with Cu wire; alternatively, the new temp. may be found from a knowledge of the original temp. and the ratio of the resistance change.—H. Pl.

***Study of an Old [Copper] Bus-Bar.** A. Romwalter and F. Macher (*Acta Techn. Acad. Sci. Hungar.*, 1952, 3, (3/4), 355–357).—[In German]. Comparison of the spectra and metallographic structure of two samples (*a*) and (*b*) of a Cu bus-bar, forming part of the D.C. supply system of the Hungarian town of Sopron, through one part of which, (*a*), D.C. had passed for 30 years while no current had passed through the other part, (*b*), showed that (*a*) had suffered no structural or chem. change during the long period of passage of D.C. through it.—J. S. G. T.

***Observations on the Bauschinger Effect in Copper and Brass.** H. Schwartzbart, M. H. Jones, and W. F. Brown, Jr. ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Research Memo.*, 1951, (E51D13), 37 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 398).—For any tensile prestrain between 0·905 and 0·65, the stress at 0·001 plastic strain in subsequent compression is ~0·7 of that at the same scalar strain for simple compression. The stress subsequently rises at a rate that is slower for a larger prestrain and levels off ∥ the compression curve at ~ 0·01 strain in compression. The subsequent compression stress/strain curve levels off ∥ and below the virgin compression curve by more and more as the prestrain increases. For large tensile prestrains, there is an appearance of a Y.P. in subsequent compression. Stress-relief annealing caused the tension and compression curves following tensile prestrain and heating to approach each other in Cu, indicating relief of residual stress, but both curves were raised in brass, probably owing to strain-ageing. Longitudinally pre-stretched material had different tensile curves in the longitudinal and transverse directions. The difference was similar for isotropic and for cubically aligned sheet. This effect was related to the Bauschinger effect, according to previous work. Since the cubically aligned sheet presumably does not develop residual stresses

---

* Denotes a paper describing the results of original research. † Denotes a first-class critical review.

between grains to nearly the same extent as isotropic material, apparently the Bauschinger effect is not due to residual stresses caused by orientation differences between different grains.

Anomalous [Changes with] Temperature of Copper, and Their Physical Significance.—II(3). Mitsuru Satô (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 661–686).—[In English]. Cf. *Sci. Rep. Tôhoku Imp. Univ.*, 1939, [i], **27**, 278; *M.A.*, **6**, 205. In literature relating to Cu, anomalous changes in its properties have been reported at 503°, 553°, 823°, and 1356° K., and it is suggested that these temp., together with abs. zero (0° K.), are related, in some way, to components of the fine structures of the energy levels E_1 and E_2. Values of the sp. heat and elect. resistance, diffusion and sintering, annealing and hardness, and the phys. significance of these anomalous temp. are briefly discussed on this basis. —J. S. G. T.

The ABC's of Germanium. J. P. Jordan (*Elect. Eng.*, 1952, **71**, (7), 619–625).—The lattice structure and bonding of Ge are discussed and illustrated with ref. to rectification and semiconducting appn. Other properties of Ge are briefly reviewed. —J. R.

Effect of Various Gases on Potassium-Ion Emission from Hot Platinum. Charles F. Robinson. P. D. Zemany and E. L. Brady (*J. Chem. Physics*, 1952, **20**, (8), 1329).—Two letters. Z. and B. reply to a criticism by R. of their recent paper (*ibid.*, (2), 294; *M.A.*, **20**, 4).—J. R.

***The Viscosities [and the Debye Characteristic Temperatures] of Liquid Lithium, Rubidium, and Cæsium.** E. N. da C. Andrade and E. R. Dobbs (*Proc. Roy. Soc.*, 1952, [A], **211**, (1104), 12–30).—The oscillating-sphere method has been used to determine the viscosity (η) of molten Li (~99·8%), Rb (>99·5%), and Cs (>99·5%), as a function of temp., under a vacuum of $<10^{-4}$ mm. Hg. The construction and operation of the viscometer is described in detail; for Li it was necessary to construct a special Ni-lined metal vessel, since liq. Li attacks silicates, such as glasses of all kinds, and alloys with many metals other than Ni. The viscosity values (in millipoises) were found to be: (1) Li—6·02, 5·918, 5·749, 5·541, 5·406, 4·917, and 4·548, at 180° (the m.p.), 183·4°, 193·2°, 208·1°, 216·0°, 250·8°, and 285·5° C., resp.; (2) Rb—6·74, 6·734, 6·713, 6·663, 6·441, 6·258, 4·844, 4·133, 3·634, and 3·234, at 38·5° (the m.p.), 38·0°, 39·0°, 40·0°, 45·0°, 50·0°, 99·7°, 140·5°, 179·0°, and 220·1° C., resp.; and (3) Cs—6·86, 6·837, 6·299, 4·753, 4·065, 3·750, and 3·343, at 28·2° (the m.p.), 28·4°, 43·4°, 99·6°, 140·5°, 168·0°, and 210·9° C., resp. The results show that: (i) the abs. values of the m.p. viscosities, η_M, accord well (generally to within ±0·3%) with the previously published formula, $\eta_M = \kappa\,(AT_M)^{\frac{1}{2}}V^{-\frac{2}{3}}$, where κ is a derived const. (of dimensions $M^{\frac{1}{2}}LT^{-1}\theta^{-\frac{1}{2}}$) having a value of $\sim 5{\cdot}4 \times 10^{-4}$ C.G.S. units, A the at. wt., V the at. vol., and T_M the m.p. in °K.; and (ii) the variations of η with temp. (T° K.) fit the simple equation, $\eta = Be^{b/T}$, and, more closely, the more elaborate equation, $\eta v^{\frac{1}{3}} = Ae^{c/vT}$, where v is the sp. vol., and A, B, b, and c are const. peculiar to the metal in question. The values for the characteristic Debye temp., Θ, of Li, Na, and K, which can be derived from the m.p. viscosities, agree well with the values derived from the sp. heats of these metals. The value of Θ for Rb and Cs, calculated from their η_M values are 75° and 53° K., resp. Values of the kinematic viscosity, ν_M, at the m.p. are derived. 16 ref.—E. N.

***Fluidity and Temperature. II.—Cases of Mercury and Gallium and a General Critique.** Asoke Kumar Mukherjee (*J. Indian Chem. Soc.*, 1951, **28**, (11), 626–630).—[I. is of no met. interest.] An equation connecting fluidity and temp. of Hg and Ga has been tested, and the results are interpreted. —W. A. M. P.

An Introduction to Arc-Cast Molybdenum and Its Alloys. J. L. Ham (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (8), 723–731; discussion, 731–732).—Investigations are being made for the appn. of Mo alloys in piercing plugs for seamless steel tubes; gas-turbine blades; electrodes for heating molten glass; die-casting dies for brass and other non-ferrous metals; certain components of turbo jets, ram jets, rockets, and nuclear reactors; and parts exposed to corrosive chemicals. The general properties of unalloyed cast Mo are described, and curves are presented to show the mech. properties at room temp., the solid-solubility limits of Si, Al, Fe, Co, and Ni in Mo at various temp., and the effect of Be, Al, Si, Ti, V, Cr, Fe, Co, Ni, Zr, Nb, Ta, and W on the hardness of Mo at various temp. Mo contg. a little Be can be hardened by quenching and softened by slow cooling. Regardless of which alloying element is used there appears to be a limit to the increase in hardness that can be tolerated without causing difficulty in hot working by the techniques so far applied. The effects of several of the alloying elements on the rate of work-hardening and on the temp. of softening and recrystn. were studied, using a series of alloys which were extruded.—H. Pl.

***Effect of Temperature on the Tensile Properties of High-Purity Nickel.** William D. Jenkins and Thomas G. Digges (*J. Research Nat. Bur. Stand.*, 1952, **48**, (4), 313–321).—99·85% Ni was tested in tension at temp. from −320° to +1500° F. (−196° to +815° C.) After fracture the specimens were examined under the microscope. Both the yield and tensile strengths decreased rapidly with increase in temp, except in the range 80°–300° F. (27°–150° C.) where the yield strength increased and the tensile strength decreased slowly. The ductility at max. load, measured by plastic extension and reduction of area, decreased with temp., although irregularities were found in the temp. ranges 80°–300° F. and 500°–700° F. (260°–370° C.). The reduction of area at complete fracture increased with temp., and the elongation had a min. value at ~500° F. Strain ageing occurred in the range 80°–300° F., and recrystn. and recovery occurred above 1200° F. (650° C.). —J. W. C.

***Physical Properties of Electrodeposited Metals. I.—Nickel: (3) The Effect of Plating Variables on the Structure and Properties of Electrodeposited Nickel.** A. Brenner, Victor Zentner, and C. W. Jennings (*Plating*, 1952, **39**, (8), 865–894, 899–927, 933).—Cf. *ibid.*, 1948, **35**, 1228; *M.A.*, **16**, 496. A report on work carried out under Research Project No. 9 of the American Electroplaters' Society. Full details are presented of the results of an extensive series of tests to determine the effect of plating variables on the compn. of deposits, the structure of deposits, internal stress, *d*, and the mech., elect., and magnetic properties of deposits. Information is included on the heat-treatment of Ni deposits. It is concluded that the most important factor that determines the properties of deposits is the type of soln. Variations in operating conditions have a relatively minor effect, provided the conditions are within the satisfactory operating range of the soln. The effect of bath compn. and operating variables upon properties, as observed in this work, is in essential agreement with data in the literature. 209 ref.—G. T. C.

***Effects of Cold Working and Heat-Treatment on the Magnetic Properties of Nickel Sheets.** Tadao Fukuroi and Kichirô Yasuhara (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 687–697).—[In English]. The magnetic properties of Ni sheets, of thickness 50 μ, were measured in directions ∥ and ⊥ the direction of cold rolling and after heat-treatments below and above the recrystn. temp., after being cold rolled to various degrees of reduction of thickness up to 77%. Anisotropies were found in all the magnetic properties measured, viz. magnetization curves, initial (χ_a) and max. ($\chi_{\max.}$) susceptibilities, coercive force, H_c, and residual magnetization. Annealing at a temp. below that of recrystn. increases the value of χ and its anistropy exhibited by the cold-rolled specimen. Variations of χ and H_c are discussed in terms of fibre structure and internal stress.—J. S. G. T.

***Activated Adsorption of Hydrogen on Nickel Catalysts.** G. C. A. Schuit and N. H. de Boer (*Nature*, 1951, **168**, (4285), 1040–1041).—The all-important parameter in studying the adsorption of H on Ni–Si catalysts was found to be the temp. of reduction. Above 500° C., when reduction is substantially complete, the adsorption was immeasurably rapid. Below 400° C., when reduction is incomplete, quite a considerable amount of H is adsorbed slowly. O adsorbed at −196° C. was found to diminish the amount of rapidly adsorbed H, while the total adsorption remained const. S. and de B. conclude that the results support the hypothesis of Roberts

(*Proc. Roy. Soc.*, 1935, [A], **152**, 445, 464, 477 ; *M.A.*, **2**, 560), that the O content of the metal is responsible for the activated adsorption. The phenomenon cannot be considered as evidence for the heterogeneity of a pure metal surface.—R. S. B.

***Effect of a Magnetic Field on the Propagation of Sound Waves in a Ferromagnetic Material. [Nickel].** J. de Klerk (*Nature*, 1951, **168**, (4283), 963–964).—de K., using Ni and Ni–Fe rods, has confirmed the observations briefly noted by Rogers and Johnson (*J. Appl. Physics*, 1950, **21**, 1067 ; *M.A.*, **18**, 644) that the attenuation of H.F. longitudinal acoustic waves decreases as the applied field increases. Both longitudinal and transverse waves give the similar result that the attenuation (μ) decreases to a min. value as the magnetization approaches saturation. A plot of $(\mu_0 - \mu_H)/H$ is strikingly similar to the I/H initial curves of magnetization of a ferromagnetic material.—R. S. B.

***Self-Diffusion in Solid Sodium.—I. II.—The Effect of Pressure.** (I) Norman H. Nachtrieb, Edward Catalano, and John A. Weil. (II) N. H. Nachtrieb, J. A. Weil, E. Catalano, and A. W. Lawson (*J. Chem. Physics.*, 1952, **20**, (8), 1185–1188 ; 1189–1194).—[I.—] Tabulated data from a radioisotope study of self-diffusion in solid Na over the range 0°–95° C. fit the equation $D = 0{\cdot}242 \exp(-10{,}450/RT)$, with an estimated uncertainty of ±300 cal./mole in the heat of activation. The entropy of activation calculated from the temp. coeff. of the elastic modulus is in agreement with the value based on the vacancy mechanism. [II.—] For solid Na under hydrostatic pressure the rate of self-diffusion is decreased, and the free energy of activation is a linear function of the m.p. ; the enthalpy of activation (ΔH) $\propto$ the m.p. and is related to the latent heat of fusion (L_m) by the equation $\Delta H = 16{\cdot}5L_m$, which is in agreement with available diffusion data for all the cubic metals except Pb and W. The change in D_0 for Na with pressure is expressible in terms of the compressibility and the activation entropy change.—J. R.

Diffusion Coefficient of Sodium Vapour in Nitrogen. R. J. Cvetanović and D. J. Le Roy (*J. Chem. Physics*, 1952, **20**, (2), 343–344).—A letter. Accuracy in the diffusion coeff. of Na vapour is of fundamental importance in rate-const. detn. by the diffusion flame method, yet the present value employed relates to a temp. higher than that normally used, and is admittedly uncertain. Detn. by C. and Le R. of the diffusion of Na vapour into N gave the value 0·68 ± 0·03 cm.^{2}/sec. at 253·5° C. and 1 atm., which indicates that employment of the previous value, suitably corrected for temp., has occasioned no serious error.—J. R.

***[Electrical] Resistivity of Evaporated Tellurium Films.** Takemaro Sakurai and Seiji Munesue (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 96–104).—[In English]. The elect. resistivity (R) of Te films, made by evaporation *in vacuo* and deposited upon a Si substrate (S) was measured at various temp. (T). When S was kept at room temp. during evaporation, log R plotted against $1/T$ gave a curve conspicuously different from that characterizing the metal in bulk. By heat-treatment or electron bombardment of the metal, R changed irreversibly, becoming closer to the value for the metal in bulk. This phenomenon is attributed to crystn. of an amorphous deposit in the film. When the Te was deposited on a hot substrate, at 470° K., R agreed well with the value for the bulk metal. Values of the activation energies in the intrinsic and extrinsic range, resp., are 0·34 and 0·04 eV., in good agreement with the resp. values, 0·36 and 0·039 eV., found by Scanlon and Lark-Horovitz (*Phys. Rev.*, 1948, [ii], **73**, 1256 ; *M.A.*, **17**, 5). A mechanism of the phenomenon is suggested and discussed.—J. S. G. T.

***Heats of Solution of Group IB Metals [Gold, Silver, and Copper] in Liquid Tin.** L. B. Ticknor and M. B. Bever (*J. Metals*, 1952, **4**, (9), 941–945).—The construction and operation of an isothermal calorimeter suitable for use with liq. metals in the temp. range 150°–320° C., and under a vacuum, is described and illustrated. It has been used for detn. of the integral heats of soln. (ΔH, in cal./g.-atom of solute, with a positive sign if heat is absorbed) of Group IB metals in liq. Sn (<0·003% impurities) as solvent. The results are : (1) Au (99·95%)— −5980 and −5610 at 240° and 300° C., resp. ; (2) Ag (99·92%)— +3800 and +3900 at 240° and 300° C., resp. ; (3) Cu (high purity)— +2570 at 300° C., and (4) a 75 : 25 Au–Ag alloy— −1320 at 240° C. These values lead to the following estimates for the partial molar heat of mixing at 300° C. : −8400, +1250, and −50 cal./g.-atom of Au, Ag, and Cu, resp. ; the negative value for Au is consistent with the Au–Sn phase diagram, which shows that the system has a strong tendency towards compound formation in the solid state ; the negligible value for Cu points to ideal solid-soln. behaviour. The heat of formation, at 0° C., of the Au–Ag alloy, determined as −1010 cal./g.-atom, compares favourably with the value of −890 cal./g.-atom at 500° C. found by other workers. 13 ref.—E. N.

Titanium and Its Alloys. A. E. Williams (*Min. J.*, 1950, **235**, (5994), 10–13).—A brief survey of the properties, sources, and uses of Ti and the properties and uses of Ti–Al alloys is given.—S. R. W.

New Structural Metals [Titanium, Zirconium, and Molybdenum]. J. J. Harwood (*Product Eng.*, 1950, **21**, (10), 96–102).—The properties and methods of processing and fabrication of Ti, Zr, and Mo are described, and the development of alloys based on them is briefly reviewed. 9 ref.—M. A. H.

***Observations on the Preparation of Iodide Titanium.** O. J. C. Runnals and L. M. Pidgeon (*J. Metals*, 1952, **4**, (8), 843–847).—A study has been made of the kinetics of the van Arkel–de Boer system (hot-wire process) for the prodn. of "iodide" Ti, to determine the relative influence of the fundamental mechanisms, viz. prodn. of the Ti iodides, gaseous counter-current diffusion, and decompn., on the rate of diffusion. This has been found to be : (1) extremely sensitive to the geometry of the system, particularly to the position of the crude metal in the finely divided charge, and (2) influenced by the diffusion of I from the wire to the crude metal. The non-volatile compound TiI_3 was found to be condensed on the cool W lead-in wires directly above the hot (1500° C.) Ti filament ; it must have originated at, or in the vicinity of the hot filament itself, and probably resulted from either the heterogeneous partial dissociation of TiI_4 and/or a homogeneous gas-phase reaction of monatomic I with TiI_4. 10 ref.
—E. N.

***Note on the Effect of Oxygen and Nitrogen on the Hardness and Lattice Parameter of High-Purity Vanadium.** S. Beatty (*J. Metals*, 1952, **4**, (9), 987–988).—Four samples of V were prepared by : (A) the iodide method, (B) Ca redn. of V_2O_5, at temp. sufficient to give a melt, (C) as (B) but remelted under pure A, and (D) Ca redn. of V_2O_5 to give a powder which was then compacted by powder-metallurgy methods, and finally annealed in a high vacuum at 800° C. The samples contained ~0·1% metallic impurities ; the C content was assumed low, and the principal impurity difference between the samples was in the O and N contents. When the hardness (in kg./mm.^{2}) and the lattice parameters (a_0 in Å. at 30° C.) of the specimens were determined and related to the O and N contents (in wt.-%) the following results were found for these factors, resp., for samples A, B, C, and D, resp. : 65 ± 10, 3·028, 0·038, 0·021 ; 113 ± 5, 3·031, 0·044, 0·032 ; 185 ± 5, 3·034, 0·075, 0·106 ; 343 ± 10, 3·0375, 0·15, 0·189. A marked increase in line broadening of the back-reflection Laue patterns was observed as the impurities increased, and is attributed to non-uniformity of compn. within the specimens rather than to a tetragonal modification of the b.c.c. lattice. It is concluded that small addn. of O and N result in a sharp increase in the lattice parameter and hardness of V, an effect similar to that occurring in Ti. 8 ref.—E. N.

***The Internal Friction of Zinc Crystals.** Genzirō Mima and Masasuke Imoto (*Technol. Rep. Osaka Univ.*, 1952, **2**, (33), 93–100).—[In English]. The internal friction of Zn crystals was found to be increased very considerably after annealing and thereafter to diminish continuously during a period of 20 hr. The higher the temp. of anneal, the greater the initial increase of internal friction. Two causes operate to produce internal friction, viz. thermal agitation of atoms, mol., &c., and scattering of elastic waves at lattice defects, e.g. dislocations. The great increase of internal friction after annealing is attributed to the liberation of dislocations, which during the subsequent period become trapped, so that the internal friction is reduced.—J. S. G. T.

The Influence of Impurities on the Properties of Metals.—I.–IV. J. D. Fast (*Metalen*, 1952, **7**, (1), 2–6; 7–12; (2), 23–27; (3), 48–50).—Effects of impurities in metals are discussed with ref. to mech. properties; the interaction between impurities and dislocations is discussed, and the theory of strain-ageing is critically examined. F. concludes that impurities should not be considered as " enemies ", but rather used in controlled amounts to make improved alloys. —I. S. M.

†Mechanical Strength and Creep in Metals. N. F. Mott (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), 1952, 173–190; discussion, 190–196).—Effects, including those due to dissolved impurities which may lock dislocations without diffusing into them, as well as those due to impurities after a process of diffusion, suggested by Cottrell (*Phys. Soc.*: *Conf. on Strength of Solids*, **1948**, 30; *M.A.*, **16**, 322), responsible for the mech. strength of metals, are briefly summarized and discussed. Frank's ideas relating to the strength of pure metals (*ibid.*, p. 46; *M.A.*, **16**, 322), which depend on the distance between locking points of a Frank–Read source (*Phys. Rev.*, 1950, [ii], **79**, 722; *M.A.*, **18**, 465) are briefly reviewed, and a theory of exhaustion creep, i.e. creep in which the slowing down of extension is due to exhaustion of easily moved dislocations, which gives theoretical results in good agreement with the experimental results of Davis and Thompson (*Proc. Phys. Soc.*, 1950, [B], **63**, 847; *M.A.*, **18**, 597), is developed and discussed. 11 ref.—J. S. G. T.

Creep Under Complex Stress Systems at Elevated Temperatures. A. E. Johnson (*Proc. Inst. Mech. Eng.*, 1951, **164**, (4), 432–446; discussion, 447).—A comparison is made of the complex-stress creep characteristics of C steel, Al alloy, Mg alloy, and Nimonic 75 based on the results of a programme of creep tests under general stress systems at elevated temp. carried out at the National Physical Laboratory. 10 ref. —W. A. M. P.

***Rheology of Metals at Elevated Temperatures.** A. E. Johnson and N. E. Frost (*J. Mechanics Physics Solids*, 1952, **1**, (1), 37–52).—This account is part of a general study being made mainly on three materials of different crystal structures at elevated temp. in the range of practical appn. The metals are 0·17% C steel, R.R.59 Al alloy, and Mg–2% Al alloy, in the temp. range 350°–550°, 20°–250°, and 20°–150° C., resp. The work involves simple tension, torsion, and combined-stress creep tests, relaxation studies, and short-period plastic-strain tests. The results show good agreement with the von Mises criterion in complex-stress creep tests. The period to fracture (P) is related to the max. principal stress σ_1 by the relation: $\log P = A - B\sigma_1$, where A and B are const. The creep rates are compared with current theoretical formulæ: in general, curve fitting was possible only over a limited temp. range, and empirical formulæ are suggested. At low strain rates for R.R.59 alloy, creep and relaxation are directly related. Finally an investigation was made of the stress/strain curve for the Mg alloy at various temp. At all temp. this material obeyed the von Mises criterion of yielding. —E. O. H.

The Time Laws of Creep. A. H. Cottrell (*J. Mechanics Physics Solids*, 1952, **1**, (1), 53–63).—A review. Many empirical laws have been developed to give the relation between creep strain and time in the early stages of creep. C. discusses the theoretical basis for these formulæ, with particular ref. to transient creep, and develops some new dislocation models which show good agreement with experiment. He considers, however, that Andrade's $t^{\frac{1}{3}}$ law can be incorporated only if work-hardening is taken into account.—E. O. H.

Fatigue Weakness of Surfaces. J. O. Almen (*Product Eng.*, 1950, **21**, (11), 117–140).—A discussion of surface weakness and the effects of surface treatments (peening, carburizing, nitriding, &c.) on the properties of materials indicates that current testing procedures are misleading. 35 ref. —M. A. H.

***Plastic Deformation and the Meyer Constants of Metals.** M. A. Meyer and K. J. Blok van Laer (*Nature*, 1952, **169**, (4293), 237–238).—It is shown that if the true stress/strain curve of annealed metals is represented by $\sigma = g\varepsilon^x$ (where x and g are const. of the metal, σ the true stress, and ε the plastic elongation) then the ultimate Meyer hardness (H_u) must increase and n decrease with increasing amounts of cold work. This disagrees with the work of Finniston, Jones, and Madsen (*ibid.*, 1949, **164**, 1128; *M.A.*, **17**, 696) which suggests that H_u and n both increase with cold work for non-cubic metals. Experiments on Sn, Zn, and Cd, taking precautions to minimize recovery, showed H_u to increase and n to decrease. It is suggested that the effects observed by F., J., and M. were possibly due to recovery.—R. S. B.

Plastic Instability Under Plane Stress. H. W. Swift (*J. Mechanics Physics Solids* 1952, **1**, (1), 1–18).—This paper is concerned with the conditions governing the onset of necking, when the proportional decrease in cross-section equals, and subsequently exceeds, the corresponding increase in yield stress. Using the von Mises–Hencky yield condition, and allowing for strain-hardening, S. examines instability in the cases of biaxial tension, a thin cylinder, and a spherical shell. Appn. are also given for non-uniform stress distributions, in the case of hydrostatic bulging and in cylindrical shell pressing; in the latter case a full analysis is not given, however.—E. O. H.

On Discontinuous Plastic States, with Special Reference to Localized Necking in Thin Sheets. R. Hill (*J. Mechanics Physics Solids*, 1952, **1**, (1), 19–30).—H. studies the conditions under which discontinuities of stress, velocity, or surface slope may arise in a plastic-rigid material, deformed in its own plane. Using the Tresca yield condition, one velocity discontinuity is shown to represent the start of necking. The influence of work-hardening is discussed, and other yield conditions are also examined. An appn. of the results is made to the case of necking in notched strip.—E. O. H.

***General Theory of Small Elastic Deformations Superposed on Finite Elastic Deformations.** A. E. Green, R. S. Rivlin, and R. T. Shield (*Proc. Roy. Soc.*, 1952, [A], **211**, (1104), 128–154).—Math. Using tensor notations, a general theory is developed for small elastic deformations, of either a compressible or incompressible isotropic elastic body, superposed on a known finite deformation, without assuming special forms for the strain-energy function. Various appn. of the theory are given, including the problem of the infinitesimally small indentation, by a spherical punch, of the plane surface of a semi-infinite body of incompressible isotropic elastic material which is first subjected to a finite pure homogeneous deformation symmetrical about the normal to the force-free plane surface. 17 ref.—E. N.

***Determination of Dynamic Coefficient of Friction for Transient Conditions.** G. G. Gould (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (7), 649–654; discussion, 654).—A numerical value cannot be assigned to the coeff. of sliding friction between two materials. For most friction-clutch and brake appn., the friction forces during the first two runs of operation are of primary interest; however, during this time the forces do not remain const. even under otherwise steady operating conditions. Powder-metallurgy materials are commonly used for heavy-duty clutch appn. They consist chiefly of Cu contg. 15% Sn, Pb, C, and numerous other additives, e.g. SiO_2 MoS_2. These, latter are used to achieve desired friction characteristics, such as low wear rate or high coeff. of friction. Brushes made from these materials were mated up against drums of various steels, cast Fe, K Monel, and Cr-plated stainless steel. Curves derived from experimental data are presented, showing the behaviour of friction during the transient period.—H. Pl.

***On the Effect of Wear of Metals Against Clay Particles.** Tamaki Watanabe (*J. Mech. Lab.* (*Tokyo*), 1951, **5**, (5), 225–230).—[In Japanese]. Under the same frictional conditions, W. found that the wear of W, Ni, Fe, Cu, and Al against clay is nearly inversely proportional to the 4th power of their cohesive energies.—Author.

The Transfer of Metals to Plastics During Sliding. E. Rabinowicz and K. V. Shooter (*Proc. Phys. Soc.*, 1952, [B], **65**, (9), 671–673).—A radioactive metal slider was moved over a plastic surface. Autoradiographic studies indicated that, in every case, metal was transferred to the plastic, showing that local adhesions occurred. Values of the coeff. of friction and amounts of metal transfer are given.—E. O. H.

***Measurements of the Internal Friction of Metal Wires and Metal Wire Ropes by Means of Longitudinal Vibrations.** Sukeo Kawashima and Haruo Kimura (*Mem. Fac. Eng. Kyushu Univ.*, 1952, **13**, (1), 119–129).—[In English]. The internal friction of 2 steel and Cu wires, one stranded Cu wire, and 4 steel wire ropes was determined by measuring the logarithmic decrement of longitudinal vibrations excited electromagnetically in the wires and ropes suspended vertically and stretched by various loads, the extension of the wire or rope, while vibrating, being amplified and determined by an elect. amplifier. The longitudinal elasticity of wire ropes, E, increases with increase of the load. The dynamical modulus of elasticity is greater than the static modulus for wire ropes. The logarithmic decrement of the longitudinal vibrations increases about proportionally to the increase of initial amplitude and decreases with increasing load, in the case of the ropes, but for the metal wires the logarithmic decrement does not vary with the load. The logarithmic decrement of steel wire ropes is especially large compared with that of the wires, and increases with the number of wires in the rope, except in the case of the unpreformed type of rope. The internal friction of wire ropes is determined almost entirely by mutual friction of each wire and strand, and the internal friction of the constituent wires is very small.—J. S. G. T.

Residual Stresses in Machined Surfaces. E. K. Henriksen (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (1), 69–76).—Extremely high residual stresses are induced in the surface of a piece of metal by the action of cutting tools even when light cuts are taken. In ductile materials the residual stress induced is tensile. The stresses are closely related to the work-hardening properties and are concentrated near the surface. The resultant stress values depend largely upon the shape and angles of the tool.—H. Pl.

***†Surface and Interfacial Tensions of Single-Phase Solids.** J. C. Fisher and C. G. Dunn (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), 1952, 317–343; discussion, 343–351).—Published values of surface and interfacial tensions of single-phase solids, viz. for Au and Ag (Sawai and Nishida, *Z. anorg. Chem.*, 1930, **190**, 375; *Met. Abs.* (*J. Inst. Metals*), 1931, **47**, 7), for Au foil (Tammann and Boehme, *Ann. Physik*, 1932, [v], **12**, 820; *Met. Abs.* (*J. Inst. Metals*), 1932, **50**, 339), for Au wire (Alexander *et al.*, *U.S. Atomic Energy Commission Publ.*, 1950, (**SEP-43**); *M.A.*, **19**, 699), and for Cu/Pb vapour (Sears, *J. Appl. Physics.*, 1950, **21**, 721; *M.A.*, **18**, 483) are summarized, tabulated, and critically examined; several errors are corrected. Interface configurations at solid Cu|liq. Pb and Pb vapour, and of Cu|Cu|Pb and Cu|Cu|vapour junctions, and values of the various interfacial tensions, viz. Cu|Cu, Cu|liq. Pb., Cu|Pb vapour, Cu|Cu vapour, and liq. Pb|Pb vapour tensions are derived. The following best values of the resp. surface tensions are deduced: $\gamma_{Ag} = 1140$, $\gamma_{Cu} = 1430$, $\gamma_{Au} = 1510$, and $\gamma_{Cu|Pb\ vapour} = 760$ dynes/cm. Values of grain-boundary tensions in Cu and in γ-Fe are found as follows: $\gamma_{Cu|Cu} = 550$; $\gamma_{Fe|Fe} = 850$ dynes/cm., and the value of the coherent twin boundary tension in Cu $= \gamma_t = 19$ dynes/cm. Variation of grain-boundary tension with difference of orientation of abutting grains for 3·5% Si–Fe, Fe, Sn are shown graphically, and values of the twin-boundary tensions in Si–Fe, Cu, and Al are tabulated. 29 ref. —J. S. G. T.

***†Theory of Diffusion [in Solids].** Clarence Zener (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 289–314).—Cf. *J. Appl. Physics*, 1951, **22**, 372; *M.A.*, **19**, 9. A statistical-mech. theory of the atomic mechanism of diffusion in solids is discussed. The diffusion coeff. (D), being expressed in the form $D = D_0 \exp(-H/RT)$, it is shown that, for chem. diffusion, D_0 indicates whether diffusion occurs homogeneously throughout the matrix, or is confined to short-circuiting paths arising from imperfections, the demonstration being based upon the essential positive sign of the entropy of activation. Abnormally low values of D_0, found in cases of chem. diffusion, are attributed to short-circuiting diffusion paths along either grain boundaries or a continuous network of solute atoms. Conditions for normal values of D_0 are discussed. 44 ref.—J. S. G. T.

†The Adsorption of Gases on Metal Filaments, Films, and Single Crystals. Albert D. Crowell (*Amer. J. Physics*, 1952, **20**, (2), 89–98).—Some of the concepts involved in the study of the adsorption of gases on solids are discussed with special ref. to the problem of uniform surfaces. An historical account is also included which is sp. concerned with (1) metal filaments and evaporated films, and (2) surfaces of metal single crystals. 21 ref.—W. A. M. P.

On the Simultaneous Determination of the Optical Constants and Thickness of Very Thin Metallic Films. Florin Abelès (*Compt. rend.*, 1950, **231**, (19), 958–960).—Math. A. demonstrates that it is possible to deduce the thickness and opt. const. of thin metallic films in terms of the transmissions, absorptions, and refractive indices of the two media on either side of the film, which is assumed homogeneous and isotropic. —N. B. V.

On the Optical Properties of Very Thin Metallic Films. Florin Abelès (*Compt. rend.*, 1952, **234**, (2), 198–199).—A. has used the formulæ previously derived (*ibid.*, 1950, **231**, 958; preceding abstract) to examine the experimental data on Au and Ag films reported by Krautkrämer (*Ann. Physik*, 1938, [v], **32**, 537; *M.A.*, **5**, 526) and by Goos (*Z. Physik*, 1936, **100**, 95; *M.A.*, **3**, 494). Close agreement has been found. —N. B. V.

***The Phase Shift Accompanying Reflection of Light from Evaporated Films and the Change of Effective Refractive Index as a Function of the Thickness of Deposit.** Shigenori Nawata (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 740–747).—[In English].—In continuation of previous work (*ibid.*, (1), 107; *M.A.*, **20**, 75) N. extends his theory of the phase shift accompanying reflection from a thin film to include reflection from multiple-layer films on a base substrate, and calculates the phase shift and effective refractive index of the film as a function of the film thickness. Phase shifts at air|Al, air|(Al opaque + ZnS), and air|(Al opaque + MgF_2 + ZnS) interfaces were determined as a function of film thickness from the displacement of Fizeau fringes. 9 ref. —J. S. G. T.

The Contribution of the Surface to the Specific Heat of Disperse Systems. H. Koppe (*J. Chem. Physics*, 1950, **18**, (5), 638–640).—Extension to binary-system interfaces of a theory by Brager and Schuchowitzky (*Acta Physicochim. U.R.S.S.*, 1946, **21**, 1001) regarding the contribution of the surface to the sp. heat shows that with some limiting exceptions, the contribution of the interface between two solids is very nearly the same as that which would arise from an equal free surface of the acoustically denser substance. Surface effects of this kind could be also expected in eutectic alloys and polycryst. substances with high elastic anisotropy.—J. R.

Theory of the Specific Heat of Chain and Layer Structures. V. V. Tarasov (*Zhur. Fiz. Khim.*, 1950, **24**, (1), 111–128).—[In Russian]. T. expounds the Debye quantum theory of sp. heats, according to which the sp. heat of a three-dimensional crystal lattice $\propto T^3$ at temp. $T \ll$ the Debye characteristic temp. θ_3, and then points out that in some substances the low-temp. sp. heat $\propto T^2$ ($MnCl_2$) or $\propto T$ (Sb_2O_3, As_2O_3, Bi_2O_3). These relationships appear to be connected resp. with "layer" and "chain" crystal structures, in which the interaction forces between layers or chains of atoms are smaller than the forces within the layers or chains. T. then works out the Debye treatment for an m-dimensional lattice, obtaining for the sp. heat C_m at temp. T the formula:

$$C_m = 3m(m+1)R\left(\frac{T}{\theta_m}\right)^m \int_0^{\theta_m/T} (x^m/e^x - 1)dx - 3mR\left(\frac{\theta_m}{T}\right)\left[\exp\left(\frac{\theta_m}{T}\right) - 1\right]^{-1}.$$

R being the gas const. and θ_m a const for any substance, analogous to the Debye characteristic temp. Tables of this function for $m = 1$ (chain structures), $m = 2$ (layer structures), and $m = 3$ (Debye function) are given. T. shows that if $\theta_m/T \gg 1$, $C_1 = \pi^2 R(T/\theta_1)$, $C_2 = 43{\cdot}272\ R(T/\theta_2)^2$, and $C_3 = (12/5).\pi^4 R(T/\theta_3)^3$. The C_2 function satisfactorily describes the low-temp. sp. heats of As, with $\theta_2 = 331^\circ$ K., and of Sb, with $\theta_2 = 223^\circ$ K. (Anderson, *J. Amer. Chem. Soc.*, 1930, **52**, 2298); and of Ga, with $\theta_2 = 257^\circ$ K. (Landolt–Börnstein, "Physikalisch-Chemische Tabellen" 2. Erg. Bd., **1931**, II, 1169). The C_1 function describes the low-temp. sp. heat of cryst. Se, with $\theta_1 = 371^\circ$ K. (Anderson, *J. Amer. Chem. Soc.*, 1937, **59**, 1036; *M.A.*, **4**, 484). T. develops his theory to allow for interaction forces between the chains or layers of atoms, by using the Debye distribution function for lattice vibration frequencies up to a critical value, and the one- or two-dimensional distribution function at higher frequencies. —G. B. H.

***Research on Materials and Phenomena of Electrical Contacts.** Hidekiyo Asai (*Technol. Rep. Osaka Univ.*, 1952, **2**, (32), 81–92).—[In English]. Metal transfer characteristics in elect. contacts of Ag, Cu, Cd, and Bi were studied. The direction and amount of metal transfer, on operating the contact, are determined mainly by two factors, viz. (1) the impact energy of electrons against the anode member, and (2) the kinetic energy of positive ions at the cathode member, both being influenced by the phys. and chem. properties of the contact materials and the atmosphere. Two factors were found capable of reversing the direction of transfer of metal, viz. (*a*) the p.d. effecting ionization of the metal or the atmosphere, and (*b*) the amperage necessary to balance the kinetic energy of the transferred metals. Of these (*a*) is definite and characteristic of both metal and atmosphere, and the process of reversion is continuous, while (*b*) is probably absent during the contact-opening operation, under certain atmospheric conditions, although the tendency to reverse is continuous.—J. S. G. T

Hall Effect and Ponderomotive Force in Simple Metals. Norman Rostoker (*Amer. J. Physics*, 1952, **20**, (2), 100–107).—The Bloch–Sommerfeld theory of conductivity is applied to the calculation of momentum lost by conduction electrons because of collisions with the crystal lattice. The calculations provide the basis for a microscopic interpretation of the ponderomotive force $f = (1/c)(\gamma \times H)$/unit vol. The relation between this force and the Hall effect is discussed.—W. A. M. P.

***Photoelectric Absorption of γ Rays in the L Layer of Atoms.** (Mme) Nadine Marty (*J. Phys. Radium*, 1952, **13**, (7/9), 401–404).—Theoretical and experimental data relating to photoelect. absorption by the L and K layers of atoms are briefly reviewed and a spectrometric method for determining the ratio, τ_K/τ_L, of the resp. effective photoelect. absorbing sections of the K and L layers for photons of energies 411 and 140 keV. is described. Experimental values of the ratio are: 14% for a Ag diffusing layer, 25–27% for Pb and Au, and 31% for U. 13 ref.—J. S. G. T.

***Supraconductivity and Electrodynamic Potential.** M. v. Laue (*Z. Physik*, 1949, **125**, (7/10), 517–530).—Math. The work done by a quasi-stationary magnetic field when bodies change their positions, or when part of a normal conductor becomes supraconducting, is calculated.—J. W. C.

Metals and Alloys: The Pauling Hypothesis. W. Hume-Rothery (*Ann. Rep. Progress Chem.*, 1949, **46**, 42–56).—A review, with many ref.—N. B. V.

---

2 — PROPERTIES OF ALLOYS

Some Characteristics and Particulars of the State of Development of Aluminium and Aluminium Alloys. W. Bleicher (*Aluminium*, 1952, **28**, (1/2), 20–27).—B. briefly reviews modern trends in alloy development and the manufacture of semi-products (continuous casting, cladding, extrusion, drawing of age-hardened sheet by taking advantage of reversion phenomena, design of riveted structures, welding materials, research, brazing of clad materials, resin bonding, surface treatments). 16 ref.—M. A. H.

***Dilatometric Determination of the Solidus Temperature [of Aluminium Alloys].** (Verö). See col. 367.

***Thermal Expansion of Aluminium and Some Aluminium Alloys.** Peter Hidnert and H. S. Krider (*J. Research Nat. Bur. Stand.*, 1952, **48**, (3), 209–220).—Coeff. of expansion between −50° and +400° C. are summarized for cast and annealed Al, and for alloys of various compn. from the systems Al–Be, Al–Cu, Al–Si, Al–Cu–Fe, Al–Cu–Mg, Al–Cu–Ni, Al–Si–Cu, Al–Si–Mg, Al–Cu–Ni–Mg, Al–Cu–Sn–Zn, Al–Si–Cu–Mn, Al–Si–Cu–Ni, Al–Si–Ni–Cu–Mn, and Al–Si–Ni–Cu–Mo. The data have been obtained over the last 25 years by the fused-quartz tube and dial-indicator method (up to 20° C.) and the precision micrometric method (20°–400° C.). Results obtained from both heating (expansion coeff.) and cooling (contraction coeff.) are given; where these do not agree there is a difference in length after heating and cooling, owing to structural changes. In the binary alloys the variation of expansion coeff. is almost linear with at.-% of solute; the decrease is greatest in the order Si>Cu>Be. The results for pure Al are compared with those of previous workers. In the ternary systems diagrams are given on which the "isodils" (curves of const. coeff. of expansion in given temp. range) are plotted. 16 ref.—J. W. C.

***Equilibrium Relations at 460° C. in Aluminium-Rich Alloys Containing 0–7% Copper, 0–7% Magnesium, and 1·2% Silicon.** H. J. Axon (*J. Inst. Metals*, 1952–53, **81**, (4), 209–213).—The equilibrium isothermal at 460° C. is given for quaternary alloys rich in Al and contg. 0–7% Mg, 0–7% Cu, and const. Si (1·2%). The phases encountered are the Al-rich solid soln., $CuAl_2$, Mg_2Si, Si, the ternary phase Al_2CuMg, and a quaternary phase which probably has the compn. $Al_5Cu_2Mg_8Si_6$. These phases give rise to fourteen separate phase fields. A new 460° C. isothermal for the Al-rich region of the Al–Cu–Si system is also given.—AUTHOR.

***Change in the Solid-Solution Range Under Hydrostatic Pressure in the Al–Mg and Al–Ag Systems.** M. I. Zakharova and V. A. Il'ina (*Zhur. Fiz. Khim.*, 1950, **24**, (6), 714–717).—[In Russian]. Al-rich Al–Mg alloys were prepared from electrolytic Al (contg. 0·048% impurities) and electrolytic Mg, by casting and annealing first at 450° C. and then at temp. of 600°, 525°, 475°, 450°, 400°, 350°, 300°, and 250° C. until equilibrium was attained. The alloys were then annealed at the same temp. under pressures up to 106 kg./mm.², applied by a Brinell press, and water-quenched under pressure. The solid solubility of Mg in Al was determined at each temp. by lattice-parameter measurements, supplemented by micrographic examination. Annealing times varied from 2 hr. at 600° C. to 34 hr. at 250° C., and in all cases further annealing led to no change in parameter. The solid solubility decreased considerably under pressure, both by comparison with Hanson and Gayler's diagram (*J. Inst. Metals*, 1920, **24**, 201) and by comparison with the same alloys annealed at the same temp. under atmospheric pressure. The extent of lowering of solid solubility decreased with temp., from 5% at 450° C. to 0·8% at 250° C., both under a pressure of 106 kg./mm.². The solid solubility at 400° C. was lowered from 13·6 to 7·5% by a pressure of 106 kg./mm.², but the change was not proportional to pressure. A similar investigation of the solid solubility of Al in Mg under pressure showed a considerable lowering; at 430° C. under 106 kg./mm.², the solid solubility was found to be 9·4%, compared with Schmid and Siebel's value (*Z. Physik*, 1933, **85**, 36; *M.A.*, **1**, 497) of 12·5% at atmospheric pressure. Similar results were obtained for the solid solubility of Ag in Al, which has the following values under a pressure of 106 kg./mm.²; at 550° C., 25·6%; at 500° C., 15·5%; at 450° C., 9·6%; at 400° C., 5·1%; at 350° C., 3·0%; at 300° C., 1·5%. The corresponding values at atmospheric pressure range from 48% at 550° C. to 3·25% at 300° C.—G. B. H.

***Study of Crystallization and Mechanical Properties of Secondary Silumin Al 4 as Affected by Addition of Nickel, Phosphorus, and Boron.** A. I. Patsiornykh (*Trudy Tsentral. Nauch.-Issledovatel. Inst., Ministerstvo Sudostroitel. Prom. S.S.S.R.*, 1947, **7**, (2), 25–30; *C. Abs.*, 1951, **45**, 1940).—[In

Russian]. The secondary Silumin used in these tests contained Si 8·5–10·5, Mg 0·17–0·30, Mn 0·25–0·50, Cu $\leqslant$ 0·5, Fe 0·7%, balance Al. Before testing, the specimens were soln.-treated, water-quenched from 525°–530° C., and artificially aged at 185°–190° C. for 5 hr. In ordinary Silumin of this kind Si appears as round or elongated grey inclusions in the white matrix of a solid soln. 0·1% Ni caused the Si to appear in a needle-like form, but its distribution remained uneven. 0·3% Ni improved the structure and decreased the size of the Si particles; 0·5% Ni decreased the size of the Si inclusions still further, but its distribution became uneven again and porosity appeared. 1% Ni coarsened the structure considerably and caused the sepn. of Fe–Si constituents and Al–Ni compounds. Thus, the optimum Ni range was 0·1–0·5%. P affected the crystn. of Silumin markedly; 0·016% imparted a fine cryst. form. Raising the P content to 0·032–0·06% somewhat coarsened the structure, but at 0·6% P, Si crystals appeared as cracked plates with uneven edges, and porosity became evident. A P content above 0·048% affected the mech. properties adversely. The range of beneficial effect of P was therefore 0·016–0·048%. B under conditions of these tests did not show any beneficial effect on Silumin. More extensive study is needed for drawing reliable conclusions.

***On the Properties of the Transition Point of Tensile Curves and Its Influence on the Ageing of Aluminium-Base Alloys.** B. Jaoul, F. Aubertin, and C. Crussard (*Rev. Mét.*, 1952, **49**, (9), 633–644; discussion, 644–646).—J., A., and C. have continued the investigations of C. and J. (*ibid.*, 1950, **47**, 589; *M.A.*, **18**, 248) on the transition points of tensile curves of Al–Mg and Al–Cu alloys by studying the effect of the alloying metals by X-ray and micrographic examination, and by considering the elect. properties of these alloys. In tension the deformation curve changes its slope when it exceeds a critical value; the transition point corresponds to an abrupt change of direction of the load/deformation curves. Its position depends essentially on the concentration of the alloying elements in solid soln., and, doubtless, on the temp.; but this investigation was carried out only at room temp. The elongation at this point $\propto$ the cube root of the at.-% of alloying element. The limit of proportionality varies according to the nature of the metal dissolved in the Al; it is low for Zn and Mg. The transition point corresponds to a change in the various phys. properties accompanying the deformation, i.e. on the mechanism of cold working. The X-ray study showed that (i) for elongations below the transition point, the spreading due to asterism is continuous and annealing produces no fragmentation of the asterism spots, and (ii) in specimens deformed beyond the transition point, the asterism spots are broken up into more or less indistinct points, corresponding to imperfect fragmentation or polygonization, which becomes more marked on annealing. Very interesting results were obtained from measurements of the change in elect. resistivity at room temp. of an Al–Mg alloy immediately after quenching and cold working. An attack which develops etch pits on a specimen electrolytically polished after quenching reveals sub-boundaries or intracryst. contours of polygonization. If the specimen is again polished, deformed, and attacked, a small amount of cold work neither alters these contours nor forms new ones; but heavy working effaces the contours, which reappear slowly at room temp. Thermoelect. measurements of the Al–4% Cu alloy show that light working after quenching below the transition point slightly hastens the structural hardening, but reduces the total extent. Further observations on this alloy are discussed. The hypotheses to cover these results have not yet been agreed upon; but the results all indicate that the transition point corresponds to a change in the mechanism of cold working. Broadly, after deformation in which the small groups of dislocations are distributed throughout the metal, there appears at the transition point a tendency to the regrouping of the dislocations in zones of marked rumpling, producing a fragmentation of the metal. By ageing, the distinctness of the folds is accentuated at the same time as the interior of the fragments becomes more perfect: this is true polygonization. 12 ref.—J. H. W.

***A Study of Poisson's Ratio in the Yield Region [of Aluminium Alloys].** G. Gerard and S. Wildhorn ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (**2561**), 30 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 358).—Al alloys 24S-T4 (rolled) and 14S-T6 and 75S-T6 (extruded) were tested in tension and compression along each of the three principal axes. The data indicate that tensile loading in the yield region is accompanied by a permanent decrease in vol., whereas compression results in a permanent increase in vol. These data were subsequently checked by *d* measurements on several specimens.

***Some Fatigue Tests on Aluminium-Alloy and Mild-Steel Sheet, With and Without Drilled Holes.** C. E. Phillips and A. J. Fenner (*Proc. Inst. Mech. Eng.*, 1951, **165**, (65), 125–129; discussion, 130–140; and (abridged) *Mech. World*, 1951, **129**, (3348), 251–252).—An account of some fatigue tests of Al alloy and mild steel panels, with and without central holes. The presence of a hole generally causes a marked reduction in fatigue strength, but the smallest holes have only a slight effect.—W. A. M. P.

***Fatigue Strengths of Aircraft Materials [Aluminium Alloys and Steels].** H. J. Grover, W. S. Hyler, and L. R. Jackson ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (**2639**), 22 pp.).—The report presents results of axial-load fatigue tests on notched specimens of three sheet materials: 24S-T3, and 75S-T6 Al alloys and normalized SAE 4130 steel. Each specimen was notched by edge notches designed to have a theoretical stress-concentration factor of 1·5. Tests were carried out at four levels of nominal mean stress: 0, 10,000, 20,000, and 30,000 lb./in.^{2}. The results of these tests extend information previously reported from tests on unnotched specimens and tests on specimens more severely notched, and afford data on the variation of fatigue-strength reduction with notch severity.—AUTHORS.

***Spectral Characteristics of Antimony-Cæsium Photocathodes in the Ultra-Violet Region Down to 1500 Å.** Vladimir Schwetzoff and (Mme) Simone Robin (*Compt. rend.*, 1951, **233**, (6), 475–478).—S. and R. have measured the characteristics of a vacuum Sb–Cs photocell formed in the absence of O, from very pure materials. The current/wave-length curve shows a max. at approx. 3550 Å. and another very small max. at $\sim$1600 Å. Discrepancies between these results and those of earlier workers are thought to be due to the presence of impurities.—N. B. V.

***Spectral Characteristics of Antimony–Potassium and Antimony–Sodium Photocathodes in the Ultra-Violet Region Down to 1500 Å.** Vladimir Schwetzoff and (Mme) Simone Robin (*Compt. rend.*, 1951, **233**, (7), 518–520).—Cf. *ibid.*, (6), 475; preceding abstract. The characteristics of Sb–K and Sb–Na photocathodes, formed, in the absence of O, from very pure materials, were measured in the light from a W filament at 2550° K. The overall sensitivity for Sb–K was 2·3 μamp./lumen and very much less for Sb–Na (cf. 12 μamp./lumen for Sb–Cs). Max. occur at $\sim$3400 Å. in the case of Sb–K and $\sim$2900 Å. in that of Sb–Na.—N. B. V.

***Influence of Chemical Composition on the Rupture Properties at 1200° F. (650° C.) of Wrought Cr–Ni–Co–Fe–Mo–W–Nb Alloys.** E. E. Reynolds, J. W. Freeman, and A. E. White (*J. Metals*, 1952, **4**, (9), 946–958).—Alloys forming 62 variations of the basic alloy Cr 20·0, Ni 20·0, Co 20·0, Fe 32·0, Mo 3·0, W 2·0, and Nb 1·0%, together with small amounts of C, Mn, Si, and N, were prepared by melting in a small induction furnace, with and without the use of deoxidizers, poured into taper ingots, forged at initial and final temp. of 2200° and $>$1800° F. (1205° and 980° C.), resp., soln. treated for 1 hr. at 2200° F., quenched in water, aged for 24 hr. at 1400° F. (760° C.), and air cooled. Stress–rupture tests at 1200° F. (650° C.) were then carried out on pieces 0·250 in. dia. with a 1-in. gauge-length, and were of sufficient duration to establish the 100 hr. and to estimate the 1000 hr. rupture strengths; hardness and metallographic examinations were made on all the alloys. Variations in alloy compn. were found to have the following effects on the rupture properties: (1) negligible—C 0·08–0·6, Mn 0·3–2·6, N 0·03–0·18, Ni 11–20, Co 20–33, and Nb 2–4%; (2) beneficial—N 0·004–0·08, Cr 10–30, Ni 0–11, Co 0–20, Mo 0–4, W 0–4, and Nb 0–1%; and

(3) detrimental—Ni 20–30, and Si 0·5–1·6%. Such variations in strength are the result of changes in inherent creep-resistance caused by alloy addn., except for compn. changes for which total elongation to fracture altered from very low (<10%) to high (>20%) values, in which case the greater ability to deform before fracture was the dominating factor. It is considered that rupture strength improvements are caused by: (i) Mo and W atoms entering substitutionally into the matrix solid soln., (ii) Ni and Co increasing the solubility of Mo and W, (iii) Cr having an effect on the ageing reaction, and (iv) small addn. of Nb increasing the ability to deform before fracture. The detrimental effects of Si are possibly connected with melting phenomena. Nb, Cr, and C produce major changes in the microstructure of the alloys. 17 ref.—E. N.

***Concentration Dependence of Diffusion Coefficients in Metallic Solid Solutions [with Particular Reference to Binary Systems with Copper, Iron, and Lead].** Donald E. Thomas and C. Ernest Birchenall (*J. Metals*, 1952, **4**, (8), 867–873).—Using chem. pure Pd and O.F.H.C. Cu, the diffusion coeff. in the Cu–Pd system have been determined, and are shown graphically as a function of concentration for three temp., 878°, 972°, and 1038° C. The concentration curves are S-shaped; the diffusion coeff. (D) is nearly const. in both dil. soln.—approx. 0–20 and 80–100 at.-% Cu—and varies strongly in the intermediate region—20–80 at.-% Cu; D approaches the value of the self-diffusion coeff. of the solvent metal as the solute concentration approaches zero. The activation energy for diffusion has been calculated as 31, 34, and 40 kg.cal./mole at 0, 50, and 100 at.-% Pd—values which are rather inaccurate and probably much too low. A study of diffusion at the high-Cu (65–100 at.-%) end of the Cu–Ni system confirms the fact that in relatively dil. soln. D is extremely dependent on compn. At 89·9 at.-% Cu and temp. of 923°–1048° C. D_{Cu} is 25–30% greater than D_{Ni}, there being no appreciable variation with temp. An analysis of the results, together with those of other workers on the diffusion of various metals in Cu, Fe, and Pb, shows that the greater the m.p. depression of the solute, the more the diffusion coeff. exceeds that of the pure solvent metal; both factors tend to become greater the more limited the solid solubility of the solute metal. 23 ref.—E. N.

***Elastic Anisotropy of (α + β) Brass Strip.** H. Asada, E. Tanaka, and K. Koike (*Rep. Inst. Sci. Technol., Tokyo*, 1952, **6**, (1), 19–25).—[In Japanese]. The elastic anisotropy of (α + β)-, α-, and β-brass, cold rolled and annealed, was measured. The results show that: (1) The elastic anisotropy of cold-rolled (α + β)-brass does not depend on the β phase but on the α phase. When the α phase is cold-rolled a preferred orientation is gradually set up, but the β phase cannot assume a preferred orientation and breaks down at 30% cold reduction or more. (2) When cold-rolled (α + β)-brass is annealed at 200° C., Young's modulus increases in each direction. (3) Elastic anisotropy of recrystallized (α + β)-brass strip (annealed at 400° C.) decreases, and the value of Young's modulus in each direction becomes the mean value of Young's modulus for α- and β-brass strip. (4) The curve of the Young's modulus calculated from the pole figure of β-brass strip is very similar to the observed curve, but for α-brass the two curves are different.—Authors.

***Observations on the Bauschinger Effect in Copper and Brass.** (Schwartzbart, Jones, and Brown). See col. 322.

***Topochemical Selenization of Single-Crystal Cuprous Chloride and a Method of Analysing the Copper Selenide Obtained.** Roger Molé (*Compt. rend.*, 1951, **233**, (7), 523–525).—By the action of gaseous H_2Se on single crystals of CuCl, a selenide $Cu_{2-x}Se$ has been obtained, which is similar to the corresponding sulphide $Cu_{2-x}S$. The compound is not stoichiometric, but lies between Cu_8Se_5 and $Cu_{8 \cdot 3}Se_5$.—N. B. V.

***The Synthesis of Two Selenides of Copper by Compression of Copper and Selenium Powders.** Raymond Hocart, Roger Molé, and (Mme) Louis Schué-Muller (*Compt. rend.*, 1951, **233**, (12), 661–662).—Mixtures of pure Cu and Se powders in the ratio of 9 Cu : 5 Se were compacted at pressures between 350 and 17,500 kg./cm.² and the products examined microscopically and by X-rays. At low pressures a selenide with an orthorhombic structure (umangite) was formed with the formula Cu_3Se_2 or more probably $Cu_{4-x}Se_2$ ($a_0 = 4 \cdot 24$, $b_0 = 6 \cdot 38$, $c_0 = 12 \cdot 44$ Å.). At high pressure a selenide of the $Cu_{2-x}Se$ type (bezelianite) is formed, with a cubic structure ($a_0 = 5 \cdot 70$ Å.). At intermediate pressures both phases occur. —N. B. V.

***The Synthesis of Two Tellurides of Copper by Compression of Copper and Tellurium Powders.** Raymond Hocart and Roger Molé (*Compt. rend.*, 1952, **234**, (1) 111–113).—Cf. preceding abstract. Mixtures of Cu and Te powders in the ratio 9 Cu : 5 Te were compacted at various pressures between 3500 and 17,500 kg./cm.² at room temp. and 150° and 250° C. Microscopic and X-ray examination showed the existence of two compounds: (1) a pseudo-cubic (probably orthorhombic) $Cu_{2-x}Te$-type phase (weissite) with $a_0' = 10 \cdot 0_5$ kX; this appeared under all conditions; and (2) a tetragonal $Cu_{4-x}Te_2$-type phase (rickardite), with $a_0 = 3 \cdot 95$, $c_0 = 6 \cdot 06$ kX; this was found chiefly in the specimens compressed at 150° C.—N. B. V.

***Constitution of Alloys Between Iron and the Compound CoAl.—I.** O. S. Ivanov and M. A. Skryabina (*Izvest. Akad. Nauk S.S.S.R.*, 1949, [Khim.], (3), 242–253; *C. Abs.*, 1951, **45**, 520).—[In Russian]. The investigation of this section of the Fe–Co–Al system was suggested by theoretical indications that the previously reported complete solubility at room temp. was incorrect. Since the CoAl compound has twice the molar heat of formation of FeAl, it was concluded that a two-phase region should exist in preference to a solid soln. in this section, as in the Fe–NiAl section. A 50-g. melt of each of 43 alloys was made in an induction furnace in an Al_2O_3 crucible under a NaF and Ca_2F flux with Armco Fe, high-purity Co, and master alloys of CoAl and FeAl, the latter contg. 0·20% Si and 0·015% C. Cast specimens 3 × 60 mm. were homogenized for 72 hr. at 1000° C., cooled at 10° C./hr. to 500° C., and then furnace-cooled. Their structures corresponded to equilibrium at 500°–600° C. The variation of phys. properties with compn. indicated the presence of a two-phase region. Saturation magnetization fell continuously from the max. value at 100% Fe. The coercive force reached a max. of 90 Oe. at 40 : 60 Fe–CoAl. Sp. elect. resistance reached a max. value of 2·3 ohm-mm.²/m. at 20 : 36 : 44 Fe–Co–Al. Debye X-ray studies confirmed the presence of a two-phase region from ~25 to 70% Fe. A given line of the 2 phases could not be separated, but superlattice lines of the CoAl phase could be observed. Photomicrographs (×500) gave weak evidence of the presence of two phases, and thermodynamic calculations also gave further support for this view.

The Magnetic Structure of Alnico 5. E. A. Nesbitt and R. D. Heidenreich (*Elect. Eng.*, 1952, **71**, (6), 530–534).—Structures and properties of Alnico 5 (Al 8, Ni 14, Co 24, Cu 3, Fe 51%) heat-treated along directional magnetic fields are illustrated and discussed. 11 ref.—J. R.

***Investigation of Plastic and Other Properties of Alloys of the System Iron–Nickel–Tantalum.** K. A. Osipov (*Izvest. Akad. Nauk S.S.S.R.*, 1951, [Tekhn.], (6), 848–851; *Appl. Mechanics Rev.*, 1952, **5**, 309).—[In Russian]. Sp. elect. resistance, hardness, and deflections of rods of several Fe–Ni–Ta alloys (contg. 5, 10, 15, 20, 25, 30, and 35% Ta) were determined. All specimens were cast, annealed for 48 hr. and quenched in oil, and the rods were subjected to bending in a centrifugal testing machine (method of I. I. Kornilov). Curves of equal sp. elect. resistance and of equal deflections in bending of alloys in the region of the λ soln. are given. In the region of the γ soln., the deflections in bending diminish, as the Fe content increases or as the limit of solubility is approached. Transition in the two-phase region (γ + Fe_2Ta) is accompanied by smaller deflections, while transition in the (γ + Ni_3Ta) region gives greater deflections.

***Solubility of Magnesium in Lithium.** J. A. Catterall (*Nature*, 1952, **169**, (4295), 336).—The boundary of the equilibrium diagram showing the solubility of Mg in Li has been redetermined by microscopical examination of the quenched alloys. At 588° C. the solubility limit was found to be 24·1 at.-% Li, a lower figure than those previously given (Henry and Cordiano, *Trans. Amer. Inst. Min. Met. Eng.*, 1934, **111**, 319; see *M.A.*, **1**, 382; and Saldau and Schamray, *Z. anorg. Chem.*, 1935, **224**, 388; *M.A.*, **2**, 577).—R. S. B.

***Magnesium-Rich Alloys of the System Magnesium–Aluminium–Manganese.** N. V. Ageev, I. I. Kornilov, and A. N. Khlapova (*Izvest. Sekt. Fiz.-Khim. Anal.*, 1948, **16**, (4), 130–143; *C. Abs.*, 1951, **45**, 521).—[In Russian]. The purpose of the investigation was to determine the limit of solubility of Al and Mn when present together in solid Mg, to determine the boundaries of phases adjacent to the ternary solid soln., and to study the ageing properties and corrosion-resistance of the alloys. The work was carried out on Mg alloys with up to 40% Al and up to 10% Mn. At the beginning of crystn. in the Mg region there separated first the α-phase, which is a ternary solid soln. having the crystal lattice of Mg. A Mn-rich phase separated next, and is referred to as *X*, the compn. of which was not determined. Next came γ, which is a solid soln. having a crystal lattice corresponding to Al_2Mg_4, and a phase having the crystal lattice of Mn. In the presence of Al, the solubility of Mn in Mg at 450° C. rose from 0·3% to 1·25–1·5% and in slowly cooled alloy from 0 to 0·5–0·8%. The hardness of the alloys increased considerably with the Al content; lowering the Mn content did not appreciably affect the hardness. Alloys within the field of ternary solid soln. contg. Al 10–14 and Mn 0·2–0·5% could be aged artificially at 150°, 175°, and 200° C., and attained a max. hardness of 90–100 Brinell. At 100° C. the alloys aged very slowly and the hardness changed insignificantly. The preferred compn. for these alloys are: Al 10–13 and Mn 0·2–0·5%; the most corrosion-resistant alloys contained Al 2–7 and Mn 0·05–1·5%.

***Isothermal Transformation of Supersaturated Solid Solutions of Aluminium and Manganese in Magnesium.** I. I. Kornilov and A. N. Khlapova (*Izvest. Sekt. Fiz.-Khim. Anal.*, 1948, **16**, (4), 144–150; *C. Abs.*, 1951, **45**, 521).—[In Russian]. The process of transformation was studied in alloys contg. Al 2–14·50 and Mn 0·10–1·15% at 100°–350° C. The alloys fall within the field of supersaturated ternary solid soln. Specimens were kept for 3 days sealed in evacuated tubes at 430° C. and were then transferred to a furnace maintained at the requisite temp. where they were kept for a definite time, after which they were water-quenched. At 350° and 300° C. the precipitated phase appeared in the form of large shiny particles throughout the entire field of the solid soln. There was no change in hardness. At 250°–150° C. the precipitated phase was pearlitic; the pptn. started along grain boundaries. Decompn. of the solid soln. at this temp. was accompanied by a considerable increase in hardness. At 100° C. no noticeable change was observed in the microstructure or hardness. The rate of transformation increased with the Al and Mn content. In consequence of the isothermal transformation the same hardness was attained as upon ordinary ageing, though in less time.

***A Note on Magnesium Telluride.** W. Klemm and K. Wahl (*Z. anorg. Chem.*, 1951, **266**, (6), 289–292).—The prepn. of MgTe from Mg and Te is described. The compound has a wurtzite structure with $a = 4·53 \pm 0·01$ and $c = 7·38 \pm 0·02$ Å. In a number of prepn. the *d* of the compound was found to be approx. 3·80, in satisfactory agreement with the $d = 3·85$ deduced from X-ray results.—J. W. C.

Navy Develops New [Manganese–Bismuth] Permanent Magnetic Material. —— (*Steel*, 1952, **131**, (4), 76–77).—The new magnetic material called Bismanol is an alloy of Mn bismuthide and unreacted Mn and Bi, and has a coercive force of 3000 Oe. The method of manufacture consists of heating at 700° C. for 8–16 hr. a mixture of powdered Mn and Bi. The magnetic portions are then separated and hot pressed to shape in a strong magnetic field.—E. J.

An Introduction to Arc-Cast Molybdenum and Its Alloys. (Ham). See col. 323.

***A Study of Order–Disorder and Precipitation Phenomena in Nickel–Chromium Alloys.** A. Taylor and K. G. Hinton (*J. Inst. Metals*, 1952–53, **81**, (4), 169–180).—Elect. resistivity, sp. heat, and X-ray measurements have been made on Ni_3Cr and $Ni_{15}Cr_4Al$. It has been shown that both these alloys undergo an order–disorder type of transformation at approx. the same temp. and that an additional high-temp. transformation occurs in the ternary alloy, consistent with the pptn. and the re-soln. of a second phase.—AUTHORS.

***Ferromagnetic Behaviour of Nickel–Cobalt Alloys.** Mikio Yamamoto (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 14–27).—[In English]. The magnetization curves for magnetic fields up to ~900 Oe., the magnetization for a number of const. fields, the saturation magnetization, magnetocryst. anisotropy const., and initial and max. susceptibilities as functions of compn. were measured for ~60 annealed Ni–Co alloys covering the whole compn. range., at room temp. The two-phase range in the alloy system was shown to extend approx. from 68 to 78%, contrary to current opinion that the range was quite narrow near 70% Co. This conclusion is supported by Y.'s measurements of the *d* of the alloys (*ibid.*, 1950, [A], **2**, 871; *M.A.*, **19**, 365) and by Masumoto's magnetic measurements (*Sci. Rep. Tôhoku Imp. Univ.*, 1929, [i], **18**, 195; *J. Inst. Metals* (Abstracts), 1929, **42**, 459). 21 ref.—J. S. G. T.

***Alloys of Palladium with Copper.** V. A. Nemilov, A. A. Rudnitsky, and R. S. Polyakova (*Izvest. Sekt. Platiny*, 1949, (24), 26–34; *C. Abs.*, 1951, **45**, 522).—[In Russian]. Thermal analysis was carried out to determine the temp. of transformations in the solid state. The alloys used contained 10–60 at.-% Pd. The alloys were annealed *in vacuo* for 5 days at 850° C., slowly cooled to 550° C., and kept at this temp. for 5 days, cooled to 300° C., and kept at this temp. for 5 days, and then cooled to room temp. within 6 hr. Alloys contg. 35–50 at.-% Pd were in addn. heated for 14 days at 550°–600° C. and cooled slowly. The heating curves showed two transformations with max. near 17 and 37 at.-% Pd and temp. of 525° and 650° C., resp. These points correspond to $PdCu_5$ and Pd_3Cu_5. For hardness detn. annealed and hardened specimens were used. The hardened specimens gave a smooth curve having a flat max., which is usual for a continuous series of solid soln. The annealed specimens gave a curve with two min. at 17 and 37 at.-% Pd. Elect. resistance was measured at 25° and 100° C. on annealed and hardened specimens. Both curves had a max. at approx. 60 at.-% Pd. The curve of annealed specimens showed two min. at 17 and 37 at.-% Pd. The results indicate the formation of 2 daltonides $PdCu_5$ and Pd_3Cu_5.

Titanium and Its Alloys. (Williams). See col. 326.

***Titanium–Chromium Phase Diagram.** F. B. Cuff, N. J. Grant, and C. F. Floe (*J. Metals*, 1952, **4**, (8), 848–853).—A thermal, dilatometric, metallographic, and X-ray study of the system has been made, using (1) sponge Ti (99·7%) and electrolytic Cr (Cr 99·03, Fe 0·40, O 0·53%) for the initial detn., and (2) iodide Ti and high-purity Cr (contg. C 0·050 and O 0·045%) for refinement of points, particularly in the Ti-rich portion of the system. The alloys were prepared by melting in a He arc furnace, using a W electrode and a water-cooled Cu crucible. The specimens were homogenized by heating at 1350° C. for 4–6 hr. and quenching in water, and then given a final heat-treatment for a sufficient time (up to 670 hr. at 600° C. for the lower soln. temp.) at the temp. to be investigated in order to approach equilibrium conditions. From the results a new phase diagram has been prepared (which is in general agreement with that proposed by McQuillan (*J. Inst. Metals*, 1951, **79**, 379; *M.A.*, **18**, 759), the main features of which are: (1) a continuous series of solid soln., with a b.c.c. structure, between the two metals, which exist over a large temp. range at either end of the system, and the small temp. range of 1350°–1400° C. for the compn. 50–70% Cr; (2) a min. in the solidus at ~50% Cr and 1400° C.; (3) no evidence of the intermetallic compound Cr_2Ti_3 reported by other workers; (4) an intermetallic compound $TiCr_2$, existing between 60 and 65% Cr, and forming, during cooling, from the solid-soln. at 1350° C.; it has a f.c.c. structure, with 24 atoms/unit cell, and a parameter of 6·91 Å.; it is characterized by extreme brittleness and twinning; (5) an extremely slow eutectoid reaction in the low-Cr side of the system, whereby the β solid soln. decomposes at 15% Cr and 670° C. into an α phase and $TiCr_2$; 0·5% Cr is soluble in α-Ti at the eutectoid temp.; (6) the solubility of Ti in Cr to be ~15% at

600° C. Alloys contg. <10% Cr have an acicular, martensitic structure when quenched from the single-phase β region. Photomicrographs show the typical structures encountered. All compn. are in wt.-%. 4 ref.—E. N.

***A Theoretical Study of Young's Modulus of Binary Solid Solutions.** Yoshio Shibuya (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 645–654).—[In English]. Continuing his work (*ibid.*, 1949, [A], **1**, 161; *M.A.*, **18**, 651), devoted to the qual. explanation of the relation of Young's modulus, *Y*, and degree of order in β-brass, S. now extends the math. treatment to discuss the relation of Young's modulus to the compn. of binary solid soln. The force const. due to interaction energies between: (1) two *A* atoms, (2) two *B* atoms, and (3) one *A* and one *B* atom being denoted by ξAA, ξBB, and ξAB, resp., it is shown that: (i) when ξAB is equal to the mean value of ξAA and ξBB, a linear relation exists between *Y* and the compn. of the solid soln.; (ii) when ξAB is not equal to the mean of ξAA and ξBB, a non-linear relation exists between *Y* and the compn.; (iii) the theoretical curves derived by the use of appropriate values of the force const. agree well with experimental ones, e.g. for the Mo–W, Ag–Pd, and Au–Pd systems of alloys. Using the value of ξAA derived from experimental values of the lattice const. and compressibility, the calculated value of *Y* for polycryst. Mo is $3{\cdot}6 \times 10^{12}$ dynes/cm.^{2}, which agrees to within 10% with the experimental values $3{\cdot}22 \times 10^{12}$ and $3{\cdot}28 \times 10^{12}$ dynes/cm.^{2} obtained by Druyvesteyn (*Physica*, 1941, **8**, 439; *M.A.*, **10**, 198), and by Köster and Rauscher (*Z. Metallkunde*, 1948, **39**, 110; *M.A.*, **16**, 489). 11 ref.—J. S. G. T.

†Diffusion in Alloys and the Kirkendall Effect. J. Bardeen and C. Herring (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), 1952, 261–288).—Reprinted from *Atom Movements* (*Amer. Soc. Metals*), 1951, 87; *M.A.*, **19**, 372.

The Reduction of Metal Compounds Facilitated by Formation of Alloys, and Calculation [of Reaction]. N. G. Schmahl (*Z. anorg. Chem.*, 1951, **266**, (1/3), 1–29).—Metal compounds which are reduced with difficulty to give a pure metal among the products, may be reduced more easily by adding a second nobler metal that alloys with the decompn. product and facilitates the reaction. Examples are given of a number of reactions of this kind. From the ratio of the equilibrium decompn. pressures, values for the activity of the less noble metal in the alloy system are obtained. New results for activities in the systems Ag–Pt, Cu–Pt, Cu–Pd, Pd–Au, Pd–Ag, Pd–Pt, and Mn–Cu are given.—J. W. C.

3 — STRUCTURE

(Metallography; Macrography; Crystal Structure.)

[For all abstracts on the constitution of alloy systems, including X-ray studies, see 2—Properties of Alloys.]

***The Sub-Grain Structure in Aluminium Deformed at Elevated Temperatures.** J. A. Ramsey (*J. Inst. Metals*, 1952–53, **81**, (4), 215–216).—Sub-grain structures in coarse-grained Al deformed at elevated temp. are shown by metallographic and X-ray examination to be associated with bands similar to kink bands. A marked resemblance thus exists between these sub-grain structures and those resulting from heating after straining at room temp.—Author.

***Effect of Simultaneous Strain on Sub-Grain Growth [in Aluminium].** Jun Hino, Paul G. Shewmon, and Paul A. Beck (*J. Metals*, 1952, **4**, (8), 873–874).—A metallographic and X-ray study has been made of the effect of simultaneous strain at a low strain rate in accelerating sub-grain growth during the high-temp. annealing of cold-worked metals, an effect discovered by Wood and Scrutton (*J. Inst. Metals*, 1950, **77**, 423; *M.A.*, **17**, 904), and designated by them as "cell growth". The experiments, which have been carried out on fine-grained, high-purity Al strip, confirm the work of W. and S., and they also give information as to the min. creep strain necessary to produce the effect. 5 ref.—E. N.

***A Study of Slip Formation in Polycrystalline Aluminium.** A. E. Johnson, Jr., and S. B. Batdorf ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1951, **(2576)**, 18 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 306).—J. and B. analyse the assumptions made by Sachs, Cox and Sopwith, Taylor, and B. and Budiansky in attempts to correlate math. and phys. theories of plasticity. 2S-O Al was tested in tension, and photomicrographs (×350) at various strains were studied. The observations showed that: (1) slip occurs first in isolated grains; (2) as straining increases, there is a tendency for interaction between those grains which first suffered plastic deformation and their neighbours; (3) slip occurs on only one slip plane within a grain; (4) microscopic and macroscopic strains are not identical; (5) microscopic and macroscopic stresses are not identical; (6) results are in better agreement, however, with the assumption that microscopic and macroscopic stresses are identical than with the assumption that the respective strains are identical; (7) the resolved shear stress is not equal in all grains; (8) the experiments are compatible with, but do not necessarily verify, the assumption that plastic deformation in strain-hardening materials is primarily due to slip within the grains, because slip was first observed on the photomicrographs in which stress was considerably beyond the elastic limit.

***On the Angular Distribution of Slip Lines in Polycrystalline Aluminium.** J. M. Hedgepeth, S. B. Batdorf, and J. L. Sanders, Jr. ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1951, **(2577)**, 18 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 306).—A photomicrograph of polycryst. Al deformed in simple tension (see preceding abstract) was analysed to give the statistical distribution of the angle between tensile axis and trace of slip planes on a polished face. Comparison with the theory of B. and Budiansky shows significant disagreement, probably because the theory considers only one slip system, whereas Al has 12 systems.

***Observations on Twinning in Semi-Continuous Cast Aluminium.** K. T. Aust, F. M. Krill, and R. F. Morral (*J. Metals*, 1952, **4**, (8), 865–866).—Large amounts of twinning have been observed in semi-continuous cast Al of 99·2% purity, together with large differences of dendritic alignment in regions separated by the twin boundary. Small amounts of certain alloying addn. to the melt resulted in equiaxed grains and no twinning. The twin boundaries in the as-cast condition were free from microscopic impurities, but pptn. occurred at the boundaries after homogenization for 8 hr. at 1000° F. (540° C.) and above, being greatest after treatment at 1075° F. (580° C.). A twinned cast structure (1) required ~90% reduction to produce a symmetrical rolling structure compared to ~ 50% reduction for an equiaxed cast structure, (2) retained its non-symmetrical structure when annealed after <90% reduction, and (3) recrystallized less readily than an equiaxed structure. It is considered that the observations can be readily understood on the basis of a growth twinning process involving nucleation and preferential crystal growth with nuclei originating in "stacking faults" on c.p. octahedral planes. High-temp. gradient and increasing purity promote the formation of coarse twinning. Rapidity of solidification inherent in the semi-continuous method of casting is important, for twinning cannot be observed at slower rates of cooling. The effects of impurities or of alloying addn. in suppressing the twin formation may be connected with the adsorption of the foreign atoms on the crystal planes, thus altering the anisotropy of the melting temp. 8 ref.—E. N.

Many Mechanisms at Work Producing Grain Refinement in Light Metals [Aluminium Alloys]. Harold Bernstein (*J. Metals*, 1952, **4**, (9), 926–930).—A review. Elementary Al of high purity (99·997%) tends to have a columnar structure which may be coarse or fine, depending on the rate of cooling in the mould. The binary alloys usually have an equiaxed grain structure. On the simple basis of wt.-% alloying element needed to accomplish refinement, they may be classified into two groups: (1) where substantial amounts of alloying elements are required, i.e. the commercial casting alloys such as Al–Mg, Al–Cu, and Al–Zn, and (2) where only traces of alloying elements are used, e.g. Ti, Nb, and B; this group also includes the use of such addn. to binary and ternary Al alloys. Distinct mechanisms of grain refinement are suggested for each group, viz. by restriction of crystal growth in the former, and by active nucleation in the latter. Theories of grain restriction include: (1) Northcott's concentration-gradient theory and its appn. by Eborall and Cibula, (ii) Northcott's adsorption effect, and (iii) Bastien's surface-tension effect. Nucleation theories include: (*a*) spontaneous nucleation, (*b*) nucleation by the formation of intermetallic compounds, including the effects of superheating. The effects of vibrations are noted. It is concluded that there appear to be many mechanisms, acting singly or in concert, which are capable of producing grain refinement; future work may help to clarify the picture. 15 ref.—E. N.

***Study by Irradiation with the Atomic Pile of Intergranular and Interdendritic Segregation of Traces of Impurities in Very Pure Aluminium.** Frédéric Montariol, Phillippe Albert, and Georges Chaudron (*Compt. rend.*, 1952, **235**, (7), 477–480).—Melting at the grain boundaries of pure Al before that of the crystal mass has been attributed to imperfection of the structure at the boundary and to the diffusion of impurities towards that imperfection at a temp. close to the m.p. (*ibid.*, 1948, **226**, 1372; *M.A.*, **16**, 58). It has also been shown that irradiation will detect and determine the amount of traces in 99·99% Al (*ibid.*, 1951, **233**, 1108; *M.A.*, **19**, 671). By the introduction of a trace of radioactive Zn into the Al, dendritic segregation can also be studied. The two methods give results in good agreement. Radioactive Ru reveals the same segregation as radioactive Zn. The homogenization of very pure metals can be followed by quant. autoradiography after annealing at different temp.—J. H. W.

***Light Figures of Bismuth Single Crystals, and the Determination of Their Crystal Orientations by the Light-Figure Method. I.—Preliminary Investigations.** Mikio Yamamoto and Jirô Watanabé (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (2), 127–138).—[In English]. The orientations of crystal planes in trigonal Bi single crystals were studied by the light-figure technique previously described by Y. (*Sci. Rep. Tôhoku Imp. Univ.*, 1940, [i], **29**, 113; *M.A.*, **8**, 9). Trigonal Bi single crystals, etched with 32% HNO_3, always showed light figures corresponding to the $\{111\}$ planes, whereas light figures of pseudo-trigonal symmetry corresponding to the $\{11\bar{1}\}$ planes were produced only by artificially produced imperfect cleavage planes. Light figures corresponding to $\{10\bar{1}\}$ and $\{11\bar{2}\}$ were never obtained, but unexpected light figures corresponding to crystal planes with high-indices of the kinds $\{uu\bar{v}\}$, $\{uuv\}$, and $\{uuw\}$ occurred, the kind and number of these varying with the etching conditions. Thus light figures corresponding to $\{44\bar{1}\}$, $\{55\bar{1}\}$, $\{66\bar{1}\}$, $\{449\}$, $\{337\}$, and $\{115\}$ occurred, but changed slightly with etching conditions. The technique of the light-figure method is briefly explained. 12 ref.—J. S. G. T.

***The Effect of Certain Solute Elements on the Recrystallization of Copper.** V. A. Phillips and Arthur Phillips (*J. Inst. Metals*, 1952–53, **81**, (4), 184–208).—Using the method developed by Thorley (*ibid.*, 1950, **77**, 141; *M.A.*, **17**, 928), activation energies were calculated for the two stages postulated in Cook and Richards' theory of recrystn. (*ibid.*, 1947, **73**, 1; *M.A.*, **15**, 275) from hardness results on pure Cu and binary alloys contg. small amounts of P, Ag, Cd, As, Te, and O. The materials, of 0·010–0·018 mm. initial grain-size, were cold-rolled to 95% reduction and annealed isothermally at 35·6°–400° C. The two activation energies, although nearly equal for pure Cu and some alloys, differed considerably in, for example, alloys contg. 0·01 at.-% or less As. A variation with temp. of the slope of x (fraction recrystallized) versus log t (t = annealing time) plots and of $1/T$ (reciprocal abs. annealing temp.) versus log $t_{\frac{1}{2}}$ (time for 50% recrystn.) plots, was found for certain alloys, as predicted from the theory Approx. measurements were made of growth and nucleation rates; they varied exponentially with temp. for pure Cu. The growth rate, G, was const. with t for pure Cu and three alloys. The rate of nucleation N at 215·8° C. decreased with increase in t for a 0·021 at.-% P alloy. Log G and log N were approx. inversely $\propto$ log $t_{\frac{1}{2}}$ for all the materials. Structures were studied micrographically and by X-rays, and pole figures determined on alloys contg. 0·021 at.-% P, 0·0271 at.-% Ag, 0·018 at.-% Cd, and 0·047 at.-% As, annealed at 350° C. The effect of grain-size and of rolling reduction from 90 to 97·5% was explored for pure Cu and two alloys with P. —AUTHORS.

***Study of the Softening of Cold-Rolled β-Brass Strip by Low-Temperature Annealing and of Its Rolling Texture.** H. Asada and E. Tanaka (*Rep. Inst. Sci. Technol., Tokyo*, 1952, **6**, (1), 13–17).—[In Japanese]. In order to study the properties of cold-rolled (α + β)-brass strip, the properties of cold-rolled β-brass strip were determined. Cold-rolled β-brass strip softened even at room temp. and almost entirely softened on annealing below the recrystn. temp. By rolling and annealing, A. and T. obtained sheet with a reduction of 80% and then examined its rolling textures by means of etch pits. It had (100) [110] as the main preferred orientation, and (111) [112], (111) [110], (111) [134], and (112) [110] as secondary ones. These textures were similar to the rolling textures of other b.c.c. metals, e.g. Fe, W, and Mo. When the cold-rolled plate recrystallized, these textures became random. —AUTHORS.

On Ordering Phenomena in the Alloy Cu_3Au. L. Kholodenko (*Zhur. Eksper. Teoret. Fiziki*, 1950, **20**, (12), 1083–1097).—[In Russian]. Kh. attempts to calculate the anomalies in thermal expansion coeff., compressibility, and sp. heat, caused by the order–disorder transformation in Cu_3Au. With a Bragg–Williams treatment modified to take account of the observed change in vol. during the transformation, the calculated relation between compressibility/temp. and sp. heat/temp. curves agrees quant. with the experimental data, but an attempt to calculate the thermal expansion and degree of ordering as functions of temp. from compressibility data is unsuccessful, giving too low a value for the degree of long-range order just below the critical temp. Kh. then attempts to allow for the dependence of the lattice vibration spectrum on the degree of ordering, using a simplified version of Stepanov's approx. method (*ibid.*, 1940, **10**, 103). This modification is no more successful in explaining the high degree of ordering just below the critical temp.—G. B. H.

***The Formation of Order in the Alloy $AuCu_3$.** I. G. Edmunds and R. M. Hinde (*Proc. Phys. Soc.*, 1952, [B], **65**, (9), 716–730).—Single-crystal specimens were prepared from the melt, and examined at various stages of ordering, using both the oscillating-crystal and a special moving-film technique. These studies allowed the intensity contours of the broadened superlattice spots to be plotted in reciprocal space, after allowance had been made for instrumental broadening. A comparison with other published work was then made, and although the shapes of the spots are in agreement with previous theory, the quant. intensities are not. E. and H. then develop a short-range-order theory involving two parameters; the growth of antiphase domains which eventually touch so that the Au atoms avoid contact is regarded as a natural consequence of the extension of short-range order. —E. O. H.

The Yield Phenomenon in Polycrystalline Mild Steel. W. M. Lomer (*J. Mechanics Physics Solids*, 1952, **1**, (1), 64–73).—L. considers the variables involved in studies of Lüders bands in mild steel. The grain-size effect is reviewed, as well as the temp. of testing. The geometry of the deformation is considered, and it is shown that the size of the specimen has a large influence on the form taken by the Lüders bands. It is concluded that the boundary of the band is characterized by a pure shear deformation, and that the strong dependence of yield stress on grain-size is evidence for the grain boundaries acting as barriers to the deformation, rather than evidence for the dislocation-locking mechanism of Cottrell and Bilby (*Proc. Phys. Soc.*, 1949, [A], 62, 49), which is more applicable to single crystals.—E. O. H.

*__Surface Effects in the Slip and Twinning of Metal Monocrystals [for Lead, Tin, and Zinc].__ John J. Gilman and T. A. Read (*J. Metals*, 1952, 4, (8), 875–883).—The plastic deformation of single crystals (produced by the Bridgman method) of Pb, Sn, and Zn (of 99·999+, 99·997, and 99·999+% purity, resp.) has been studied in relation to (1) the crystal shape, and (2) the presence of evaporated and electrodeposited films of Cu, Ni, Au, and Zn on the surface of the crystals. The methods used for preparing the single crystals and the surface films and for creep and tensile testing are described in detail. The results show that: (1) crystals of Sn and Zn whose shapes have odd orders of symmetry, e.g. are triangular, sometimes twist axially when elongated in simple tension; (2) the critical shear strength of Zn crystals appears to be affected by their shape; (3) triangular and rectangular Sn crystals, when pulled in tension, develop serrated corners, the serrations consisting of minute tensile kink bands, and the affected corners always lying in planes $\sim\perp$ the plane contg. the slip direction and tensile axis; the serrations are more intense when surface films are present; (4) the plastic properties of all crystals are affected by the presence of surface films, viz. (*a*) evaporated films (~2000 Å. thick) of Cu and Ag reduce the creep rate of Zn crystals, an effect which increases with time; the stress/strain curve is also raised; (*b*) Cu films plated on to Zn crystals during creep tests, reduce the creep rate by an amount depending on film thickness; the effect is probably due to the film supporting part of the load; (*c*) all four types of plated films raise the tensile stress/strain curves of all crystals, by an amount too large to be explained by the strength of the film alone; the difference between the curves is greater at moderate strains, e.g. 10%, than at small strains; and (*d*) Cu films (~200 Å. thick) plated on to Zn crystals increase the critical tensile stress for twinning by 50–100%, and cause the deformation bands to become narrower. 29 ref.—E. N.

*__Grain Movements During Creep [of Lead–Thallium Alloy].__ R. C. Gifkins (*Nature*, 1952, 169, (4293), 238–239).—Pb alloys contg. ~5 at.-% Tl extend by large amounts (~300%) before breaking when strained slowly, the creep curve being similar to that already published (*J. Inst. Metals*, 1951, 79, 233; *M.A.*, 18, 705). X-ray photographs confirmed that no obvious recrystn. took place. No cracks occurred and very little, if any, preferred orientation was produced. The grains, which were still equiaxed after the deformation, were also much the same size, although some of the larger grains had irregular boundaries and peninsular-like extensions between other grains. It is suggested that deformation has taken place to a great extent by the relative movement of the grains.—R. S. B.

*__Deformation of Magnesium at Various Rates and Temperatures.__ J. W. Suiter and W. A. Wood (*J. Inst. Metals*, 1952–53, 81, (4), 181–184).—It is shown by X-ray and metallographic studies that polycryst. Mg, when deformed, behaves in a similar way to Al and Zn in that, at both elevated temp. and slow strain rates, a sub-grain or cell structure is formed within the grains. The work, however, brings out a new feature, namely that it appears necessary to postulate the formation of crystallite "debris" at the grain boundaries in order to correlate the X-ray and metallographic observations.—AUTHORS.

*__Physical Properties of Electrodeposited Metals. I.—Nickel: (3) The Effect of Plating Variables on the Structure and Properties of Electrodeposited Nickel.__ (Brenner, Zentner, and Jennings). See col. 324.

*__Electron-Diffraction Study of Nickel Surfaces Tarnished in the Course of Heat-Treatment in Hydrogen.__ (Ogawa and Watanabe). See col. 352.

*__The [Crystal] Structure of Nickel Boride, NiB.__ P. Blum (*J. Phys. Radium*, 1952, 13, (7/9), 430–431).—The phase diagram of the Ni–B system is still very imperfectly known, only Ni_2B and, possibly, a new phase less rich in B than Ni_2B, having hitherto been identified, although the formula and lattice are still not known. Now, single crystals of NiB have been isolated, and its lattice const. have been determined as follows: $a = 2{\cdot}925$, $b = 7{\cdot}396$, $c = 2{\cdot}966$ Å.; the vol. of the lattice cell is $V = 64{\cdot}17$ Å.^{3} and the cell contains 4 NiB mol. corresponding to a calculated $d = 7{\cdot}13$. Each metal atom is surrounded by 6 neighbours arranged at the apices of a trigonal prism, 4 at a distance of 2·584 Å., and 2 at 2·619 Å. The B atoms are arranged in zig-zag chains along the *c*-axis. Each B atom is surrounded by 6 Ni atoms, 4 at 2·177 Å. distance, and 2 at 2·632 Å. The distance between B atoms is 1·72 Å., corresponding to a normal value of the B atomic radius of 0·86 Å. The crystal structure of NiB supports B.'s theory of the directions and velocities of growth of crystals produced, as in the case of NiB, by electrolytic cementation.—J. S. G. T.

*__Oriented Chemical [Halide] Overgrowths and Surface Topography [of Silver Crystals].__ D. W. Pashley (*Proc. Roy. Soc.*, 1952, [A], 210, (1102), 354–376).—The growth of thin halide tarnish layers on Ag crystals has been studied by means of electron-diffraction reflection patterns, thus enabling the true contact planes between the Ag and the halide to be determined for many of the observed orientations of the latter. The Ag crystal specimens were prepared in various ways—rough etching, electrolytic polishing, and evaporation *in vacuo*—so that, from all the data obtained, the film orientations on their (100), (110), and (111) facets or atomically flat surfaces could be deduced. The films were grown by exposing the Ag crystals to the halogens (Cl, Br, and I) in a glass vessel at room temp. The results show that: (1) the contact planes are often not ‖ the macroscopic surface of the substrate; (2) the orientation of the halide on the (*hkl*) type Ag facets is largely independent of the position of these facets with respect to the macroscopic surface; (3) the orientation obtained on a given facet is also found to occur on a flat surface having the same indices, but varies for facets of different indices; (4) certain orientations on a given (*hkl*) facet or flat surface occur only when that facet or surface contains small irregularities (probably steps or cracks) along which the growth can originate; (5) a small % misfit between the substrate and the overgrowth lattice is not an essential condition for the appearance of the orientation of the overgrowth; and (6) some unexplained rotational (tilting) disorientations occur, as well as orientations corresponding to a large (up to 73%) misfit between the substrate and overgrowth. 10 ref.—E. N.

*__Migration of Silver Atoms on the Surface of Zinc Oxide Crystals.__ A. B. Shekhter, A. I. Echeistova, and I. I. Tret'yakov (*Zhur. Fiz. Khim.*, 1950, 24, (2), 202–206).—[In Russian]. Electron micrographs were taken of ZnO smoke deposits at magnifications of × 35,000 and × 110,000, before and after deposition of Ag by vacuum evaporation. The "conventional number of atomic layers" of Ag, calculated on the assumption that all the evaporated metal condensed on the specimen in a uniform film, varied from 25 to 60. The Ag deposits were found initially to consist of discrete particles, with dia. from <50 (resolution limit of microscope) to 1300 Å. Heating in air for 15 hr. at 150° C., followed by re-examination of the same field, produced no significant change in the structure of the deposits. At 220°–230° C., however, the coarser particles began to grow at the expense of the finer ones after 2 hr., and the process continued on further heating for 10 hr.; at 300° C. coarsening of the particles began after 1 hr. When a fresh Ag deposit was evaporated on to a specimen with coarse particles, it formed a fairly uniform film, which broke up into coarse particles on heating at 280° C. Sh., E., and T. attribute these phenomena to the high mobility of Ag atoms on Ag, and the comparatively low mobility of Ag on ZnO, so that a uniform Ag layer rapidly breaks up into discrete particles, but the particles undergo "grain growth" only on prolonged annealing.—G. B. H.

*__Note on the Effect of Oxygen and Nitrogen on the Hardness and Lattice Parameter of High-Purity Vanadium.__ (Beatty). See col. 326.

*__Solid Films on Electropolishing Anodes.__ T. P. Hoar and T. W. Farthing (*Nature*, 1952, 169, (4295), 324–325).—It has been postulated by H. and Mowat (*ibid.*, 1950, **165**, 64; *M.A.*, **18**, 33) that a compact solid film forms on an electropolishing anode. By dropping Hg on to the anode while the current was flowing, H. and F. were able to show that the Hg spread over the surface when the cell was in the etching condition, but did not do so when in the polishing condition. This they believe is direct evidence of the presence of a solid film on the electropolishing anode, which in their case was either Cu or α-brass in a 50 : 50 v./v. orthophosphoric acid–water bath.—R. S. B.

†**Solid-State Physics in Electronics and in Metallurgy.** W. Shockley (*J. Metals*, 1952, **4**, (8), 829–842).—Institute of Metals Division Lecture, 1952, consisting of an admirable exposition of the basis for the conclusion that "the met. industry will find profit in supporting fundamental research on dislocations". S. emphasizes an important feature common to both solid-state electronics (in which physics leads technology) and metallurgy (where theory lags behind practice), viz. that in both cases the important properties of the materials arise from imperfections, i.e. deviations, of the materials from perfect single crystals. The lecture is divided into four parts: (1) transistor electronics, dealing with the four types of imperfections in Ge crystals, (2) the relationship between transistor electronics and dislocation theory, (3) elements of dislocation theory, and (4) new experimental results involving dislocations—growth spirals, grain-boundary energies, grain-boundary movements, strength of small metal crystals (filamentary growths of "whiskers" produced on surfaces of Al, Cd, Ag, Sn, and Zn, and found to be thin, elongated single or twinned crystals $\sim$10,000 atoms in dia. and 10,000,000 atoms in length), and the surface hardening of Al crystals. 28 ref.—E. N.

***Preliminary Report on Researches Relating to Primary Crystallization Associated with Viscosity Change.** Tadashi Yanagihara (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 657–660).—[In English]. Bi–Sn alloy, contg. 90% Bi, heated to a definite temp. (300° C.) was allowed to drain through a capillary, in the base of an elect. furnace, and the relation between time of draining and of cooling to various temp. (230°, 220°, and 215° C.) was studied when the temp. of the melt was reduced at the rate of $\sim$3·5° C./min. All primary crystn. of the alloy occurred after holding it for $\sim$10 min., and reached the equilibrium value after $\sim$30 min. Cooling at a slower rate than 0·5° C./min. was sufficient to crystallize primary crystals in the state of equilibrium. —J. S. G. T.

***Segregation of Two Solutes, with Particular Reference to Semi-Conductors.** W. G. Pfann (*J. Metals*, 1952, **4**, (8), 861–865).—If a donor or acceptor is present alone in a solid soln. the conductivity $\propto$ its concentration; if both are present then the conductivity $\propto$ their difference in atomic concentration. If the donor and acceptor segregate at different rates during the directional solidification of the ingot, then a "*p–n*" barrier may form, in certain circumstances, at some point in the ingot. Accordingly, equations are derived, and illustrated with examples for Ge, which express the effect of segregation on: (1) the concentration, C, and the concentration gradient, dC/dl, of a single solute, with a discussion of the assumptions; (2) the difference, ΔC, between two solute concentrations and the means for minimizing its variation in an ingot; and (3) the location, g_{pn}, and the concentration gradient, $d(\Delta C)/dl$, of a *p–n* junction. The conditions under which a *p–n* barrier will form are determined; a critical ratio, r_c is found for which ΔC will go through a max. at any desired location. By locating $\Delta C_{\text{max.}}$ near the centre of an ingot, substantial uniformity of conductivity can be secured over a large portion of an ingot. 11 ref.—E. N.

***Theory of Catalysis of Nucleation by Surface Patches.** David Turnbull (*J. Chem. Physics*, 1952, **20**, (8), 1327).—A letter. Recent studies of the kinetics of solidification of supercooled Hg droplets (*ibid.*, (3), 411; *M.A.*, **20**, 21) and of cryst.-nucleus formation in supercooled liq. Sn (*J. Amer. Chem. Soc.*, 1952, **74**, (9), 2323; *M.A.*, **20**, 22) are discussed in relation to a proposal that nucleation can be catalysed by surface "patches".—J. R.

Calculation of the Internal Stress Accompanying the Phase Change of Solids. Shigeru Fujita (*J. Sci. Research Inst.* (*Tokyo*), 1952, **46**, (1274), 53–58).—[In English]. On the assumption that the grains are spherical and that transformation starts from the centres of the grains, an expression is derived for the internal stress accompanying a single-phase transformation of a solid, in terms of the modulus of rigidity of the solid in the first state, the resp. bulk moduli in the first and second states, and the resp. radii of the spherical grains in the first and second states. A simpler form of the expression for the stress is derived for the case when the transformation starts from some centres of widely separated grains.—J. S. G. T.

***(I) The Superlattice of Binary Alloys with a Large Unit Cell, and of Multiple-Component Alloys.—I.–II. (II) Note on Second-Order Superlattices for a Face-Centred Cubic Lattice.** Hiroshi Sato (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 1–13; (2), 150–159; 160–163).—[In English]. (I) [I.—] The classification and nomenclatures of types of superlattice are briefly discussed. For the interpretation of the superlattice with a large unit cell, interaction between distant atoms, as well as between nearest neighbours, must be taken into account. Various orders of sublattices, viz. first, second, &c., are defined in terms of the energy of the resp. atomic configurations. A superlattice of higher order corresponds to a lower-temp. phase, and hence transitions occur successively to lower-order stages and finally to the completely disordered state with rise of temp., although direct transition to the completely disordered stage does occur. Owing to the short range of the interaction between atoms in metals, the formation of higher-order superlattices, above the second, is frozen at a lower-ordered stage and is unrealized, if the ordering energy is solely composed of interaction energy between pairs of atoms as in the formal theory. Stabilization of such superlattices with a large unit cell is attributable to a lowering of electronic kinetic energy by splitting of the Brillouin zones corresponding to the ordering of the atoms. (I) [II.—] The theory developed in [I] is extended to treat the superlattice of multiple-component—mainly ternary—alloys. The math. treatment of ternary-alloy superlattices differs from that developed in [I] for the treatment of binary alloys, principally as a result of the necessity for the introduction of *partial* differentiation into the analysis, owing to the increase in the number of independent variables, compared with the one variable required in the discussion of binary alloys. Conditions for the discrimination of the distribution of atoms are determined not only by the signs and relative magnitudes of interactions between constituents, but also by the number of equivalent sublattices and the concentration of atoms. (II) The discussion of a second-order superlattice necessitates account being taken of the interaction between second-nearest atoms. The unit cell of the second-order superlattice for a f.c.c. lattice is described; it comprises 8 sublattices. The alloy CuPt is one of the second-order superlattices for a f.c.c. lattice, the existence of which is considered to be rather difficult to realize, according to formal theory. According to the general theory of higher-order superlattices, there may be two typical cases of second-order superlattice for f.c.c. lattices, viz. the trigonal case and the orthorhombic case. CuPt belongs to the trigonal, CuAu II may belong to the orthorhombic case (Johansson and Linde, *Ann. Physik*, 1936, [v], **25**, 1; *M.A.*, **3**, 453); the unit cells are different. The symmetry of the lattice and the lattice itself change on superlattice formation, and this deformation has an important effect on the nuclear formation of superlattice at an early stage of the transformation. 10 ref.—J. S. G. T.

Do Metals Recrystallize? Paul A. Beck (*J. Metals*, 1952, **4**, (9), 979–980).—Recent investigations on the recrystn. of cold-worked metals such as Al, Cu, Ni, Fe, Au, and brass, lead to the conclusion that in many cases (and probably always) the new grains formed on annealing in deformed metals do not grow at the expense of "deformed material", as was supposed in the traditional picture of recrystn. Not only is most of the stored energy of cold work released before "recrystn." but in many cases softening also takes place in this early stage of the annealing process. The microscopically observed "recrystn." now appears to be a process fundamentally akin to a type of grain growth ("coarsening", "discontinuous grain growth", or "secondary recrystn.") which occurs in substantially strain-free matrix with strong orientation texture; grains with an orientation strongly deviating from that of the matrix grow much larger than the matrix grains. The alternative mechanism for complete softening, viz. subgrain growth, resembles continuous or gradual grain growth, where the type of grain-size distribution and the preferred orientation do not change appreciably. In both processes the total interface area and also the interface energy/unit vol. decrease. It is concluded that the recent evidence tends to support C. S. Smith's suggestion (contribution to the discussion on a paper by B., *Physics of Powder Metallurgy* (McGraw-Hill), **1951**, p. 50; *M.A.*, **20**, 24) that the driving energy of "recrystn." is interface energy associated with the sub-boundaries. 18 ref.—E. N.

***X-Ray Diffraction Studies of [the Effects of] Cold Work in Metals.** B. E. Warren and B. L. Averbach (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 152–166; discussion, 166–172).—Cold work in a polygrained metal produces a broadening of the X-ray powder pattern lines, due primarily to distortion rather than to fragmentation; the term *frozen heat motion* is an incorrect description of cold-work distortion. A new method of Fourier analysis of reflections of X-rays from cold-worked metals, dependent upon precise spectrometer measurements of the shape of a reflection from a cold-worked sample, and of the same from an annealed sample, and designed to obtain as much information as possible from the X-ray data without making any *a priori* assumptions about the nature of cold-work distortion, is developed and discussed. For cold-worked (70 : 30) α-brass filings, X-ray data indicate that the stress is approx. isotropic. R.M.S. values of the stress averaged over various distances, *L*, for cold-worked (70 : 30) α-brass suggest the non-uniform nature of the stresses and strains in cold-worked metals. The stored elastic energy/unit vol., due to plastic deformation of α-brass is ~1·4 cal./g. 6 ref. In the discussion, *P. B. Hirsch* showed that heavily deformed Al consists of particles of mean dia. 2 μ, making mean angles of ~1° with each other. The overall *d* of dislocations in the particle boundaries is $\sim 10^9 \rightarrow 10^{10}$ cm.$^{-2}$, a result low compared with that deduced from calorimetric detn., which suggests that dislocations are present *within* the particles. (See *Nature*, 1950, **165**, 554; *M.A.*, **18**, 359).—J. S. G. T.

†**Experimental Information on Slip Lines.** W. T. Read, Jr. (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 129–146; discussion, 146–151).—Experimental information on slip lines, relating mainly to metal single crystals, is briefly reviewed but not interpreted. The investigations reported include: (1) the detn. of slip direction and slip plane; (2) the form—straight or wavy, continuous or discontinuous—of the slip lines; (3) the spacing of the various classes of slip lines and the amount of slip/line. The slip direction is always found to be the line of closest atomic spacing, but other features of slip, viz. slip plane, distribution of slip, and form of the slip lines, vary with the metal, temp., stress, and measure of loading. The influence of experimental conditions on slip lines has been studied by opt. and electron-microscope observations, multiple-beam interferometry, and photoelasticity. Classes of slip lines are distinguished. Metals discussed include hexagonal metals, Mo, Na, K, Fe, Hg, α-brass, and Al. What appears to be a single line to the eye is seen as a cluster of fine lines in a low-power microscope, and is resolved into clusters of lines by the electron microscope. Recent and current studies of cross slip and deformation bands, and the influence of surface treatment on slip lines, are discussed. 38 ref.—J. S. G. T.

*†**The Geometry of [Crystal] Dislocations.** W. T. Read, Jr., and W. Shockley (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 77–94).—Dislocations are only one of the few types of defects that occur in crystals. They are classified according to a scheme utilizing the Burgers circuit and Burgers vector. Slipping and diffusive motions of dislocations are illustrated, and a dislocation theory of straight *versus* wavy slip bands is suggested. The intersection of dislocations produces kinks or jogs in the dislocation lines, which impede the slipping motions and may create vacancies and interstitial atoms. By purely topological reasoning it is shown how an unlimited amount of slip can result from the motion of a single L-shaped dislocation line. The Frank–Read mechanism of dislocation multiplication (*Phys. Rev.*, 1950, [ii], **79**, 722; *M.A.*, **18**, 465) is discussed as a possible explanation of the occurrence of slip bands. 8 ref.—J. S. G. T.

*†**[Crystal] Imperfections from Transformation and Deformation.** C. S. Barrett (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 97–125; discussion, 126–128).—A stacking fault is produced when an atom layer in a crystal is displaced to a position destroying the regular sequence of the layers; they can be detected and analysed by X-ray-diffraction patterns made with polycryst. metals oscillating through small angles. They can also be determined, though less reliably, by X-ray line-broadening in powder patterns. Plastic deformation causes faulting in some Cu–Si, Au–Cd, Ag–Sn, and Ag–Sb alloys, and possibly in Cu–Zn, Cu, and Ag at low temp., but not in Al or the Al alloy 24S-T. Faulting accompanying transformation in Co, Li, and Li–Mg alloys is ascribed to strains due to the martensitic type of transformation. The more fundamental characteristics of crystal distortions due to plastic flow are briefly reviewed, these include elastically bent glide lamellæ, distortion around slip bands, rotational slip, deformation- and kink-bands, deformation twins, martensitic (diffusionless) transformations both spontaneous and strain-induced, grain-boundary and sub-boundary flow, and strain gradients near grain boundaries. In the discussion, *W. T. Read* suggested that hardening in the Cu–Si alloys may be attributable to an increase of slip resistance due to a fault. 45 ref.—J. S. G. T.

†**Imperfections in Nearly Perfect Crystals: A Synthesis.** Frederick Seitz (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 3–76).—The various types of crystal imperfections are discussed from a unified viewpoint. Six primary types of imperfections, viz. (1) phonons, (2) electrons and holes, (3) excitons, (4) vacant lattice sites and interstitial atoms, (5) foreign atoms in either interstitial or substitutional positions, and (6) dislocations, are discussed with a brief account of the historical introduction of each type; other kinds of imperfections may exist in materials with specialized properties, e.g. ferromagnetic materials and supraconductors. Two or more imperfections of the same or different types may interact to generate other imperfections, the whole forming a closely interlocked family. This feature of imperfections is regarded as an essential part of their pattern of behaviour, intimately related to the normal properties of the imperfections. A classified scheme of the available experimental information concerning the interaction of imperfections is given, emphasis being laid both on characteristics of imperfections and their means of generation. 97 ref.
—J. S. G. T.

†**Substructures in Crystals.** A. Guinier (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 402–436; discussion, 436–440).—Sub-boundaries divide cryst. grains into slightly disoriented sub-grains which are revealed by the opt. or electron microscope, by special etching agents, or by their property of favouring the formation of precipitates in supersaturated solid soln. or the gathering of impurities; such a structure of disoriented sub-grains has often been called mosaic structure, but the dimensions of coherent blocks of this mosaic structure are much smaller than those of sub-grains. Generally sub-grains are formed by annealing following slight deformation (polygonization), or during creep, when deformation and annealing are simultaneous. Along sub-boundaries, dislocations of the same sign are regularly spaced at rather great distances apart, and this accounts for the properties of sub-boundaries, viz. easy mobility and atomic disorder concentrated at special points. A sub-structured crystal is characterized by imperfections occurring in special planes

between rather perfect lattices. The sub-structure is a metastable stage of energy intermediate between that of the cold-worked and the recrystallized metal. At present, fragmentation of a metallic cold-worked crystal into perfect crystallites or into crystallites without too many defects, cannot be regarded as proved. Hence, a cold-worked metal cannot be regarded as an assembly of sub-grains, small but little deformed; it is more likely that such metal comprises small domains having an average orientation rather badly defined owing to internal distortions and stresses. 43 ref.
—J. S. G. T.

†**The Properties and Effects of Grain Boundaries.** Bruce Chalmers (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 441–450; discussion, 450).—A distinction is made between the properties and effects of grain boundaries; the effects of boundaries on plastic deformation are discussed, and the possibility of a non-crystallographic mode of deformation is considered. The following properties are also dealt with: strength (shear and normal) and fusion (chem. and electrochem.). 19 ref.—Author.

*†**Dislocation Models of Grain Boundaries.** W. T. Read, Jr., and W. Shockley (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 352–371; discussion, 371–376).—Grain boundaries with small angles of misfit provide a very suitable field for the appn. of dislocation theory. For grain boundaries joining two crystallites which have been relatively rotated through a small angle θ, the geometrical conditions lead to predictions which can be compared with experimental results. For any fixed arbitrary orientation, θ, of the grain boundary and axis of relative rotation of adjoining grains, the energy/unit area, E, satisfies the equation $E = E_0\theta(A - \log \theta)$, where E_0 and A are parameters dependent on the grain model; E_0 is directly calculable from values of the elastic const., and A requires, in addn., the energy of atomic disorder at the core of a dislocation. The available experimental graphs of relative E plotted against θ for Si–Sn, Sn, and Pb are in good agreement with values calculated by the formula, even for values of θ larger than are justified by the derivation of the formula. The calculated value of the abs. energy of a random grain boundary in Cu, viz. 715 ergs/cm.^{2} is 30% higher than the mean of measured values. The difference between energy and free energy is unimportant for boundaries with small values of θ. 13 ref.—J. S. G. T.

*†**Movement and Diffusion Phenomena in Grain Boundaries.** R. Smoluchowski (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 451–471; discussion, 472–475).—A theory of the mobility of grain boundaries, based upon Mott's suggestion that a grain boundary can be described as a sequence of "islands" of good and bad fit between grains (*Proc. Phys. Soc.*, 1948, **60**, 391; *M.A.*, **16**, 339) is developed. Experimental data for the mobilities of grain boundaries for Ag, Al, and brass agree with calculated values to within the expected limits of one or two orders of magnitude, taking into account the effect of surface tension in a curved grain boundary. Various phys. factors, viz. surface energy, strain, external stresses, time and temp., and relative orientation, influencing the mobility of grain boundaries, are briefly discussed and illustrated. Geometrical factors, viz. thickness of specimen, grain-size, obstacles, and impurities influencing such mobility are also discussed. The mechanism of grain-boundary diffusion is discussed on the basis of current theories of vol. diffusion. The interpretation of experimental results, e.g. those relating to grain-boundary diffusion in the Al–Cu system (Keller and Brown, *Trans. Amer. Inst. Min. Met. Eng.*, 1944, **156**, 377; see *M.A.*, **11**, 143) and the Cu–Ag system (Berman and Harrington, *Trans. Amer. Soc. Metals*, 1945, **34**, 143; see *M.A.*, **12**, 101) is discussed. Other matters discussed include: (1) the influence of relative orientation of grains and of grain boundaries upon diffusion, (2) the relationship between grain-boundary diffusion and the dislocation model, and (3) recent observations of diffusion anisotropy in a grain boundary. 30 ref.—J. S. G. T.

*†**Interphase Interfaces.** Cyril Stanley Smith (*Imperfections in Nearly Perfect Crystals* (John Wiley and Sons, Inc.), **1952**, 377–401).—Values are given for the interfacial free energy for interfaces involving one solid and a second solid or liq. phase, viz. Pb/Pb vapour, Cu/Pb vapour, Cu/Pb liq., Cu/Cu, Cu_2S liq./Cu liq., Fe_γ/Cu liq., Fe_γ/Cu_2S liq., and Fe_γ/Fe_γ. The data are fragmentary. Values of relative interfacial energies are tabulated for the Cu–Zn, Cu–Al, Cu–Sn, Cu–Sn–Pb, Cu–Sb, Cu–Ag, Cu–Si, Fe–C, Fe–Cu, Zn–Cu–Al, and Zn–Sn systems of alloys, and relative values of the interfacial tensions compared with grain-boundary energies are tabulated for the Cu–Pb, Cu–Zn–Pb, Cu–Ag, Fe–Cu, Fe–Ag, Al–Sn, Fe–Cu_2S, Fe–FeS systems. The energy between two different crystal lattices in metallic phases is almost invariably smaller than the energy of a grain boundary between two crystals of the same phase differing only in orientation. It is shown that a grain-boundary interface must have some atoms differing in co-ordination number, but that an interface between crystals differing in co-ordination number can have perfect coherence regardless of lattice parameter and orientation, and without any atoms having co-ordination different from that of either lattice or the intermediate layer. In two-phase solids, the equilibrium angle between a grain boundary and an interface boundary is important, for it determines the equilibrium microstructure. Two-dimensional lattices differing in the number of sides in the unit cell, e.g. triangular or sq. lattices, can be fitted together without "dislocations" at the boundary, although the lattices differ in orientation and spacing—a principle which may be important in the discussion of diffusionless transformations. 12 ref.
—J. S. G. T.

***Point Counter and Counter Tube in Metallographic Surface Investigations.** J. Kramer (*Z. Physik*, 1949, **125**, (11/12), 739–756).—Changes in the counting rate of a point counter or counter tube were used to investigate the nature of the metallic surface forming one of the electrodes of the counter; a point counter was found the more convenient for this purpose. When the metal surface was introduced, the counting rate fell sharply at first and then more slowly. Decay curves are given for emery-rubbed Al, Sn, Sb, Pt, and Fe. The curves were not exponential, and varied considerably with surface condition. When the metal surface was heated after reaching a steady counting rate, the counting rate rose rapidly again, but the initial rate of increase decreased with the time before raising the temp. With polished surfaces of Fe, Ni, Pt, Cr, and Sb, the counting rate suddenly increased at a characteristic temp., interpreted as the transition temp. of a non-metallic phase, above which the surface is metallic. Electrodeposited Cr, with a h.c.p. structure, gave the same transition temp. as polished Cr. Wood's metal was melted and re-solidified in the counter; there was no anomaly on heating, but a large increase in counting rate during solidification. Cu gave an increased counting rate under static or dynamic load.—J. W. C.

***An Example of Strain-Relief in Powder Specimens.** J. Gordon Parr (*J. Inst. Metals*, 1952–53, **81**, (4), 214).—The more complete strain-relief in powder than in solid specimens of an Fe–Mn alloy is demonstrated by the lower hardness value of quenched powders, and their more rapid attainment of equilibrium on tempering.—Author.

The Electron Microscope in Metallurgy. Alfred L. Ellis (*Blast Furn. and Steel Plant*, 1950, **38**, 681–686).—The techniques of the prodn. of replicas are briefly described, and the following examples of the appn. of the microscope to met. problems are given: particle-size study, quant. phase detn., and detn. of the structure of martensite.
—S. R. W.

Electron Metallography. R. D. Heidenreich (*Bell Lab. Record*, 1952, **30**, (3), 101–105).—H. briefly describes the technique of electron metallography, illustrated with micrographs of Pb–Ag alloy and Al.—D. M. D.

***On the Non-Orthogonality Problem Connected with the Use of Atomic Wave Functions in the Theory of Molecules and Crystals.** Per-Olov Löwdin (*J. Chem. Physics*, 1950, **18**, (3), 365–375).—Though the two principal quantum-mech. treatments of mol. or crystals can be based on the generally overlapping atomic orbitals, corresponding overlap integrals have been neglected in the literature. The construction of molecular orbitals when taking the overlap integrals into consideration is simplified if the orthonormalized functions given by a presented equation are considered as the real atomic orbitals; the soln. is worked out in detail for various methods of treating mol. and crystals, and it is shown that the overlap effects are responsible for the repulsive forces in ionic crystals and add interesting new features to molecular and crystal properties.—J. R.

6 — CORROSION AND RELATED PHENOMENA

***The Dissolution of Metals Over the Temperature Range 25° to −60° C. II.—The Dissolution of Cadmium in Hydrochloric Acid–Methyl Alcohol–Water Solutions.** A. B. Garrett and J. R. Heiks (*J. Phys. Colloid Chem.*, 1952, **56**, (4), 449–451).—Cf. *ibid.*, 1950, **54**, 437; *M.A.*, **19**, 5. Data have been obtained for the rate of dissoln. of Cd in methanol-water-hydrochloric acid soln. under static and dynamic conditions, in the presence and absence of a depolarizer over the temp. range 25° to −60° C. The energy of activation of the static depolarized process is in the order of 4000 cal./mole and of the non-depolarized process is ~10,000 cal./mole. The non-depolarized dissoln. process is first-order with respect to the acid concentration.—D. M. D.

Chrome Carbide Provides High Corrosion-Resistance. J. D. Kennedy (*Steel*, 1952, **131**, (5), 92–94).—Describes a new sintered product contg. Cr carbide 83, WC 2, and Ni 15%. Its properties include high abrasion and heat resistance, and high corrosion-resistance to H_2SO_4.—E. J.

***The High-Temperature Oxidation of Some Cobalt-Base and Nickel-Base Alloys.** A. Preece and G. Lucas (*J. Inst. Metals*, 1952–53, **81**, (4), 219–227).—A description is given of the oxidation characteristics of Co and Ni, and of a number of alloys based on these metals, in the temp. range 800°–1200° C. A simple apparatus was designed to supply an atmosphere similar in compn. to that produced in gas turbines, paraffin contg. 2% S being used as fuel. Reactions occurring within the scales are compared with those that take place when mixtures of oxides are heated at similar temp. Of the reactions noted, spinel formation is shown to be detrimental to the formation of a protective oxide layer. The effects of a number of minor alloying elements, viz. V, B, Nb, Be, Ti, Zr, Ca, Ta, Al, Ce, Si, and Th, on the oxidation of a Co–32% Cr alloy are described. Several elements increased the resistance to oxidation, in particular Th and Si; V and B, however, were highly deleterious, owing to the formation of low-m.p. oxides. The rate of oxidation of Co shows a sharp decrease in the region of 950° C., and this coincides with the upper limit of stability of Co_3O_4. It appears that the presence of this oxide at the outer surface of the scale increases the rate of O transfer to the underlying CoO. From a consideration of scale structure and the occurrence of internal oxidation in a number of the alloys, it is suggested that O diffusion through the scale occurs to a considerable extent.—AUTHORS.

***Corrosion-Resistance of Copper Alloys Containing Chromium [and Nickel].** Tetsutaro Mitsuhashi and Manabu Ueno (*J. Mech. Lab.* (*Tokyo*), 1950, **4**, (2), 50–53).—[In Japanese]. A report on the corrosion-resistance of Cu-base alloys contg. Cr 5–25 and Ni 5–25%.—AUTHOR.

***Scaling of Lead in Air.** Elmer Weber and W. M. Baldwin, Jr. (*J. Metals*, 1952, **4**, (8), 854–859).—A study has been made of the oxidation of Pb (~99·94%) at temp. of 250°–800° C., i.e. covering both the solid and the quiescent liq. metal, X-ray analysis being used to determine the nature of the oxide films formed. Solid Pb, between 254° and 321° C., obeys a single parabolic law of wt. increase/time, the scale formed being β (red, tetragonal) PbO. Liq. Pb scales according to three successive parabolic relations, the time of transition becoming shorter with increasing temp. The first parabola has a low const.; the scale (sooty black in colour) is thin and yields diffraction lines of Pb_3O_4, Pb, and α (yellow, orthorhombic) PbO. The second parabola has a much higher const.; the scale contains Pb_3O_4, and β- or (α + β)-PbO, according as to whether the temp. is < or > 486° C. (the equilibrium temp. of β- and α-PbO), resp. The third parabola has a const. only slightly lower than that of the second; the scale consists of β-PbO if formed below 486° C. or (α + β)-PbO if formed above. The results are discussed in relation to those of Gruhl (*Z. Metallkunde*, 1949, **40**, 225; *M.A.*, **17**, 418), but no mention is made of the more recent work of Hofmann and Mahlich (*Werkstoffe u. Korrosion*, 1951, **2**, 55; *M.A.*, **19**, 603). 18 ref.—E. N.

***Electron-Diffraction Study of Nickel Surfaces Tarnished in the Course of Heat-Treatment in Hydrogen.** Shiro Ogawa and Denjiro Watanabe (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 41–47).—[In English].—Ni plates for use in vacuum tubes, annealed in H at ~900°–1000° C. often tarnish and acquire a milk-white appearance, which considerably reduces their commercial value. This tarnishing and the milk-white appearance have been shown by electron-diffraction studies to be due to the presence of silicates of light metals. Silicates, e.g. $(Mg, Ca)SiO_3$ and Mg_2SiO_4, must be removed during melting, as if even a small amount remains dispersed throughout the material, it may gradually collect in the grain boundaries during heat-treatment of the Ni and cause tarnishing of the surface.—J. S. G. T.

Material Problems in Chemical Process Equipment [Use of Tantalum and Titanium]. L. Piatti (*Sulzer Tech. Rev.*, 1952, (1), 21–34).—In a general description of the problems of materials for the chem. industry, the corrosion and erosion attack of ferrous and non-ferrous alloys is discussed. Included in the survey are details of the use of Ta and Ti.—E. J.

***Studies in the Corrosion of Metals Occasioned by Aqueous Solutions of Some Surface-Active Agents. III.—Tin.** T. K. Ross (*J. Appl. Chem.*, 1952, **2**, (9), 526–531).—Cf. Holness and R., *ibid.*, 1951, **1**, 158; *M.A.*, **19**, 36. [Part II deals with mild steel.] The action of dil. aq. soln. of representative anionic, cationic, and non-ionic wetting agents on sheet Sn has been studied, using (1) distilled water and (2) London tap-water with and without Zeolite softening treatment. The effects of varying the temp. and *p*H of the soln. were also investigated. Most of the anion-active agents were corrosive towards Sn, exceptions being sulphonated castor oil, which was practically inactive, and Na alkylarylsulphonate, with which attack was very slight. With two of the cationic compounds moderate corrosion, increasing with increasing concentration, was noted, but no attack on Sn immersed in triethanolamine soln. was observed, though a precipitate of $Mg(OH)_2$ was deposited on the Sn when these soln. had been made up with tap-water. The sample of the non-ionic group examined produced no measurable attack, but the whole surface of the Sn specimens was in many cases covered with coloured films after immersion. In general, tap-water soln. were less corrosive than those in distilled or softened water, and the effects of *p*H changes were very variable. The results are graphically expressed, and a mechanism is suggested.—J. R.

***Study of [Corrosion of] Zinc Casting Alloys.** Kazuo Katori and Kingo Naoki (*J. Mech. Lab.* (*Tokyo*), 1952, **6**, (1), 26–30).—[In Japanese]. Cf. *ibid.*, 1950, **4**, 268. An investigation of cracking phenomena on test-pieces exposed to air for 3 years is reported. Results differed considerably from those obtained in steam tests for 100 hr., and the difference increased with poorer qualities of alloy. Mg content of the alloy had little effect.—AUTHORS.

[Corrosion-Resistance of] Metals for High-Pressure Hydrogenation Plant. G. A. Nelson (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (2), 205–211; discussion, 211–213).—After exposure for about one year to a temp. of 500° C. and a pressure of 2840 lb./in.^{2} in an NH_3 catalyst chamber, the Hastelloys

A, B, C, and D were found to be completely unaffected by either H or N. Brass is quite resistant to H and S attack, although no use is made of this alloy. It is considered that Zn acted as the catalyst to provide protection against H_2S. N. deals mainly with the various steels used in this type of plant.—H. Pl.

***Corrosion–Erosion at Boiler Feed Pumps and Regulating Valves at Maryville. Second Test Programme.** J. M. Decker, H. A. Wagner, and J. C. Marsh (*Trans. Amer. Soc. Mech. Eng.*, 1950, **72**, (1), 19–24; discussion, 24–28).—Test results obtained for leaded bronze indicated that it is not well suited for corrosion–erosion service at temp. from 250° to 400° F. (120°–205° C.). After six months at 400° F. leaded bronze rings in a coil-drains pump were badly corroded in comparison with 12% Cr steel rings. Indications are that even a thin coating of Cr on the interior surface of C steel pump casings might provide excellent protection for pumps already in service. Navy M bronze is satisfactory at temp. up to $\sim$320° F. (160° C.)—H. Pl.

Analysis of Some Corrosion Problems in Petroleum Refineries. J. F. Mason, Jr. (*Trans. Amer. Soc. Mech. Eng.*, 1952, **74**, (3), 305–312; discussion, 312–313).—15 corrosion problems have been presented for consideration. Where mineral acids or chlorides may be present, as well as alkaline soln. at elevated temp., the trend is towards the use of highly-alloyed materials such as Monel and Ni together with the Cu-base alloys such as 70 : 30 cupro-nickel, Admiralty brass, and Muntz metal.
—H. Pl.

***A Standard Laboratory Corrosion Test for Metals in Phosphoric-Acid Service.** H. F. Ebling and M. A. Scheil (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (10), 975–987).—Where high concentrations of H_3PO_4 are handled Hastelloys A, B, C, and D, Ta, Chlorimet 2, La Bour R55, and sterling Ag have proved satisfactory metals. However, in a catalytic process using 65% H_3PO_4 under pressures of 1000–1350 lb./in.^{2} in the range 500°–600° F. (260°–315° C.) only Ag, Cu, and some Cu alloys were satisfactory. In three processes where H_3PO_4 was the principal corroding agent, there was little agreement in selecting a good corrosion-resistant material. The reasons given for disagreement include: (1) concentration, temp., and pressure of the H_3PO_4 soln. influence the corrosion-resistance of the material; (2) small quantities of reducing mineral acids and the occurrence of formation of organic acids influence corrosion-resistance of the material; (3) stagnant liquids or solids in contact with the surface of the material can cause localized attack or pitting; and (4) agitated or flowing soln. of H_3PO_4 can cause a serious problem of erosion together with corrosion.—H. Pl.

Chemical Causes of Wear. J. J. Broeze (*Engineering*, 1952, **173**, (4505), 693–694).—Presented at a Conference on Abrasion and Wear, held at Delft, November 1951. B. deals with three typical fields in which wear of mating metallic surfaces may be attributed to chem. attack: (i) attack by the atmosphere (oxidation), (ii) attack by lubricants or their derivatives, and (iii) attack by combustion products encountered in internal-combustion engines. 9 ref.
—D. K. W.

Corrosion: Design It Out! R. B. Mears (*Canad. Metals*, 1952, **15**, (6), 24, 26).—It is considered that the most effective method of reducing corrosion is by the use of appropriate design.—W. A. M. P.

Corrosion: Causes and Prevention. (Speller). See col. 382.

7 — PROTECTION

(Other than by Electrodeposition)

The Anodic Oxidation of Aluminium. G. C. Thompson (*Electronic Eng.*, 1952, **24**, (292), 281–283).—The theory of anodizing is discussed, and details of the chromic-acid process (Bengough–Stuart process) and the sulphuric-acid process are given. The chromic-acid process is unsuitable for use with Al alloys contg. $>5\%$ Cu, but the corrosion-resistance of the film produced by it is better than that produced by the sulphuric-acid process. The sulphuric acid oxide film is transparent, whereas the chromic acid film is not. Notes are given on the final treatment, multichrome dyeing, and a method of removing patchy and streaky films before re-anodizing.—H. Pl.

***The Connection Between the Continuity of the Oxide Films and the Surface Cleanliness of Aluminium.** A. V. Shreider (*Zhur. Fiz. Khim.*, 1950, **24**, (4), 455–458).—[In Russian]. Specimens of Al contg. 0·3% Si + Fe, of alloy AMG (Mn 0·24, Si 0·3, Mg 2·51%, remainder Al), and of alloy AK-4 (Ti 0·07, Cu 2·4, Mn 0·16, Si 0·29, Mg 1·52, Ni 1·41, Zn 0·21, Fe 1·19%, remainder Al) were anodized in 3% CrO_3 soln. at 36° C., with a p.d. raised from 0 to 40 V. in 15 min., then kept const. for 45 min., and a c.d. of 0·6–1·5 amp./dm.^{2}. This treatment produced less increase in surface roughness than treatments in stronger CrO_3 soln., in 5% $(COOH)_2$ soln., or in 20% H_2SO_4 soln. Specimens of all the alloys were polished, finished with emery paper, and sand-blasted (R.M.S. roughnesses, measured with a Brash profilometer, ranged from 7 to 200 μ-in.) before anodizing. Increased surface roughness was in all cases completely correlated with decreased resistance to the V.I.A.M. drip corrosion test (Akimov, Tomashov, and Tyukina, " Investigation of the Anodic Treatment Process for Aircraft Parts in Aluminium Alloys " and " Development of a More Rapid Process for Anodic Oxidation of Aluminium Alloys in Sulphuric Acid ", Moscow and Leningrad, **1946**), with increased loss of weight on anodizing and with a more positive electrode potential, as measured by a null method in 0·5*N*-NaCl soln. Sh. shows that the last effect can be deduced from the assumption that the anodic film on rough specimens is cracked by the changes in vol. during its formation.—G. B. H.

***Air Agitation: Its Effect in Sulphuric Acid Anodizing.** R. C. Spooner (*Metal Ind.*, 1952, **81**, (13), 248–250).—S. summarizes the advantages and disadvantages of air agitation in the anodizing bath; describes experiments on the c.d., film thickness, and anodizing temp. in the H_2SO_4 anodizing of pure Al and Al alloys; and discusses the results. It is considered that air agitation would have similar effects with other electrolytes, such as chromic or oxalic acid soln.
—J. H. W.

Vitreous Enamel Finishes for Aluminium. J. C. Bailey (*Light Metals*, 1951, **14**, (165), 647–656).—A survey of researches into the formulation and appn. of vitreous finishes for Al ware. B. enumerates desirable properties (hardness, strength, durability, non-toxicity, good appearance, low m.p., high coeff. of expansion, adhesion to base). Compn. of Pb-bearing and of Pb-free enamels are tabulated, and methods of prepn., appn., firing, and testing are described; the results of a series of tests developed by B. and Whitaker are tabulated for various enamels on a base of commercially pure Al (*ibid.*, 1949, **12**, 139; *M.A.*, **17**, 725). Future research into optimum compn., prepn. and appn. of frit, factors governing adhesion, and possible alternatives to firing is desirable.—P. R.

Hot-Dip Aluminizing Affords Protective Surface. J. B. Russell (*Steel*, 1952, **131**, (2), 100, 102).—The coating of steel with Al and an Al–Be–Si alloy by a continuous process is described in detail. It is claimed that the process will be used to an increasing extent in the future.
—E. J.

***On the Part Played by Chromous Chloride and Hydrogen in the Process of Thermal Chromium Coating.** N. G. Klyuchnikov (*Doklady Akad. Nauk S.S.S.R.*, 1950, **74**, (2), 267–269; *C. Abs.*, 1951, **45**, 476).—[In Russian]. A layer of Cr metal was produced on steel by heating for 3 hr. at 800°–850° C. in an atmosphere of $CrCl_2$, formed by reduction of $CrCl_3$ with H, and with a const. supply of H. Fe samples treated in this way showed a gain of wt. greater, for equal surface area, than after 4 hr. at 900° C. in a powder mixture of ferrochrome with 1% $CrCl_3$. This gain in wt. precludes the interpretation of the process on the basis of the reaction $CrCl_2 + Fe \rightarrow Cr + FeCl_2$ which would result in a loss of wt. Rather is the process due to the reaction $CrCl_2 + 2H \rightarrow Cr + 2HCl$, which involves no loss of Fe. Proof of this is provided by the considerable gain of wt. observed after treatment for 4 hr. with $CrCl_2$ at 900° C. of steel coated with electrolytic Fe, and especially of steel coated with electrolytic Cr. In the latter case, the thickness of the thermally produced additional Cr layer was 5–10 times greater than in the thermal Cr coating of H-free Fe. This is obviously due to the H occluded in the electrolytic Fe or Cr.

***Spectrographic Analysis of Metallic Surfaces, in Particular Those Produced by Chromizing.** (Malamand). See col. 364.

***Electrolytic Determination of Tin and Tin-Iron Alloy Coating Weights on Tinplate.** C. T. Kunze and A. R. Willey (*J. Electrochem. Soc.*, 1952, **99**, (9), 354–359).—Sn can be stripped anodically in 1*N*-HCl soln. with full current efficiency. The action does not stop at the Sn–Fe alloy layer. The voltage changes that occur indicate two end-points, from which the free and alloyed Sn may be determined. Apparatus with a specially designed D.C. power source for recording the voltage changes is described. This measurement of alloy thickness is claimed to surpass in reliability and speed all other methods. 6 ref.—G. T. C.

Galvanizing by the Sendzimir Process. K. Oganowski (*Blast Furn. and Steel Plant*, 1950, **38**, 912–916).—See *Products Finishing*, 1950, **14**, (11), 42; *M.A.*, **18**, 190; *Metal Finishing*, 1950, **48**, (10), 63; *M.A.*, **18**, 372.—S. R. W.

Metallic Surface Coatings for Grey Iron. Charles O. Burgess (*Product Eng.*, 1950, **21**, (12), 108–113).—Methods of applying coatings to grey Fe for various appn. are outlined, including spraying, hot dipping, hard-facing by welding, cementation, and electrodeposition.—M. A. H.

Ceramic Coatings for Metals Offer Protection at High Operating Temperatures. —— (*Ceramics*, 1951, **3**, (30), 321–323).—The appn. of ceramic coatings to metals for increasing the heat-resistance of the metals, particularly ZrO coatings on Mo, are discussed. Tests on Mo coated mainly with ZrO, by either dipping or spraying with slips and firing at 2150° F. (1180° C.) in an inert atmosphere, have shown that oxidation of the metal at high temp. (900° C.) is greatly retarded.—S. R. W.

Recent Ceramic Coatings for High-Temperature Alloys. —— (*Ceramics*, 1952, **3**, (35), 581–582).—The method of applying ceramic coatings to heat-resistant alloys in order to protect them from scaling and intergranular corrosion is briefly described. Compn. of ceramics recently developed for this purpose are given.—S. R. W.

Gespritzte Metallüberzüge. (Reininger). See col. 383.

---

8 — ELECTRODEPOSITION

***On the Barrel Plating Method for Chromium. I.—The Plating Solution.** Matsuhei Kishi (*J. Mech. Lab.* (*Tokyo*), 1950, **4**, (6), 232–235).—[In Japanese]. The fluoride and sulphate baths for Cr electrodeposition have been examined and compared. It is found that: (i) the fluoride bath gives better throwing power and has a greater range of c.d. for bright deposition; (ii) the range of c.d. for bright deposits in a fluoride bath becomes higher in value but the width remains almost const., whereas the range of c.d. for bright deposits in the sulphate bath is very narrow at room temp. but increases rapidly with increase of temp.; and (iii) a better plated surface is obtained in a fluoride bath by using intermittent currents.—Author.

Hard Chromium Plate to Improve the Corrosion-Resistance of Tool Steel. H. E. Ricks (*Plating*, 1952, **39**, (9), 1029–1030).—It is shown that the corrosion protection afforded by Cr plating $\propto$ the thickness of the Cr and the surface smoothness of the steel before plating. For moulds, it is recommended that the surface be given a finish of 5 μ-in. and that the Cr thickness be at least 0·0005 in.—G. T. C.

Rebuilding Engine Cylinders and Liners: Worn Machine Surfaces Built Up with Iron and Porous Chromium. Fred M. Burt (*Metal Finishing*, 1952, **50**, (8), 67–72).—A factory where Fe and Cr plating of engine cylinders and liners is carried out is described. No information is included on the compn. of soln. used.—G. T. C.

***Effect of Chromium Plating on the Plastic Deformation of SAE 4130 Steel.** Hugh L. Logan (*J. Research Nat. Bur. Stand.*, 1951, **46**, (6), 472–479; and (abridged) *Plating*, 1951, **38**, (7), 701–702; *Metal Finishing*, 1951, **49**, (8), 53–60).—Tensile, tensile impact, bending, and crushing tests were made on Cr-plated steel in the as-plated condition, and after baking for several hr. at 200° C. The Cr plate had nominal thicknesses of 0·0001, 0·001, 0·010, and 0·015 in. The plating reduced the plastic deformation preceding fracture in all tests except the impact test. In tensile tests the elongation was reduced to as little as 55% of the value for unplated steel, and comparable reductions occurred in the true stress at the beginning of fracture, and in the decrease in cross-sectional area before fracture. Tensile and yield strengths decreased as the plating thickness increased, but were >90% of the values for untreated steel in all specimens examined. Baking at 200° C. increased the ductility of the plated steel.—J. W. C.

***Radioactive-Isotope Dilution Method for Determining Sulphate Concentration in Chromium-Plating Baths.** Stanley L. Eisler (*Plating*, 1952, **39**, (9), 1019–1023, 1031).—A method using the isotope-dilution technique is described. It is shown to be more accurate and quicker than other methods. In addn., it dispenses with weighing operations and incorporates automatic counting. The results obtained are accurate to within 1–3% of the correct values. An average deviation of 2% was found from the mean of a series of seven samples run on two different occasions. 7 ref. —G. T. C.

***Electrochemistry of Fluoride Solutions. II.—Electrodeposition of Copper from Fluoride Solutions.** A. N. Kappanna and E. R. Talaty (*J. Indian Chem. Soc.*, 1951, **28**, (9), 523–529).—Cf. *ibid.*, (8), 413; *M.A.*, **20**, 357. An account is given of some experimental work on the electrolysis of CuF_2 soln. The Cu deposits on polished Pt and Cu cathodes are adherent, ductile, and fairly lustrous. It is concluded that the fluoride bath can give good deposits, but a limitation is the low max. c.d. used. The deposits obtained are much brighter than those from the acid sulphate bath.—W. A. M. P.

White-Brass Plating at Hamilton Mfg. Corp. C. C. Weekly (*Metal Finishing*, 1952, **50**, (9), 67–69).—Brief general information is presented, but no details are included on soln. compn.—G. T. C.

***Effect of Impurities and Purification of Electroplating Solutions. I.—Nickel Solutions. (5). The Effects and Removal of Zinc.** D. T. Ewing *et al.* (*Plating*, 1952, **39**, (9), 1033–1037). —Cf. *ibid.*, 1950, **37**, 1157; *M.A.*, **18**, 377. A report of work carried out under Research Project No. 5 of the American Electroplaters' Society. The experimental work is described in full detail. The following conclusions are reached: there is no change in appearance of coatings from a Watts bath (*p*H 2·2) until the Zn concentration is 300 mg./l. Beyond this there is a tendency to brighten the plate. In a Watts bath (*p*H 5·2) concentrations of Zn from 10 to 300 mg./l. cause darkening of the plate. In an alloy-type bright-Ni bath the presence of Zn decreases the brightness of the deposit. In a bath contg. organic brighteners Zn in concentrations of 150–300 mg./l. enhances the brightness. Deposits from all baths were darkened in very low c.d. areas (10 amp./ft.^{2}) by concentrations of Zn as low as 10 mg./l. Zn present up to 300 mg./l. causes no change in the adhesion of the deposits from any bath. In the Watts-type bath, increase in

Zn results in decrease in deposit ductility. No appreciable change was noted in the ductility of bright Ni when Zn was present up to 150 mg./l., but at 300 mg./l. the deposit from the Ni–Co bath is very brittle and highly stressed. A slight decrease in throwing power over the range 10–300 mg./l. Zn was noted in the Watts bath (pH 2·2). In the other baths there was a general tendency for throwing power to increase with Zn concentration over this range. In all cases, increase in Zn concentration improved the corrosion-resistance of the deposits, particularly in the case of bright plate. There was an increase in hardness with increase in Zn concentration in the case of plate from the Watts baths (pH 2·2 and 5·2) and of the bright Ni soln. contg. organic brighteners. At certain concentrations of Zn there was a slight decrease in hardness in the case of alloy bright Ni baths. Zn can be removed from Ni-plating soln. most economically by electrolysis at optimum c.d. of 2–4 amp./ft.^{2}. This method permits a reduction in the concentration of Zn to below 1 mg./l. Concentrations of Zn below 250 mg./l. cannot be removed from Ni soln. by the high pH pptn. method without appreciable loss of Ni. 7 ref.
—G. T. C.

***Physical Properties of Electrodeposited Metals. I.—Nickel: (3) The Effect of Plating Variables on the Structure and Properties of Electrodeposited Nickel.** (Brenner, Zentner, and Jennings). See col. 324.

The Production of Cast Nickel Anodes. Edmund R. Thews (*Metal Finishing*, 1952, **50**, (9), 70–74, 77).—Full details are presented of the advantages and disadvantages of this type of anode, of impurities likely to be present, of gas absorption, of melting, deoxidizing, casting, and cleaning the anodes. Comments are made on the remelting of scrap anodes. 31 ref.
—G. T. C.

Material Control in Nickel/Chrome Plating. Robert T. Hood (*Metal Finishing*, 1952, **50**, (9), 64–66).—Information is presented which is intended to assist in estimating and standardizing costs in Ni and Cr plating.—G. T. C.

***Determination of Boric Acid in Nickel and in Zinc Electroplating Baths.** A. Wogrinz and G. Kudernatsch (*Prakt. Chem.*, **1950**, 197–199; *C. Abs.*, 1951, **45**, 2368).—To determine H_3BO_3 in Ni-plating baths, add 30 ml. conc. NH_4OH, 5 g. $(NH_4)_2SO_4$, and 2 drops of conc. H_2O_2. Filter off any $Fe(OH)_3$ formed and remove the Ni by electrolysis. Then add Bromocresol Purple indicator and neutralize with HCl. Heat, cool, neutralize again, and add 100 g. of invert sugar syrup to activate the H_3BO_3 so that H^+ can be titrated. In baths contg. Zn, a similar procedure is fairly successful without removing Zn, but many baths contain alum and the Al must be removed as hydroxide before the titration is made. An excellent procedure for determining H_3BO_3 consists in distilling off $(CH_3)_3BO_3$ by treating the sample with $CH_3OH + H_2SO_4$ and catching the distillate in NaOH soln. Then the H_3BO_3 can be titrated as above by making neutral to Bromocresol Purple, activating with invert sugar soln., and titrating with NaOH.

***Electrochemistry of Fluoride Solutions. I.—Electrolysis of Silver Fluoride Solutions.** A. N. Kappanna and E. R. Talaty (*J. Indian Chem. Soc.*, 1951, **28**, (8), 413–424; errata, (9), 529).—An account of some experimental work on the electrolysis of AgF soln. A cryst. but adherent Ag deposit is obtained on Pt, and the addn. of NH_4F gives lustrous deposits with a fine, frosted appearance. Smooth, matt deposits can be obtained only with the aid of certain addn. agents, such as: (1) currant essence, (2) boric acid with citronella oil, and (3) NH_4F with clove oil. In no case are the deposits obtained as good as those from cyanide baths.—W. A. M. P.

***The Electrodeposition of Silver on an Iron-Nickel Alloy for a Glass-to-Metal Seal Arrangement.** R. T. Foley, R. D. Alvord, and J. K. Easley (*J. Electrochem. Soc.*, 1952, **99**, (9), 349–353).—A method is described for Ag-plating a 58 : 42 Fe–Ni alloy to produce a deposit capable of withstanding an anneal of 850° C. No Cu or Ni undercoating is used to achieve an adherent coating. The diffusion of O through the Ag plate is hindered by a thin coating of In, which also facilitates the glass-to-metal seal. Full details are given of two methods of preparing the alloy before plating. One involves pickling in an acid mixture contg. by vol.: HCl (37%) 1, HNO_3 (70%) 1, H_2SO_4 (96%) 1, H_2O 9 parts for 30 sec. at 90°–95° C., followed by a dip for 15–20 sec. in a mixture of H_2SO_4 2, HNO_3 1 part and then a dip in KCN soln. (50 g./l.). The second method involves cathodic treatment at room temp. in a 10% soln. of H_2SO_4 for 15 min. and c.d. 2·6 amp./dm.^{2}. The Ag plating is carried out from a bath contg. KCN 42, K_2CO_3 45, AgCN 30 g./l. C.d. is 0·4 amp./dm.^{2} at room temp. The properties of the plate produced are briefly described.
—G. T. C.

Late Developments in Electrodeposition Techniques. William Blum (*Canad. Metals*, 1952, **15**, (6), 52, 54).—B. discusses recent developments in Cr plating, testing of plated coatings, and plating from non-aq. soln.—W. A. M. P.

***Quantitative Measurement of Adhesion of Electrodeposited Metals.** H. C. Schlaupitz and W. D. Robertson (*Plating*, 1952, **39**, (7), 750–753, 764; (8), 862–864, 932).—The principal features of an ideal adhesion test for electrodeposits are discussed and four different types of known quant. adhesion tests are examined in the light of these features. Some experiments using Knapp's modification of the Ollard test are described. Further tests were carried out using a slightly modified Knapp fixture. All these tests tended to confirm previous evaluations of this and other Ollard-type test methods. It is concluded that the problem of evaluating the adhesion of electrodeposits is analogous to that of measuring the fracture strength of metals in the absence of appreciable plastic deformation. It is necessary to design a test specimen incorporating: (1) symmetry across the interface in question, (2) absence of local stress concentration, (3) a min. base-metal and deposit thickness to limit plastic deformation, and (4) a means for transmitting the load to the interface that subjects the base metal and deposit to the stress state of high triaxial tension. The present work shows that it is possible to cause non-ductile fracture in annealed Cu foil when the design of the test specimen is such that a triaxial state of tensile stress is obtained. The method may be adapted to the evaluation of the adhesion of electrodeposits in a manner which conforms with the above four principles. 18 ref.
—G. T. C.

Surface-Tension Control in the Plating Room. Joseph B. Kushner (*Products Finishing*, 1952, **16**, (12), 32–36).—The significance of surface tension in plating and allied operations is described, and details are presented of a simple instrument for measuring it in the plating shop.—G. T. C.

Rinsing [in Electroplating Operations]. H. L. Pinkerton (*Plating*, 1952, **39**, (9), 1016–1017, 1031).—Equations are derived to express the limiting average concentration attained in a rinse tank at equilibrium, the range of concentration fluctuation during a rinsing cycle at equilibrium (this depends on tank vol.), and the relationship between rinse tank concentration and time. It is shown that the size of the rinse tank does not influence the equilibrium concentration and that the time required to reach equilibrium is directly proportional to rinse tank vol. and inversely proportional to rate of rinse water flow. 7 ref.—G. T. C.

Heating Systems for Plating Vats. —— (*Product Finishing* (*Lond.*), 1952, **5**, (2), 48–56).—The advantages and disadvantages of various methods of heating electroplating soln. are discussed.—H. A. H.

Motor Generators for Electroplating. E. G. Schroeder (*Metal Finishing*, 1952, **50**, (9), 75–77).—Information is presented on the types available and on their characteristics.
—G. T. C.

High Operating Efficiency in Job-Lot Finishing of Small Parts. Frank L. Bonem (*Products Finishing*, 1952, **16**, (11), 18–28, 30).—Brief information is presented on the compn. and operating conditions for various plating soln. and cleaning baths used for finishing small parts.—G. T. C.

A Course in the Principles and Practices of Electroplating. Louis Serota (*Metal Finishing*, 1952, **50**, (8), 73–76).—The planning and operation of a two-year course is described.
—G. T. C.

9 — ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

The Lead Dioxide/Cadmium Reserve-Type Cell. L. E. Pucher (*J. Electrochem. Soc.*, 1952, 99, (8), 204C–205C).—Cd anodes are used in these cells and either Pb, Pb-alloy, or Cu grids are used for the cathode.—G. T. C.

***The Existence of a Limit to the Spontaneous Discharge of Copper Ions on Metallic Zinc.** (Mme) Marguerite Bancie-Grillot and Edmond Grillot (*Compt. rend.*, 1951, 233, (22), 1362–1364).—By the use of the radioactive tracer Cu^{64}, it has been found that there is a limit to the spontaneous discharge of Cu^{++} on Zn. This limit makes it impossible to deposit Cu from a boiling 10% $ZnSO_4$ soln. contg. $<5 \times 10^{-6}$ g. Cu/g. Zn.—N. B. V.

Magnesium Primary Batteries. C. K. Morehouse (*J. Electrochem. Soc.*, 1952, 99, (8), 187C–189C).—The use of Mg in this appn. is reviewed. 31 ref.—G. T. C.

High-Capacity Magnesium Dry Cells. R. C. Kirk, P. F. George, and A. B. Fry (*J. Electrochem. Soc.*, 1952, 99, (8), 323–327).—Mg dry cells, similar in type to the Leclanché cell, appear to be suitable for a number of appn. The anode material is a Mg alloy contg. Al 3, Zn 1, Mn 0·2, and Ca 0·15%. Details are presented on the construction and characteristics of such cells.—G. T. C.

The Cuprous Chloride/Magnesium Reserve Battery. L. E. Pucher (*J. Electrochem. Soc.*, 1952, 99, (8), 203C–204C).—Mg of at least commercial purity is used as the anode in this battery and Cu or Cu alloy grids or foil are used for forming the cathode.—G. T. C.

Silver Chloride/Magnesium Reserve Battery. Ivan C. Blake (*J. Electrochem. Soc.*, 1952, 99, (8), 202C–203C).—The use of high-purity Mg as anode and of Ag as cathode in this battery is described. 6 ref.—G. T. C.

Nickel/Cadmium Batteries: Pocket Type. Sven Bergstrom (*J. Electrochem. Soc.*, 1952, 99, (9), 248C–250C).—The construction and characteristics of these batteries are described. —G. T. C.

Sintered-Plate Nickel/Cadmium Batteries. G. B. Ellis, H. Mandel, and D. Linden (*J. Electrochem. Soc.*, 1952, 99, (9), 250C–252C).—The construction and elect. characteristics of these batteries are described.—G. T. C.

The Edison Nickel/Iron/Alkaline Cell. Fayette C. Anderson (*J. Electrochem. Soc.*, 1952, 99, (9), 244C–248C).—The construction and operating characteristics of this type of cell are described. 13 ref.—G. T. C.

***Investigation of the Overvoltage During the Evolution of Oxygen at a Nickel Electrode in Alkaline Solution.** V. N. Fiseisky and Ya N. Tur'yan (*Zhur. Fiz. Khim.*, 1950, 24, (5), 567–573).—[In Russian]. F. and T. investigated the O overvoltage on a pure Ni anode in 7·5*N*-KOH soln., purified by prolonged electrolysis, at temp. of 25° and 80° C. and c.d. of 5×10^{-5}–0·3 amp./cm.^{2}. An HgO comparison electrode was used, and the anode and cathode compartments were not separated during measurement. The curve of overvoltage against log of c.d. can be divided into three parts; at c.d. of 0·3–0·05 amp./cm.^{2} it is a straight line of slope $2{\cdot}3(RT/0{\cdot}32F)$, at c.d. of 0·05–$10^{-3}$ amp./cm.^{2} it is somewhat steeper, and at c.d. of 10^{-3}–5×10^{-5} amp./cm.^{2} it is a straight line of slope $2{\cdot}3(RT/2F)$. In the highest range of c.d. the surface of the anode was covered with pink NiO_2, but in the lowest range a new anode remained bright during electrolysis. In both the highest and lowest ranges a const. overvoltage was rapidly established after changing the c.d., but in the middle range it took some time to establish a const. value. To explain these facts F. and T. make the following hypotheses: Förster's scheme of electrode reactions (*Z. Elektrochem.*, 1907, 13, 414) is accepted as a basis. At high c.d., the anode is covered with NiO_2 and the rate-controlling process is the discharge of OH′ ions, though other reactions, such as the formation of H_2O_2 or O″ ions, may take place. In the middle range of c.d., the rate of reaction is controlled, partly by the discharge of OH′ ions, partly by the formation of NiO_2, and change of c.d. involves change in the compn. of the surface layer on the anode, which is a slow process. At low c.d. the rate-controlling process is the formation of NiO_2, and the compn. of the surface layer is independent of c.d. If, however, the anode had not been previously subjected to a high c.d., the curve agreed with this last hypothesis only at 80° C.; at 25° C. Volchkova and Krasil'shchikov's curve (*Zhur. Fiz. Khim.*, 1949, 23, 441; *M.A.*, 19, 790) was confirmed. —G. B. H.

***Nitrogen Overvoltage [on Platinum] in the Electrolysis of Solutions in Liquid Ammonia.** N. M. Gessler and V. A. Pleskov (*Zhur. Fiz. Khim.*, 1950, 24, (4), 445–453).—[In Russian]. The discrepancy between the high decompn. potentials of soln. in liq. NH_3 and the low free energy of decompn. can be explained by the large overvoltage, especially on the anode. This was proved by direct measurements of the N overvoltage on Pt in soln. of NH_4Cl, NH_4Br, NH_4ClO_4, and NH_4NO_3 in liq. NH_3 at −50° C., which gave values of 1·2–1·4 V., irrespective of the nature of the anion. The overvoltage increased linearly with the log of the c.d., the slope of the line being $(2RT/F)$. G. and P. consider that the primary anodic process in soln. of strong acids is the electrochem. oxidation of NH_3, and that the anion is not discharged, though with NH_4I some I is evolved. The overvoltage arises mainly from the slowness of the electrochem. stage, not from the desorption of N from the electrode. In soln. of KNH_2, the evolution of N proceeds by discharge of NH_2' ions, since some hydrazine is formed.—G. B. H.

***Study of the Sintering Process in Platinized Platinum by Electrochemical Methods. I.—Heat-Treatment in an Atmosphere of Hydrogen.** M. I. Nikolaeva and A. I. Shlygin (*Zhur. Fiz. Khim.*, 1950, 24, (4), 427–431).—[In Russian]. A platinized Pt electrode was prepared by electrolytic deposition, and curves were taken showing its electrode potential in relation to the quantity of electricity which had passed through it, during the electrolysis of $N/10$-H_2SO_4 under an atmosphere of H. Hence the effective surface area of the electrode for adsorption could be deduced. The same electrode was then used to catalyse the decompn. of H_2O_2 and the velocity const. of the reaction was measured. These measurements were repeated after the electrode had been heated in H for 1 hr. at temp. from 40° to 300° C. The adsorption surface fell steadily with rising temp. of heat-treatment, except for a sudden drop at 80° C., reaching 4% of the initial value after treatment at 300° C. The catalytic activity fell to 30% of the initial value after treatment at 60° C., then rose to a max. at 80° C. and fell to 2·3% of the initial value at 300° C. N. and Sh. suggest that the catalytic activity is caused mainly by lattice imperfections, which disappear at temp. <80° C., and that the subsequent fall in catalytic activity and adsorption surface is caused by recrystn. Their results differ from those of previous experiments, in which platinized Pt electrodes were annealed *in vacuo* with no effect on their activity at temp. below 250°–300° C.—G. B. H.

The Zinc/Silver Peroxide/Alkaline Storage Battery. Samuel Eidensohn (*J. Electrochem. Soc.*, 1952, 99, (9), 252C–254C).—The construction and elect. characteristics of this battery are described.—G. T. C.

The Lead Dioxide/Zinc Reserve-Type Cell. W. J. Schlotter (*J. Electrochem. Soc.*, 1952, 99, (8), 205C–206C).—Chem. pure Zn amalgamated with Hg is used as the anode and a Pb-alloy grid for the cathode in this cell.—G. T. C.

The Influence of Electrolytes on Hydrogen Overvoltage and the Theory of Slow Discharge. A. N. Frumkin (*Zhur. Fiz. Khim.*, 1950, 24, (2), 244–253).—[In Russian]. F. shows that, after a secondary correction by de Béthune (*J. Amer. Chem. Soc.*, 1949, 71, 1556; *M.A.*, 17, 369) the overvoltage theory of Eyring and his collaborators (*J. Chem. Physics*, 1939, 7, 1053) gives the same basic kinetic relations as the slow-discharge theory previously developed by F.

and his collaborators (*Z. physikal. Chem.*, 1933, [A], **164**, 121). The only remaining difference between the two theories is that the American workers, having split the jump in potential at the cathode into two parts and calculated the surface concentration of H ions, do not deduce from this any phys. based picture of the structure of the double layer, do not give any experimental data on the magnitude of the surface charge, and do not admit that the charge is const. on the inner part of the ionic sheath in the double layer. F. shows that the equations obtained in this way must be regarded as approximations to the equations which arise from the theory of the double layer, and that the approximation used is valid only in certain conditions. He shows that, in the presence of *n*-valent cations, the potential at a distance of one ionic radius from the electrode surface varies with the cation concentration $[Me^{n+}]$ in soln. according to the equation $\psi_1 \cong (RT/nF)\ln[Me^{n+}] + \text{const.}$, provided that $[Me^{n+}]$ exceeds a certain value, given by $[Me^{n+}]/[H_3O^+]^n \gg 1/y^{n-1}$, where y is the root of the equation $\varepsilon = -2dFy - \sqrt{(n-1)RTD/2\pi n} \,.\, \sqrt{y}$ (ε is the charge density and d the thickness of the Helmholtz double layer, and D is the dielect. const. of the solvent).
—G. B. H.

10 — REFINING

Refined Aluminium: Its Manufacture and Applications. —— (*J. Four Elect.*, 1952, **61**, (1), 17–21).—Refined Al is 99·99–99·999% pure, and is produced electrolytically. This review deals with the principles of the refining process, the Hoopes and Péchiney chloro-fluoride techniques, the refining of Al scrap contg. $\not>$10% metallic impurities, means of reducing the consumption of elect. energy during processing, and the properties and appn. of the refined metal.—H. A. H.

Additional Note on the Subject of Refined Aluminium. —— (*J. Four Elect.*, 1952, **61**, (1), 21–22).—See preceding abstract. Two further procedures for the production of refined Al receive brief mention, and several general patents are listed.—H. A. H.

***The Volatilization [Catalytic Distillation] of Aluminium [from Scrap Aluminium and Its Alloys] as Aluminium Monofluoride.** Armin Schneider and Werner Schmidt (*Z. Metallkunde*, 1951, **42**, (2), 43–54).—A study has been made of the catalytic distillation of Al according to the reaction $2Al + AlF_3 \rightleftharpoons 3AlF$, in regard to its appn. to the sepn. of Al from scrap Al-base alloys contg. Cu, Fe, Mn, and Si. Under a vacuum of 1 mm. Hg the rate of AlF formation, which is slow at 800° C., attains a max. at 900° C., at which temp., also, the sublimate produced contains a max. amount of free Al. Sepn. from Cu and Fe is almost quant. in the range 850°–900° C., but Si and Mn are not so easily removed. Direct heating of the sublimate to melt and thereby remove its content of finely divided Al powder leads to large losses in Al and AlF_3. A flux melting procedure must be adopted, therefore, and binary and ternary fluxing mixtures of alkali and alkaline-earth halides—fluorides and chlorides of Al, Ca, K, Li, and Na—have been investigated. Those most suitable are mixtures near to the eutectic compn., and contg. NaF. A cyclic process could not be developed as, owing to their contamination with the fluxing salts, the AlF_3 residues were found to be unsuitable for use in treating fresh quantities of scrap Al. 19 ref.—E. N.

***†A New Process of Refining Aluminium Scrap.** A. Domony and R. Várhelyi (*Acta Techn. Acad. Sci. Hungar.*, 1951, **2**, (1), 143–158).—[In German]. After a brief account of the relative proportions of impurities in Al scrap or residues—dross, dissolved and occluded gases, metallic and non-metallic impurities—and of the advantages and disadvantages of current refining processes, based principally upon treatment of Al residues for removal of Mg and Zn as chlorides, and of other metallic impurities as sulphides, an account is given of an experimental refining process, worked out by D. and V., utilizing simultaneously all the advantages of treatment by S and Cl. It is based upon treatment of the residues or scrap with liquid S_2Cl_2, which is decomposed by Mg in accordance with the equation $3Mg + S_2Cl_2 = 2MgS + MgCl_2$. The optimum removal (nearly 100%) of Mg was effected by the addn. of ~3 g. S_2Cl_2/1 g. Mg to an alloy contg. 4–6% Mg. The temp. of the molten bath was 730°–740° C. and in some cases 800° C. The process appears to offer advantages over the customary process. The thermochemistry of the process is briefly discussed. Treatment by S_2Cl_2 for removal of Ti and V was completely unsuccessful. S_2Cl_2, for the almost complete removal of dross and gases from Al scrap contg. Mg, must be used only in the dry state, as it is decomposed by moisture, and is, accordingly, obtainable in commerce, only in the dry state.—J. S. G. T.

***Decarburization of High-Carbon Cobalt Metal.** John H. Dismant, J. Hugh Hamilton, W. Martin Fassell, and John R. Lewis (*J. Metals*, 1952, **4**, (8), 884).—Precipitated Co oxides were reduced with petroleum coke in a small 3-phase arc furnace, giving a metal contg. 2·5–3·5% C, a value much too high for the metal to be used in the prodn. of permanent magnets. Two methods, based on the fact that molten Co is decarburized by CO_2, were used to reduce the C content, viz., bubbling of CO_2 through, and maintenance of a CO_2 atmosphere, for ~30 min., above the molten metal contained in a H.F. induction furnace. The resulting product contained 0·003–0·03% C. It was found, also, that decarburization could be carried out by heating the metal with an acetylene blowpipe adjusted to give an oxidizing flame. The suggestion is made that the same basic principle could be used for the decarburization of Ni and, perhaps, Mn. 3 ref.—E. N.

***Refining Secondary Copper Alloys.** Marvin Glasenberg, L. F. Mondolfo, and A. H. Hesse (*Trans. Amer. Found. Soc.*, 1951, **59**, 465–471; discussion, 471).—Cu-base scrap contg. Zn, Sn, Pb, and Ce >1%, and Ni, S, Sb, Al, Si, Mn, and P <1%, was blown by using either air or commercially pure O. The analyses taken at different intervals show that the speed of removal of Al, Mn, Si, P, Fe, and Zn decreases in that order, other elements not being affected. The use of O did not lead to an increased rate of oxidation of these elements out of Cu.—V. K.

***The Hot-Wire Process for Zirconium.** Robert B. Holden and Bernard Kopelman (*U.S. Atomic Energy Commission Publ.*, 1951, (**SEP-58(x)(rev.)**), 28 pp.).—To provide data for the prodn. of Zr by the van Arkel method, a study was made of the thermal decompn. of a ZrI_4 mol. isolated from possible reaction with other tetra-iodide mol. and with the reaction products. It was found that the probability of a given ZrI_4 mol. decomposing on a single contact with a hot surface *in vacuo* is very high and approaches unity in the range 1300°–1500° C. Pressure measurements were also made in a vessel in which Zr was being prepared by iodide decompn., and it was concluded that the rate of prodn. of Zr was determined by gaseous transport of the reactants and that the initial synthesis of ZrI_4 was probably catalysed photochem. by the light radiated from the W filament. The total amount of free I in the cracking chamber was very small, and it is suggested that the decompn. process could probably be accelerated by a forced circulation of the vapours. 5 ref.
—B. W. M.

†Vacuum Metallurgy: Its Characteristics and Its Scope [in Refining]. W. J. Kroll (*Vacuum*, 1951, **1**, (3), 163–183).—Actual and potential app. of vacuum technology in metallurgy are discussed. The performance and limitations of existing vacuum-furnace designs are surveyed, and met. procedures which could be improved by the appn. of vacuum technique are reviewed. Among matters discussed are: phenomena observed in vacuum fusion, condensation of vapour, distillation of metals in a vertical retort, plant materials, heating systems and furnaces, the melting and sepn. of metals (the distillation of Mn *in vacuo* yields very pure material suitable for use in the manufacture of ductile Mn alloys, but the method cannot compete with the electrolytic process). The appn. of vacuum may be a costly matter, but it is a first-rate means of ensuring precision and refinement in processing. It is certain that, in time, vacuum technique will enter branches of metallurgy not even thought of at present. 88 ref.—J. S. G. T.

---

11 — ANALYSIS

***Spectrographic Estimation of Traces of Aluminium in Tin Bronzes.** E. V. Rouir and A. M. Vanbokestal (*Rep. 14me Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, **1951**, 15–20, 21, 23; discussion 20).—A multisource unit and soln. of samples were used for the detn. of Al in the range 0·001–0·1% in Sn bronzes. The special feature of the method was the use of a rotating disc as the lower electrode; this dips in the soln., and its upper edge is sparked. The metal is dissolved in HNO_3, and the soln. treated with NH_4Cl and NH_4NO_3 as well as $NaNO_2$ so that Cu, Pb, and Sn each form various complex compounds in which they are not readily ionized in the spark; this increases the sensitivity of the method for Al at the expense of other elements. Synthetic standard soln. are used for calibration.—E. van S.

***The Determination of Hydrogen in Aluminium.** E. Papp, L. Zombory, and I. Magyarossy (*Acta Techn. Acad. Sci. Hungar.*, 1952, **3**, (3/4), 335–354).—[In German]. After a review of the literature on the subject, a method of determining H present, with other gases, viz. O, N, CO, and CO_2, in Al is described. Essentially the method consists of melting the metal *in vacuo*, drawing off the gases and leading them in a stream of N into an analysing apparatus resembling a rather simplified form of the Shakespear katharometer, in which a wire, heated elect., is surrounded by the gas, and in which the % H is determined in terms of the temp. attained by the wire, heated by a const. current and cooled by the surrounding gas, the cooling of the wire being greater the greater the % H present, as the thermal conductivity of H is much greater than that of any other gas. The indicator is graduated in terms of the H % present. 22 ref.—J. S. G. T.

***Studies in the Determination of Arsenic. VII.—Determination of Arsenic in Lead, Copper, Nickel, and Cobalt.** Yachiyo Kakita (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 698–706).—[In English]. Cf. *ibid.*, (5), 622; *M.A.*, **20**, 113. Rapid detn. of As in Pb, Cu, Ni, and Co, both iodometrically and colorimetrically after sepn. in the metallic state, e.g. by means of $SnCl_2$ in the case of As in Pb, are briefly described. As in Pb is rapidly determined colorimetrically after sepn. of Pb as $PbSO_4$. As in Cu is rapidly determined colorimetrically after sepn. of Cu as $Cu(OH)_2$, while rapid detn. of As in Ni and in Co is effected photometrically or colorimetrically without preliminary sepn. Methods for the detn. of As in Ag and in Cd, Zn, and Mg, colorimetrically and otherwise are very briefly referred to.—J. S. G. T.

***Determination of Metallic Beryllium and Beryllium Carbide in Beryllium Metal.** Walter A. Bergholz (*J. Research Nat. Bur. Stand.*, 1952, **48**, (3), 201–205).—The sample is dissolved at a controlled rate in KOH soln. The liberated H and CH_4 are burned in a N atmosphere to form H_2O and CO_2, and these are absorbed in Mg perchlorate and soda–asbestos, resp. Metallic Be is determined from the H_2O and Be_2C from the CO_2 formed, appropriate corrections being made for impurities. Suitable apparatus is described in detail. Soln. in HCl simplifies the technique and gives satisfactory results for Be but low values for Be_2C. 31 ref.
—J. W. C.

***Quantitative Chromatographic Separation and Estimation of Cadmium in Copper Alloys, Using Paper Strips.** R. L. Davies (*Nature*, 1951, **168**, (4280), 834).—A method of estimating small quantities of Cd in Cu–Cd alloys is described, which is similar to that for estimating Zn in soldering alloys, described by Bishop and Liebmann (*ibid.*, 1951, **167**, 524; *M.A.*, **19**, 405). The sepn. is achieved by paper chromatography, this being followed by polarographic estn. For alloys contg. 0·2–1·3% Cd, the coeff. of variation has been found to be ~1%, which compares very favourably with the spectrographic soln.-spark method (~3%).—R. S. B.

***Spectrographic Analysis of Metallic Surfaces, in Particular Those Produced by Chromizing.** (Mlle) F. Malamand (*Rep. 14me Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, **1951**, 85–113).—A controlled spark source, with a Ag counter-electrode traversing a sample at 0·5 mm./sec. has given quant. results for shallow layers of Cr–Fe alloys formed during vapour-phase chromizing. Using either Fe or Ag lines as internal standards, good agreement with chem. analysis is obtained in the range 10–50% Cr. The variations of Cr with depth are traced by electrolytic polishing, so as to remove ~10 μ/min. from the surface to be examined. The results of a series of measurements are found to agree with those predicted from the theory of intermetallic diffusion. 19 ref.—E. van S.

***Quantitative Inorganic Chromatography. II.—Determination and Detection of Copper in Zinc. III.—Separation and Determination of Zinc in Cadmium.** Yoshinaga Oka and Akira Murata (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 707–710; 711–715).—[In English]. Part I is of no met. interest. [II.—] The chromatographic detection and detn. of Cu in Zn were studied with filter-paper impregnated with $Al(OH)_3$. Cu can be separated from Zn by development with water. 1·9γ Cu was determined in a dilution of 1/50 in Zn, and 0·06γ Cu can be detected in a dilution of 1/100,000 in Zn. The appropriate techniques are briefly described. [III.—] Zn and Cd cannot be separated chromatographically by development with water; when KI is added, however, Cd, forming a complex, moves to the front of the solvent zone and is completely separated from Zn, which remains at its original position. 3γ Zn in a dilution of 1/1000 in Cd, and 0·06γ Zn can be detected chromatographically in a dilution of 1/10,000 in Cd. The necessary techniques are briefly described.—J. S. G. T.

***Catalysis Analysis. XIX.—Rapid Microdetermination of Copper in Various Metals.** Hidehiro Gotô and Shin Suzuki (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 35–40).—[In English]. Cu has a catalytic action upon the reduction of the ferric ion in acid soln. contg. $Na_2S_2O_3$. Effects of Cu, Pb, Sn, Ni, Bi, Mg, Zn, Mn, Co, Cr, Al, Hg, and Cd on this reaction were studied. By measuring, with a Pulfrich or other photometer, the time required to decolourize the dark-violet colour of the ferric thiosulphate complex in the presence of K thiocyanate, under const. conditions, as little as 0·001–~0·1% Cu in the various metals can be determined. Suitable acid concentrations and effects of the various metals are discussed.—J. S. G. T.

***Quantitative Spectrochemical Analysis of Ancient Bronzes.** M. van Doorselaer (*Verhandel. Koninkl. Vlaam. Acad. Wetenschap., Klasse Wetenschap.*, 1950, **12**, (35), 3–27; *C. Abs.*, 1951, **45**, 2367).—Dissolve 4 mg. of sample in $HNO_3 + HCl$; allow the soln. to be absorbed in a pure graphite electrode at 80° C. and evaporate to dryness. To determine Cu, Sn, and Pb, obtain the condensed-spark spectrum of a soln. contg. 1% bronze and 2% added Cu and compare the intensities of the line-pairs Sn 2429·5/Cu 2441·6 and Pb 2833·07/Cu 2882·9. From the results, by means of standard curves, the content of all 3 metals can be determined with a precision of approx. 3%. For the minor and trace elements, obtain the Pfeilsticker interrupted-arc spectrum of a 3% soln. of the bronze and compare the intensities of the lines Ni 3050·8/Cu 3073·8, Fe 2488·15/Cu 2400·1, Zn 3345·6/Cu 3349·3, As 2349·8/Cu 2400·1, Sb 2877·9/Cu 2882·9, Co 2407·26/Cu 2400·1, Au 2428·0/Cu 2400·1, Bi 3067·73/Cu 3073·80, and Ag 3382·9/Cu 3349·26. Other line-pairs are recommended for special cases. For these elements the precision is approx. 10%. Analyses of various bronze artifacts are included. The methods are also applicable to a variety of commercial Cu alloys.

***Estimation of Lead: Gravimetric Analysis with Phenylarsonic Acid.** Anil Kumar Majumdar and Ramendra Nath Sen Sarma (*J. Indian Chem. Soc.*, 1951, **28**, (11), 654–656).—Phenylarsonic acid can be used to estimate Pb in the presence of alkali and alkaline-earth metals over a *p*H range of 6·8–7·4. Alkali metals and Ca co-precipitate when present over a certain amount, but co-pptn. of Ca can be avoided by a double pptn.—W. A. M. P.

***The Applicability of Lead Amalgam Electrodes in the Estimation of $Pb^{\cdot\cdot}$, SO_4'', and $Ba^{\cdot\cdot}$, by Potentiometric Titration.** Sunil Kumar Dutta and B. N. Ghosh (*J. Indian Chem. Soc.*, 1951, **28**, (7), 383–389).—A two-phase Pb amalgam electrode has been used in the estn. of $Pb^{\cdot\cdot}$, $Ba^{\cdot\cdot}$, and SO_4'' from aq. soln.—W. A. M. P.

***Quantitative Spectrographic Analysis of Nickel Cathodes for Valves.** B. Vallantin and R. Hervé (*Rep. 14^me^ Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, 1951, 183–185, 187–193; discussion, 185).—An A.C. arc with H.F. ignition and soln. of Ni evaporated on C electrodes have been used in the detn. of Mn, Mg, and Ca in cathode Ni. Soln. are necessary because of the lack of standard samples and the variability of sample form. The microphotometer used is fitted with a const.-intensity lamp monitored by a second photo-cell.—E. VAN S.

***New Colorimetric Method for Determining Palladium in Refined Silver and Other Silver Products.** G. V. Knyazheva (*Izvest. Sekt. Platiny*, 1948, (22), 129–135; *C. Abs.*, 1950, **44**, 10584).—[In Russian]. The method described is a modification of the method wherein H_2PtCl_4 is acted upon by $SnCl_2$. The reaction product is extracted with ether and has the characteristic yellow colour. Frequently this colour was marred by an unstable pink colour, which was traced to the presence of Pd. The pink colour could be stabilized by Cu, in which case it remained stable for several days. Thus, to determine Pd, take a HCl soln. of the substance or a neutral nitrate soln. and add to it conc. HCl 10, H_2O 10, $CuCl_2$ 1 (contg. 0·2 mg. Cu), 10% SnCl 20–25, and ether 10–15 ml. If Ag is present, AgCl precipitates but does not interfere. If Pt is present, it collects in the ether. Add a little NaH_2PO_2 to destroy the Pd colour and determine Pt colorimetrically in the usual way. Prepare a standard soln. to which are added all the above reagents and in addn. a quantity of Pt found in the analysed sample. Add slowly a Pd standard soln. with shaking until the colour of the analysed sample and that of the standard match. The results of Pd detn. made by this method agree well with grav. results, but whereas the grav. method requires a 100-g. sample, the colorimetric requires only 1-g. For Pd up to 0·0006 g. the amount of Cu required is 0·02 mg. or more. Colour comparison can be made in transmitted or reflected daylight, but artificial light causes difficulties. Cu^{++}, Fe^{++}, Fe^{+++}, Zn, Al, Pb^{++}, Cd, and Ag do not interfere. Pt up to a Pd : Pt ratio of 2 : 1 does not interfere; above that Pd should be determined as outlined. Se and Te should be separated.

***Determination of Platinum and Rhodium in Their Alloys.** D. I. Ryabchikov (*Izvest. Sekt. Platiny*, 1948, (22), 28–34; *C. Abs.*, 1950, **44**, 10586).—[In Russian]. The detn. of Rh is based on the reduction of Pt^{4+} to the Pt^{2+} state and on destroying the yellow colour of $PtCl_6^{2-}$ with HSO_3^-. The characteristic rose colour of $RhCl_3^{3-}$ remains unaffected, and is used for the colorimetric detn. of Rh. The reaction proceeds in 2 stages: $M_2[PtCl_6] + KHSO_3 + H_2O = M_2[PtCl_4] + KHSO_4 + 2HCl$ and $M_2[PtCl_4] + 4KHSO_3 = M_2[Pt(SO_3)_4] + 4HCl$. In the presence of HCl the colour of Rh is more stable. To determine Rh dissolve 1 g. of alloy in *aqua regia* and heat to remove N oxides. Evaporate almost to dryness, transfer to a 100-ml. volumetric flask, and acidify with HCl. Prepare a standard by dissolving 0·2513 g. of very pure $(NH_4)_3[RhCl_6]$ in a 100-ml. volumetric flask and acidify with HCl. Prepare a saturated $KHSO_3$ soln. The standard and $KHSO_3$ soln. should be prepared daily. Into the left cylinder of a Duboscq colorimeter put 10 ml. of standard, 3 ml. of saturated $KHSO_3$, 2 ml. of conc. HCl, and mix thoroughly. Into the right cylinder put 10 ml. of soln. to be analysed, 3 ml. of $KHSO_3$ soln., 2 ml. of HCl, mix, and read the colour after 2–3 min. The order in which the reactants are added must be as given. Take readings on several aliquots and determine the average value. Pt is determined by either reduction or oxidation potentiometric titration. The reduction titration is: $PtCl_6^{--} + 2CuCl = PtCl_4^{--} + 2CuCl_2$. This reaction must be carried out in an atmosphere of CO_2. To make sure that all the Pt is quadrivalent, add a few drops of freshly prepared Cl water. The titration is carried out to the 2nd break. The 1st break is due to the reduction of excess Cl_2, but the 2nd is caused by the reduction of all of the Pt. In the oxidizing titration the Pt in soln. is reduced with Cu_2Cl_2 and then titrated with $KMnO_4$. In this titration also there are 2 breaks, the 1st being due to oxidation of excess Cu_2Cl_2. Throughout this procedure of $KMnO_4$ titration it is advisable to use as little HCl as possible and to use the Reinhardt reagent.

***Solution of Metallic Rhodium.** N. K. Pshenitsyn and V. A. Golovnaya (*Izvest. Sekt. Platiny*, 1948, (22), 111–113; *C. Abs.*, 1950, **44**, 10587).—[In Russian]. The method described is a modification of the Hefts method (cf. *Ann. Chem. u. Pharm.*, 1868, **146**, 271), and consists of fusing Rh with Zn. Place a sample of Rh in a porcelain crucible, add 20 times its wt. of granulated Zn, and cover with a well-tamped layer of flux consisting of intimately mixed NaCl 50 and KCl 50%. The salts used for the flux should be thoroughly dry, and their combined wt. should be twice the wt. of Zn. Cover with a porcelain lid, place the crucible in a cold or slightly heated crucible furnace, raise the temp. to 700° C., and maintain at this temp. for 8 hr. Remove from the furnace, cool, place in a large beaker, add 5% HCl soln., cover with a watch glass, and allow to stand until all the uncombined Zn dissolves. Wash the watch glass and beaker walls with hot H_2O and filter. The undissolved residue is Rh–Zn alloy. Wash the residue with hot H_2O, place the filter in an Erlenmeyer flask, add *aqua regia* (HNO_3 1 and HCl 7 parts), cover, and boil for 1 hr. Transfer to a porcelain dish, evaporate to dryness on a water bath, add a few drops of conc. HCl, and again evaporate. Dissolve in hot H_2O and wash the filter with hot H_2O. The residue consists of silicates and Ir if present. Reduce the Rh in the filtrate to metal by any of the known methods, filter, wash, dry, ignite, reduce in H, cool, and weigh. The steps in this procedure must be followed exactly. By this method 99·7–100% of Rh was dissolved in one fusion.

***Studies in the Determination of Tellurium and Selenium. —I.–II.** (I) Hidehiro Gotô and Yachiyo Kakita; (II) Hidehiro Gotô and Tadahiko Ogawa (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1952, [A], **4**, (1), 28–34; (2), 121–126).—[In English]. [I.—] Detn. of Te and Se in HCl soln. contg. $HClO_4$, by reduction with SO_2, were studied in detail. Te was completely precipitated in HCl concentrations of 2–4N and Se at a concentration of the acid $>6N$ even in the presence of 5 ml. $HClO_4$ in a total vol. of 100 ml. The difference of acid concentration range required for pptn. of Se and Te, permits quant. sepn. of the two elements. Se is first precipitated in HCl concentration of $8{\cdot}5 \sim 9{\cdot}5N$, and Te then precipitated from the dil. filtrate at an HCl concentration of $2 \sim 4N$. [II.—] Te was precipitated quant. by SO_2 from the HCl soln. of concentration $2N$, even in the presence of 5 ml. H_2SO_4 in 100 ml. soln., but not in a concentration of HCl $>9N$. Se was precipitated quant. under the same conditions as Te when the HCl concentration was $>4N$. Hence sepn. of Se and Te, in the presence of H_2SO_4 is possible if Se is first precipitated at a HCl concentration $>9N$ and then Te precipitated from the filtrate after dilution to $2N$-HCl.
—J. S. G. T.

Vanillic Acid: A New Reagent for Thorium. K. V. S. Krishnamurty and A. Purushottam (*Rec. Trav. Chim.*, 1952, **71**, (6), 671–675).—Th may be precipitated from soln.

of *p*H ~3·6 by vanillic acid. Other elements, e.g. Ca, Sr, Ba, Mg, Zn, Cd, Ni, Co, Be, and Al, do not interfere. The precipitate is ignited and weighed as oxide. Experimental error is shown to be small.—J. W. C.

***Analytical Chemistry of Thorium. VIII.—Aliphatic Dicarboxylic Acids.** T. V. S. Suryanarayana and Bh. S. V. Raghava Rao (*J. Indian Chem. Soc.*, 1951, **28**, (9), 511–514).—Cf. *ibid.*, (5), 261; *M.A.*, **19**, 857. Experiments with succinic and adipic acids are described. Succinic acid is found to form a complex with Th, and at the optimum *p*H for the pptn. of Th, the cerite earths are also precipitated. Adipic acid is shown to precipitate Th quant. from cerite earths after a double pptn. in the *p*H range 4·2–4·4. Th in monazite has been successfully estimated in this way. —W. A. M. P.

***Analytical Chemistry of Thorium. IX.—Delayed Precipitation or Precipitation from Homogeneous Solution.** C. Lakshman Rao, M. Venkataramaniah, and Bh. S. V. Raghava Rao (*J. Indian Chem. Soc.*, 1951, **28**, (9), 515–518).—Cf. preceding abstract. The use of two new reagents, NH_4 picrate and the NH_4 salt of 2 : 4-dinitrophenol, for the estn. of Th and its sepn. from cerite earths is described. Both reagents function satisfactorily in the *p*H range 4·8–5·2, and Th in monazite has been successfully estimated with both reagents.—W. A. M. P.

***Analytical Chemistry of Thorium. X.—Ammonium Furoate (Pyromuciate) and Sodium Sulphanilate.** D. Lakshminarayana and Bh. S. V. Raghava Rao (*J. Indian Chem. Soc.*, 1951, **28**, (10), 551–554).—Cf. preceding abstract. The use of two new reagents, NH_4 furoate and Na sulphanilate, for the estn. and sepn. of Th from the rare earths is described. Both reagents have been used to estimate Th in monazite. —W. A. M. P.

***The Identification of Zinc in the Presence of Nickel and Cobalt.** E. M. Levine and Gerald Krakower (*J. Chem. Education*, 1952, **29**, (9), 437).—Centrifugal sepn. of Ni and Co hydroxides from $Zn(OH)_4^{--}$, in the method in which the latter is treated with H_2S to give a ZnS precipitate, is not always complete, this being shown by a blackening of the precipitate. Complete sepn. can, however, be obtained by filtration of the $Zn(OH)_4$. It is shown that the presence of Ni and Co in large amounts makes semi-quant. detn. of Zn unreliable.—J. R.

***Analytical Chemistry of Zirconium. III.—Salicylic and Phenoxyacetic Acids.** T. Viswanadha Sastry and Bh. S. V. Raghava Rao (*J. Indian Chem. Soc.*, 1951, **28**, (9), 530–532).—Cf. *ibid.*, (5), 257; *M.A.*, **19**, 857. Some experimental work on the use of salicylic acid and phenoxyacetic acid for the estn. of Zr is described. Ti and Ce interfere, but Th is removed in a double pptn. Fe may also be removed, and other divalent and trivalent elements cause no difficulty. —W. A. M. P.

***An Analysis of the Factors in Spectrochemical Microanalysis.** J. K. Hurwitz (*J. Opt. Soc. Amer.*, 1952, **42**, (7), 484–489).—As an accessory to the met. use of the spectrograph, a sample-traversing and plate-moving attachment is described which can be applied to local analysis; the two moving parts are driven by the same motor, and the ratio of the movements can be varied. The reasonable limits to this variation are discussed. The width of the track scanned (on the sample) was ~0·6 mm. using a spark source and pointed Ag counter-electrode; a discrimination factor for such scanning methods is defined, and a typical steel segregation analysis is illustrated.—E. van S.

X-Ray Fluorescence Analysis: Non-Destructive Testing at Shop Level. Fred Behr (*Steel*, 1952, **130**, (12), 70–71).—Using high-intensity X-ray equipment and Geiger counters, a method has been developed for the analysis of all elements from Ti to U. With modifications it is also possible to extend the range several elements lower than Ti in atomic number. Results are obtained in 3–5 min. after specimens are prepared, and an optimum accuracy of the order of $\frac{1}{2}$–1% of the element present is claimed under favourable circumstances. —E. J.

12—LABORATORY APPARATUS, INSTRUMENTS, &c.

(See also "Testing" and "Temperature Measurement and Control".)

***Dilatometric Determination of the Solidus Temperature [of Aluminium Alloys].** J. A. Verö (*Acta Techn. Acad. Sci. Hungar.*, 1951, **2**, (1), 97–113).—[In English]. The fully automatic detn. of the solidus temp. of alloys of Al (purity 99·995%) with electrolytic Cu, Zn, Cr, and Sn (purity 99·85%) by a photographic–dilatometric method is discussed and illustrated. Melting of a small quantity of eutectic in cast alloys gives a distinct break in the dilatometer curve. Peritectic alloys, in which the main part of the alloy becomes liquefied at the solidus temp., collapse, so that the dilatometer curve ends at this temp. Wrought alloys must be heated very slowly or re-cast. Alloys consisting of a homogeneous solid soln. give a dilatometer curve which becomes steadily steeper when liquefaction begins, and the solidus temp. cannot, in such a case, be accurately determined. Correct values can, however, be obtained if the heating of the furnace is controlled by a regulator, as then loops are formed at the solidus temp. The correctness of the straight solidus line of high-purity Al–Cu alloys, given by Dix and Richardson (*Trans. Amer. Inst. Min. Met. Eng.*, 1926, **73**, 560; see *J. Inst. Metals* (Abstracts), 1926, **35**, 535) for the range of the Al-rich solid soln. was confirmed. With Al–Zn alloys contg. 60–80% Zn, a lower solidus temp. was found than in previous detn. A straight line is now drawn through the newly determined points in this range in place of the generally accepted upwardly curved graph. 9 ref. —J. S. G. T.

Note on the Preparation of Spherical Single Crystals of Low-Melting [-Point] Metals [Bismuth, Tin, and Zinc]. Mikio Yamamoto and Jirô Watanabé (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1951, [A], **3**, (6), 655–656).—[In English]. Spherical single crystals of Bi are easily prepared by pouring molten Bi into a soft-glass test-tube, swollen into a spherical bulb at its lower end and preheated at 300° C. in an elect. furnace, and then cooled at 5·6° C./min. This is essentially the Tammann method. Spherical single crystals of Sn can be prepared by sucking up molten Sn through a capillary 3–4 cm. long into a bulb near the lower end of a tube by means of a water-jet pump, sealing the capillary at its lower end, and lowering the metal through a furnace, maintained at 325° C., at the rate of 0·6 or 1·6 cm./hr. Spherical single crystals of Zn can be prepared in a similar manner to that used for Sn, except that hard-glass bulbs are used, and the metal sucked into the bulb near the subsequently sealed capillary end of the tube is lowered at the rate of 1·9 cm./hr. through a furnace maintained at 470° or 500° C., preferably the former. The crystals showed preferred orientations almost ⊥ the principal crystallographic axis.—J. S. G. T.

Moving Pictures of Metal Structures Develop New Technique for Metal-Structure Study. R. B. Pond and N. K. Chen (*Iron Age*, 1952, **170**, (6), 122–126).—A combination of a cine-camera attached to a microscope and a special micro-tensile testing machine, equipped with strain gauge and a Dynalog recorder, was used to study plastic deformation of high-purity Al crystals produced by gradual solidification from liq. metal.—J. H. W.

***The Isolation and Examination of Films from Metal Surfaces: An Improved Technique.** T. J. Nurse and F. Wormwell (*J. Appl. Chem.*, 1952, **2**, (9), 550–554).—The technique previously described by Vernon, W., and N. (*J. Chem. Soc.*, 1939, 621; *J. Iron Steel Inst.*, 1944, **150**, 81P) for the isolation of films from metal surfaces has been improved by preliminary reinforcement of the film with Formvar resin. In tests of the improved technique with some non-ferrous metals, oxide films could not be removed in an uncontaminated condition from Cu, brass, or Al–Cu alloy, owing to the formation of insoluble CuI (an I soln. is used in the isolation), and film removal from heated Zn or Ti was fragmentary. However, quant. removal of the films from anodized Al and from Al-alloy surfaces uncontaminated with Cu or Si was obtained, but the process is relatively long compared with that for mild steel, as the whole of the anodized metal has to be completely dissolved; on the other hand, the elaborate precautions required with mild steel are unnecessary with anodized Al, and the sepn. time can be greatly reduced by the use of Br instead of I in the soln. Oxide surface films were more easily removed from commercial Al than from Al of super-purity.—J. R.

Uniform Metallization of Surfaces by a Process of Thermal Evaporation, Initiated at One or More Emissive Sources. P. Prugne and P. Leger (*J. Phys. Radium*, 1952, **13**, (Supplement 7/9), 129A–130A).—" Riders " of the metal to be evaporated are arranged, *in vacuo*, at one or more points on the circumference of a W filament, bent into the form of a horizontal loop, and arranged at a definite distance from the plane substrate upon which metal is to be deposited, which rotates about a vertical axis. Formulæ for calculating the thickness of deposited metal and the total mass of the metal " riders " are given.—J. S. G. T.

***A Bent-Crystal X-Ray Monochromator.** Yu. A. Bagaryatsky and E. V. Kolontsova (*Zavod. Lab.*, 1950, **16**, (8), 955–961).—[In Russian]. The monochromator is intended principally for single-crystal work, for which flat monochromators (which give a || beam) have too little intensity. A small aperture (1½–2°) is used, and the geometrical arrangement allows the whole focal spot to be " viewed ". Focusing is necessarily imperfect, but this is immaterial for many appn. where the main requirement is to remove background scattering. The image of the direct beam is either square or in the form of a narrow line. Quartz or topaz was used; the α doublet was not resolved. Formulæ and a graph for calculating geometrical quantities involved are given. The adaptation of ordinary single-crystal cameras for use with this monochromator is described. The main innovation is the use of a wire cross as a " sight-line " (as on a gun); together with the single collimator pinhole, this defines the mean direction of the beam. (Abstracted from Fulmer Research Institute Translation No. 33.)
—R. W. C.

New Developments in the Technique of Focusing X-Rays. Yu A. Bagaryatsky (*Uspekhi Fiz. Nauk*, 1951, **44**, (2), 292–295).—[In Russian]. A review of some new methods in the monochromatizing of X-rays. (The data on Pines' method are incorporated in the following abstract.) Attention is called to the advantage of using a (singly bent) monochromator with the reflecting planes at an angle to the surface. This confers the following advantages: (i) crystal–film distance is large, allowing a large specimen–film distance (transmission method) and therefore high resoln.; (ii) films can be placed ⊥ the beam, again assisting resoln.; (iii) source–crystal distance small, giving good intensity. This method is specially recommended for studies of low-angle scattering. (Abstracted from Fulmer Research Institute Translation No. 34.)
—R. W. C.

B. Ya. Pines' High-Illuminating-Power X-Ray Monochromators. D. B. Gogoberidze (*Uspekhi Fiz. Nauk*, 1951, **44**, (2), 291–292).—[In Russian]. A note. The difficulties encountered when one uses flat or singly bent crystals, or a pair of singly bent crystals, for producing monochromatized X-ray beams, are discussed. The main difficulty, insufficient intensity, has been overcome by B. Ya. Pines (*Symposium in Honour of A. F. Joffe* (*Acad. Sci. U.S.S.R.*), **1950**, p. 448) by using crystals bent into a toroidal shape (double curvature). A point-focus X-ray tube is used, and the monochromatic X-rays are focused to a point. Single-crystal Zn plates 1–1·5 mm. thick are bent to tore radii of 14 and 60 mm., giving a short path length; this is advantageous for soft radiations (e.g. Cr). Rock-salt crystals bent at 680° C. have also been used. A toroidal stamp and die are used. No data on resolution are given. (Part of this information is taken from the paper by Bagaryatsky, see preceding abstract.) (Abstracted from Fulmer Research Institute Translation No. 32.)—R. W. C.

Pendulum Hysteresis Meter: A Torsion Pendulum with Double Elastic Suspension for Measuring the Internal Friction and Elastic Modulus of Metals from the Lowest Temperatures up to 1300° C. Christian Boulanger (*Compt. rend.*, 1951, **233**, (14), 732–733).—An apparatus is described in which a disc is suspended by a wire of a metal with low internal friction ($\sim 10^{-4}$), maintained at a const. temp., which forms a reservoir of potential energy and controls the period of the pendulum (0·5–5 sec.). To the under side of the disc is attached an axial rod of an alloy which is rigid at high. temp. and has a low thermal conductivity. This rod passes into an elect. furnace or a refrigerator. A wire of the metal to be studied (250 mm. long × 0·35 mm. dia.), wound into a spiral of low pitch, is placed between this rod and another one, of the same material, coaxial with the first and firmly secured by clamps. Measurements are possible over the range of stress 1–500 g./mm.².—N. B. V.

A New Spectrograph for Qualitative and Quantitative Analysis. Jean Cojan (*Rep. 14^{me} Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, **1951**, 225–234; discussion, 227).—An unusually compact two-prism quartz spectrograph, of dispersion similar to the Hilger Quartz Intermediate and of resolving power sufficient to resolve the Fe 3100 triplet has a vertical slit and by means of a mirror produces spectra dispersed vertically and focused on an inclined plane. The plate size is 6 × 13 cm. and an opt. bench can be fitted for source and accessories.—E. van S.

Photomultiplier Cells and Circuits Used in Direct [Spectrographic] Analysis. — Macq (*Rep. 14^{me} Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, **1951**, 25–61).—A description of the photomultiplier cell, and of the chief circuits used in its appn. to spectrochem. analysis, expounded with notable simplicity and adequate detail.
—E. van S.

The Zirconium Arc Lamp. — Beaudouin (*Rep. 14^{me} Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, **1951**, 235–236, 239–243; discussion, 237–238).—An arc may be struck between a cathode contg. ZrO_2 and a surrounding metal disc, pierced with a hole, all enclosed in a bulb contg. an inert gas, and provided with a plane window. A source of high tension, such as a spark-ignition coil, is used to initiate the arc which is then limited by a ballast resistance to give ~2 amp. from a 30 V. battery; the intrinsic brilliance approaches that of a C arc crater, and the radiation is like that of a black body at 2500° K. The whole forms an approach to the point source of high intensity.
—E. van S.

13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

A Fatigue-Testing Machine at the B.W.R.A. Laboratory. —— (*Engineer*, 1952, **194**, (5032), 15).—A description of the 100–200-ton Losenhausen universal testing machine installed in the new British Welding Research Association Laboratory.
—D. K. W.

Microhardness Testing. D. L. Paterson (*Canad. Metals*, 1952, **15**, (1), 46, 48, 50–51).—The microhardness testing of case-hardened layers, weld boundary zones, and small components is discussed. Testing procedure is briefly described.
—W. A. M. P.

***Hardness Testing at High Temperatures.** I. L. Mirkin and D. E. Livshits (*Zavod. Lab.*, 1949, **15**, 1080–1087 ; *C. Abs.*, 1951, **45**, 514).—[In Russian]. A Brinell machine equipped with a metallo-ceramic indenter was used for testing the hardness of pure metals (W, Mo, Co, Ni, Fe) and heat-resistant alloys up to 800° C. The high-temp. hardness was correlated to long-time tensile tests at elevated temp. By determining the high-temp. hardness for 3 different loading periods 30, 300, and 3000 sec., a preliminary evaluation of the heat-resistant properties of an alloy could be obtained.

***Measurement of Dynamic Strain by Electrical Resistance Wire.** Sukemitsu Ito (*J. Mech. Lab.* (*Tokyo*), 1952, **6**, (4), 148–153).—[In Japanese]. The principal methods of measuring dynamic strain by changes in the elect. resistance of a wire and the methods of recording these changes are considered. It has been found by experiments reported here that an electromagnetic oscillograph recorder measuring strain up to a max. of 10^{-3} and using a power source on an initially unbalanced bridge circuit of 4500 c./s. and 2 V. has a max. sensibility of 3·3 mm./kg./mm.^{2}, an accuracy of 2·5% when the stress is >6 kg./mm.^{2} and 0·084 μ when it is <6 kg./mm.^{2} on a gauge-length of 12 mm.—Author.

Equipment for Non-Destructive Testing. G. Rigby (*Canad. Metals*, 1952, **15**, (6), 22–23).—Brief notes on magnetic-particle inspection, penetrant methods, and ultrasonic inspection.—W. A. M. P.

Advantages of X-Ray Inspection. J. Rigby (*Canad. Metals*, 1952, **15**, (4), 36–37).—The advantages and limitations of X-ray inspection are discussed, and the use of γ radiography is also mentioned.—W. A. M. P.

Using Iridium 192 for Castings Inspection. —— (*Canad. Metals*, 1952, **15**, (3), 41).—The use of the radioactive isotope Ir^{192} to supplement the X-ray examination of foundry work is described. The method is considered to give better negatives. —W. A. M. P.

Natural and Artificial Sources for Gamma Radiography. B. N. Clack (*Engineer*, 1952, **194**, (5037), 195–198).—The main requirements of a source suitable for indust. γ radiography are enumerated, and a review is made of the extent to which these have been satisfied by natural and artificial radioelements and fission products. A typical source container, suitable for transport, and manipulative devices are briefly described. The advantages of Rn over Ra for radiography are discussed and the prepn. of purified Rn from a Ra source, is described in some detail.—D. K. W.

Protective Measures for Radioactive Isotopes. V. E. Pullin (*Engineer*, 1952, **193**, (5025), 660–662).—P. considers the protection of personnel from radioactive subtances employed in indust. radiography, when in storage, and when in use. —D. K. W.

The Industrial Uses of Radioactive Materials. K. F. Fearnside (*Chem. and Ind.*, **1952**, (38), 920–922).—Includes illustrations and a discussion of the continuous measurement of Al, Cu, or other sheet or coating thicknesses by transmission or back-scattering of β rays.—J. R.

14 — TEMPERATURE MEASUREMENT AND CONTROL

The Measurement of Temperature of Copper Alloys in the Foundry. Albert Blondel (*Fonderie*, 1952, (**78**), 3007–3022).—A Pt/Pt–10% Rh thermocouple in a SiO_2 sheath undoubtedly gives the most accurate results, but the thermocouples are expensive and fragile. The Cromel/Alumel couple in a steel sheath is robust and cheaper, but it is less accurate and responds slowly at temp. above 1240° C. If due precautions are taken, the opt. pyrometer can be of great use in the Cu foundry. 18 ref.—J. H. W.

Study of Platinum/Platinum–Rhodium Thermocouples and Their Industrial Applications. Marcel Chaussain (*Fonderie*, 1952, (77), 2955–2971).—C. discusses the standardization of Pt/Pt–10% Rh thermocouples for use in measuring temp. between 1000° and 1700° C. He describes the study and method of testing and inspecting the wires and the indust. appn. of the thermocouples to high-temp. measurement. —J. H. W.

***Behaviour of Platinum/Platinum–Rhodium Thermocouples at High Temperatures.** R. C. Jewell and E. G. Knowles (*J. Sci. Instruments*, 1951, **28**, (11), 353).—In a dismantled Pt–13% Rh-alloy-wound furnace, it was found by analysis that the Rh had volatilized preferentially. This result is not in accordance with the conclusions of McQuillan (*ibid.*, 1949, **26**, 329 ; *M.A.*, **17**, 379) that changes in the e.m.f. of Pt/Pt–Rh thermocouples after exposure to high temp. are due to preferential oxidation and volatilization of the Pt.—P. C. L. P.

A Period Immersion Pyrometer. R. F. Wright and E. Scorah, with an Appendix by T. Land and R. Barber (*J. Iron Steel Inst.*, 1951, **169**, (3), 243–244).—A thermoelect. pyrometer is described which is capable of continuously measuring the temp. of molten steel for periods up to 2 hr. The pyrometer is intended for use in induction furnaces. The thermocouple wires are protected by an alumina tube, coated with $\frac{1}{8}$ in. of alumina cement to reduce the thermal shock. An alternative method of reducing the thermal shock is given in an Appendix ; in this method the alumina sheath is enclosed in a heavy steel case which slowly heats up and melts away.—R. W. R.

Pyrometers for Surface-Temperature Measurement. M. D. Drury, K. P. Perry, and T. Land (*J. Iron Steel Inst.*, 1951, **169**, (3), 245–250).—Three improved pyrometers for the measurement of the temp. of hot surfaces are described: (1) An instrument incorporating a back-silvered glass hemisphere which permits nearly black-body visible radiation to escape through a small hole and fall on a calibrated barrier-layer photo-cell. This instrument is suitable for temp. >900° C. (2) A similar instrument for use in the temp. range 300°–900° C., in which the photo-cell is replaced by a thermopile and the glass hemisphere by a Pt-plated steel one. The effect of surface emissivity is small. (3) An instrument in which a plated-steel spherical segment focuses radiation on to a thermocouple placed on the hot surface, thus obviating the necessity for good contact between the thermocouple and the surface. This instrument is suitable for temp. up to 500° C. The results of tests of each instrument are described, and a number of theoretical considerations are discussed in appendices. 3 ref.—R. W. R.

***The Measurement of Mean Surface Temperature by High-Frequency Induction.** P. Fortescue and D. M. Harris (*Atomic Energy Research Establishment Rep.*, 1951, (**E/R720**), 11 pp.).—A method is described of measuring the surface temp. of thin metal tubes in a coolant stream without disturbing the flow conditions at the surface. The method depends on measuring the change of surface resistivity with temp. A relation $\Delta R_p = 0{\cdot}039 K^2 n^2 (a/b)\sqrt{\rho f}$ is derived, giving the change in the resistive component of impedance (ΔR_p) of a coil of length b contg. n turns when a hollow non-magnetic core of radius a is placed concentrically inside it, where ρ is the resistivity of the core material, f the current frequency, and K a coupling coeff. of the order of (radius of core/radius of coil)^2. The formula applies when $f > 10^5 \rho/t^2$, i.e. for values of f sufficiently large so that the effective depth of penetration of the current in the core is less than half the min. core tube thickness (t). Under these conditions the resistance of the core has no effect on the coil inductance. The resistive component is measured on a Wheatstone bridge network after tuning the reactance out with a shunt capacity. Experiments conducted at 17° C. with cores of varying resistivity showed that the coil inductance was independent of the coil resistivity, that the resistive component $\propto \sqrt{\rho}$, and that the degree of concentricity was not critical.

Changes of core resistivity of ½%, corresponding to changes of 1½ °C. for Al, could be measured. Corrections for changes in coil temp. and for lead resistance are discussed.
—P. E. M.

A New Method of Converting Platinum Resistance Values to Degrees Centigrade. Frank D. Werner and Alson C. Frazer (*Rev. Sci. Instruments*, 1952, **23**, (4), 163–169).—By interpolation from a single "standard" table, applying corrections which have been worked out for each particular thermometer, it is possible to derive the values of temp. from values of resistance of a Pt resistance thermometer.
—E. J.

***The Calibration of Platinum Resistance Thermometers in the Temperature Range 11° to 90° K.** J. M. Los and J. A. Morrison (*Canad. J. Physics*, 1951, **29**, (2), 142–150).
—N. B. V.

The Low-Temperature Characteristics of Carbon-Composition Thermometers. J. R. Clement and E. H. Quinnell (*Rev. Sci. Instruments*, 1952, **23**, (5), 213–216).—The low-temp. characteristics of some commercially manufactured C-compn. thermometers are discussed. Thermometric sensitivity is such that by measuring resistance to one part in 5000 it is possible to detect temp. changes <0·0001° at 2° K.
—E. J.

15 — FOUNDRY PRACTICE AND APPLIANCES

Melting Aluminium-Base Alloys. L. W. Eastwood (*Canad. Metals*, 1952, **15**, (6), 40, 42–43).—E. briefly discusses charging, melting, and the effect of temp. and holding time on Al alloys.—W. A. M. P.

Large Aluminium Alloy Castings. J. P. Meister (*Aluminium Suisse*, 1952, **2**, (2), 67).—[In French and German]. A ship's ventilator casting, approx. 11 ft. × 9 ft. 5 in. × 5 ft. 9 in. weighing ~1·4 metric-tons and a ventilator fan casting 5 ft. in dia. weighing ~3 cwt., are shown and briefly described.
—A. W. B.

[Casting] Technique for Palladium. —— (*Canad. Metals*, 1952, **15**, (7), 56–57).—A brief note on the casting of Pd for jewellery purposes by the lost-wax method.—W. A. M. P.

***Study on the Zinc Casting Alloys. IV.—On the Fluidity of Zamak Alloys.** Kazuo Katori, Mikio Suzuki, and Kingo Naoki (*J. Mech. Lab.* (*Tokyo*), 1950, **4**, (7), 268–273).—[In Japanese]. The fluidity of Zamak alloys has been determined by the length cast in a spiral die. The effect of Mg addn. and the differences between Zamak 2 and Zamak 3 are discussed. Results show that Mg tends to reduce fluidity of both alloys and that Zamak 3 is superior when Pb is present.—AUTHORS.

Chemical Treatment of Molten Non-Ferrous Metals.—I, —II. D. A. Dodson (*Canad. Metals*, 1952, **15**, (6), 32, 34–35, 64).—[I.—] D. discusses the principles involved in the chem. treatment of molten Cu, brasses, bronzes, and Ni-Cu alloys. [II.—] The chem. treatment of molten Al and Mg alloys is discussed.—W. A. M. P.

Casting Developments. A. Dunlop (*Metal Ind.*, 1952, **81**, (10), 183–185; (11), 206–207).—D. describes the principles, technique, defects, and limitations of methods of moulding for casting by surrounding the pattern with expendable moulding material in what is, in fact, an investment process. In particular, he describes the Mercast, the Shaw, and the "C" processes.—J. H. W.

Sand Casting with Croning Process Shell Moulds. Richard Herold (*Canad. Metals*, 1952, **15**, (5), 30–32, 34–35).—Cf. *M.A.*, **20**, 205. Data are given on casting size, prodn. speed, support material and methods, heat sources, coring, mould assembly, pattern metals, gating and risering, pouring temp., &c. Shell moulding is considered to be particularly suitable for Al and low-Pb, low-Sn bronzes.—W. A. M. P.

***Fundamental Study on Precision Casting. I.—Consideration of a Trial Bronze Casting.** Kazuo Katori, Tsuneyuki Okakura, and Yotaro Shindo (*J. Mech. Lab.* (*Tokyo*), 1951, **5**, (7), 312–319).—[In Japanese]. The lost-wax process has been studied and is discussed. In the investigation beeswax patterns were cast in $CaSO_4$ moulds and natural, and heat-treated silica sands with a soln. of hydrolysed ethyl silicate in water as binder were used for coating and investment. The effects of several factors, including grain-size of the coating material, rate of removal of the wax pattern, heating speed of the mould, and surface hardness of the coating on the metal surface are reported qual. Effects of *d* of the binding material, grain distributions of filler, impurities, and drying conditions on the cracking of the coating are also reported. It was found that if the coating is stable, hardened, and strengthened, control of the properties of the backing investment is not always important and necessary, but if the coating is unstable control of the backing investment is essential.—AUTHORS.

***Fundamental Study on Precision Casting. II.—Some Influences Upon Cracking and Stripping of Refractory Coatings.** Kazuo Katori, Tsuneyuki Okakura, and Tomohiko Maruo (*J. Mech. Lab.* (*Tokyo*), 1952, **6**, (1), 1–8).—[In Japanese]. Cf. *ibid.*, 1951, **5**, 312; preceding abstract. Results of experiments on the cracking and stripping of the refractory coating before pouring are reported and discussed. It has been found that: (i) cracking occurs mainly during air drying and baking processes; the molten metal that enters develops cracking patterns on the casting surfaces; (ii) cracking is caused by the colloidal condition of the coating grains, rapid drying, and the high concentration of hydrolysed silicic acid; (iii) cracking at elevated temp. is mainly due to combined water content; and (iv) a low adhesive power of the coating to investment, a thin coating, and thermal expansion cause stripping. It is suggested that the use of coating material contg. no combined water will prevent cracking, and the use of multiple coatings will prevent stripping.—AUTHORS.

How Northrop Aircraft Vacuum-Impregnates Metal Castings. Thomas A. Dickinson (*Steel*, 1952, **131**, (2), 88–90).—An account is given of the process of impregnating Al and Cu alloy castings with sodium silicate, vinyl lacquers, styrene, and phenolic resins.—E. J.

The Arcair Torch [for Metal Removal]. —— (*Canad. Metals*, 1952, **15**, (1), 26–27).—A new technique for removing unwanted metal is described. A compact gouging and cutting torch is used, which combines the functions of an arc and air blast in a single unit. No gas cylinders are necessary.
—W. A. M. P.

Sand Treatment. A. S. Beech (*Canad. Metals*, 1952, **15**, (2), 28, 30–31).—B. discusses the relative merits of batch and continuously operating sand mills. It is considered that the continuous type of sand mill shows definite advantages over the batch mill.—W. A. M. P.

Automatic Core Maker Boosts Productivity. —— (*Canad. Metals*, 1952, **15**, (6), 38).—A new automatic method for core making is described.—W. A. M. P.

Foundry Mechanization. A. S. Beech (*Canad. Metals*, 1952, **15**, (3), 34, 36, 39).—A description of items of foundry plant used in Great Britain and elsewhere which are considered superior to those used in Canada and the United States.—W. A. M. P.

Modernized Foundry Boosts Castings Output. Robert H. Herrmann (*Foundry*, 1952, **80**, (7), 100–105, 216, 218).—A detailed description is given of the plant and equipment of the modernized foundry of the Dodge Mfg. Co., Mishawaka, Ind., U.S.A.—R. W. R.

17 — FURNACES, FUELS, AND REFRACTORIES

Melting [Equipment for] Aluminium and Magnesium-Base Alloys. L. W. Eastwood (*Trans. Amer. Found. Soc.*, 1951, **59**, 139–150; discussion, 150).—Melting, running, and prodn. characteristics of elect. and fuel-fired furnaces normally employed for melting light alloys are described.—V. K.

Frequency Conversion Might Well Affect Melting Technique. J. Morgan (*Canad. Metals*, 1952, **15**, (3), 32).—A brief note on some trouble experienced in casting Cu-base alloys. It was found that frequency conversion from 25 to 60 c./s. resulted in intense radiation from the elect. furnace arc which adversely affected the refractory lining, the lining being attacked by the metal. As the basic material of the lining was SiC, it was considered possible that Si was being absorbed into the melt, and the use of a flux to remove Si confirmed this view.—W. A. M. P.

A High-Frequency Induction Furnace for Melting Metals in a Vacuum and in Gaseous Media. M. G. Lozinsky (*Vestn. Mashinostroeniya*, 1950, **30**, (7), 47–53; *C. Abs.*, 1951, **45**, 475).—[In Russian]. Mech. and elect. details of the furnace are presented.

Rotary Melting Furnaces: Some Results Obtained with an Interesting Type of Equipment. —— (*Metallurgia*, 1952, **46**, (274), 99–100).—A brief description of the furnace and its operation.—F. M. L.

High Vacuum Melting. —— (*Canad. Metals*, 1952, **15**, (1), 31).—A description of a new Swiss high-vacuum plant for melting special steels, Cu alloys with readily oxidizable constituents, Ti, Zr, Mn, Cr, and their alloys. The sintering of Mo and W can also be carried out in the same type of plant. —W. A. M. P.

Automatic Furnace Controls. Leo Walter (*Canad. Metals*, 1952, **15**, (1), 20).—Different systems are briefly discussed. —W. A. M. P.

High-Frequency Heating. A. E. Williams (*Engineer*, 1952, **194**, (5038), 215–219).—A general survey of the appn. of H.F. heating to the heating and melting of metals, the selective hardening of steel, soldering, &c., and in the wood-using and plastics industries.—D. K. W.

High-Frequency Heating in Industry. K. A. Zandstra (*J. Trans. Soc. Eng.*, 1950, **41**, (4), 149–167; discussion, 167–172).—Z. discusses briefly the theory involved, the generation of H.F. currents, and the appn. of H.F. currents in industry.—S. R. W.

***Preparation of Refractories from Uranium Dioxide.** R. E. Corwin and G. B. Eyerly (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECD-3349**), 14 pp.).—UO_2 crucibles may be formed by pressing a powder contg. 6–8% distilled water at a pressure of 6–20 tons/in^2. or by slip casting after leaching in 5N-HCl. For the pressing operation, only a very slight taper on the die can be tolerated, as the material is extremely fragile, and it is preferable to incorporate a support for the base in the form of a Cr-plated brass ring bevelled at 45° where it makes contact with the crucible. The leached powder for slip casting is washed to give a pH of 1·5–2·1 and an addn. of 2 wt.-% carbo-wax will improve the dry strength. After the crucibles have been dried for several days at room temp., they are fired in a Mo resistance furnace at a max. temp. of 1750° C. in a purified H atmosphere. The slip castings are raised to 600° C. in 8 hr. to volatilize the carbowax and to 1750° C. in 20 hr. After a soak for 2 hr., they are cooled to 50° C. before removal from the furnace and stored in an evacuated desiccator. The linear shrinkage on firing is $\sim$14·8% for pressed ware and $\sim$15·2% for cast crucibles. The variation of d with the firing treatment is given, and for the above firing conditions, d of pressed and cast samples are 9·13 and 9·6 g./c.c., resp., compared with a theoretical value of 10·9. Notes are included on the necessary health precautions in handling UO_2.—B. W. M.

A Note on Fused Stabilized Zirconia: A Modern High-Temperature Refractory. N. Clarke Jones (*J. Inst. Fuel*, 1952, **25**, (142), 66–67; and *Ceramics*, 1952, **4**, (38), 75–78).—The manufacture, properties, and appn. are given.—S. R. W.

Castable Refractories for the Foundry. L. F. Train (*Canad. Metals*, 1952, **15**, (5), 36–37, 42).—T. discusses the appn. of castable refractories.—W. A. M. P.

18 — HEAT-TREATMENT

***A Study of the Annealing Kinetics in Cold-Worked Copper.** (Bowen, Eggleston, and Kropschot). See col. 321.

Bright-Annealing Copper in Continuous Controlled-Atmosphere Furnace. J. H. Bradley (*Canad. Metals*, 1952, **15**, (3), 56, 58).—A continuous controlled-atmosphere furnace for the bright annealing of Cu and Cu alloys is described. —W. A. M. P.

Induction Heating for Continuous Heat-Treatment of Sheet and Strip. F. C. Hull and Howard Scott (*Metal Progress*, 1952, **61**, (2), 57–61).—The article is mainly devoted to ferrous material and its processing; brass and Al alloys are mentioned. In treating non-ferrous materials the induction flux must be transverse and not longitudinal to the direction of the strip, and transverse flux induction heating heads were developed. 54-in. wide sheet or strip can be heat-treated. —F. A. F.

Low-Temperature Heat-Treatment [Furnace]. —— (*Metal Ind.*, 1951, **79**, (18), 381–382).—A forced-air-circulation furnace for use at temp. up to 700° C. is described.—N. B. V.

Direct Resistance Heating [for Heat-Treatment of Metals]. Lewis B. Reed (*Steel*, 1952, **131**, (2), 86–87).—E. J.

Transient Thermal Stresses [in Heat-Treatment]. M. Heisler (*U.S. Atomic Energy Commission Publ.*, 1952, (**NAA-SR-176**), 33 pp.).—H. gives a math. treatment for the stresses induced in large plates or shells for a number of heating and cooling processes, including heating by: (*a*) H.F. induction, (*b*) const. ambient temp., (*c*) uniformly varying ambient temp., and (*d*) internal heat sources. It is concluded that the abs. magnitude of the stress is the same for (*a*) and (*b*) and for (*c*) and (*d*), and that the temp. and stress equations for (*c*) and (*d*) can be obtained from a simple integration of (*a*) and (*b*). Graphs are included to facilitate the numerical evaluation of the stresses. 23 ref.—B. W. M.

19 — WORKING

Tapered Aircraft [Aluminium Alloy] Skins Machined. —— (*Steel*, 1952, **131**, (8), 98–99).—The prodn. of smooth tapered skins in Al alloy sheet for aircraft, using abrasive belts, is described.—E. J.

Precision Spinning [of Aluminium]. Bernard Blackman (*Steel*, 1952, **130**, (12), 68–69).—Describes how parts of instruments are made in Al by combining watch-case spinning techniques and die stamping.—E. J.

A Spinning Machine for the Manufacture of [Aluminium] Reflectors. Raoul Gasser (*Rev. Aluminium*, 1952, (185), 70).—Describes a machine which produces reflectors in high-purity Al by spinning and also polishes the reflecting surface. —A. W. B.

Sawing of Aluminium Alloys. Gaston Laval and René Schweyckart (*Rev. Aluminium*, 1952, (185), 71–78; (186), 111–116; (187), 153–163).—Various types of circular and band saws are described, and details are given of tooth spacing, cutting angles, &c. Figures are then given for recommended cutting speeds appropriate to various alloys, and suitable cutting lubricants are mentioned. For circular saws a high-speed steel with a hardness around Rockwell 60C is preferred. With a double-relieved blade thicknesses up to 300 mm. dia. can be cut with a blade 2 mm. thick. The latter part of the account is devoted to a detailed description of various blades and saws, including small portable saws for site use and large automatic saws, either mech. or hydraulically operated, such as would be used for billet cutting.—A. W. B.

Fabricating Aluminium Washing Machine Tubs. —— (*Canad. Metals*, 1952, **15**, (2), 46).—A brief note.—W. A. M. P.

Titanium Sheet-Metal Parts Successfully Made. Andrew N. Eshman (*Iron Age*, 1952, **170**, (3), 132–135).—Commercially pure RC-70 and RC-130A Ti alloys are fabricated for fuselage sections in place of stainless steel where operating temp. are ≯800° F. (425 ° C.).—J. H. W.

Titanium: It Can Be Forged and Drawn. Carter C. Higgins (*Steel*, 1952, **131**, (10), 92–93).—Using the same dies as those used for other materials, Ti can often be cold pressed satisfactorily. Existing Ti alloys cannot be deep-pressed, however, and more pressure is usually required. H. also details other properties of Ti and its alloys.—E. J.

Calculations on the Influence of Friction and Die Geometry in Sheet Drawing. A. P. Green and R. Hill (*J. Mechanics Physics Solids*, 1952, **1**, (1), 31–36).—A review is first given of the theory of sheet drawing through smooth dies. The results have been extended in some cases empirically, to determine the limits of reduction, and are summarized by empirical formulæ. The pressure distribution over the die is also discussed, and finally rough dies are considered, with empirical formulæ for drawing stress and die pressure.—E. O. H.

Springback: Problem in Metal [Sheet] Forming. Frederico Strasser (*Steel*, 1952, **131**, (5), 90–91).—The factors are considered upon which the amount of springback depends during forming and bending of sheet. Various remedies such as overbending, restriking, and the use of special dies are given. —E. J.

Does New Process Challenge Foundries? Lorne Frame (*Canad. Metals*, 1952, **15**, (3), 28–29).—A cored forging and pressing process is described. Several advantages are claimed over casting and forging: (1) no waste metal, (2) no rejects, (3) machining time reduced or eliminated, (4) parts can be made lighter and smaller, yet of higher tensile strength, and (5) complexity of design is only limited by the die.—W. A. M. P.

Saving Material by Cold Forming. J. R. Fawcett (*Mech. World*, 1952, **131**, (3393), 168–170).—The advantages are discussed of cold heading, cold extrusion, and other cold-working operations for the prodn. of small metal parts. —R. W. R.

***Calculation of the Forces Operative in a Rolling Mill.** A. Geleji (*Acta Techn. Acad. Sci. Hungar.*, 1951, **2**, (1), 123-142). —[In German]. In continuation of previous work (*ibid.*; 1950, **1**, 78; *M.A.*, **19**, 155) formulæ are derived for calculating the resistance to change of shape, the operative pressure, and the power requirements of a rolling mill. A comparatively simple method for determining the distribution of the resistance to deformation is developed, and it is shown how roll-turning moment, together with other characteristics of the rolling process, can be derived from the diagram representing the distribution of the resistance to deformation.—J. S. G. T.

***Measurement of Strip Tension in Tandem Mills.** R. B. Sims. (*Engineering*, 1952, **174**, (4517), 232–233; (4518), 262–264).—Cf. *M.A.*, **19**, 156, 323. A relationship between the tension in a strip passing inelastically over a roller, and the resulting thrust on the roller, is deduced, and shown to be in reasonable agreement with experimental results. The theory is used in the design and construction of a deflection roller-type tensiometer, which has been calibrated against a standard type instrument for strip thicknesses varying from 0·015 to 0·110 in. An accuracy of ±5% is obtained over the complete range. An indust. form of the tensiometer for dealing with strip up to 56 in. wide, is described.—D. K. W.

***Study on the Back-Tension Drawing Method. II.—The N.N.S.-Type Back-Tension Drawing Machine.** Kenichi Nakamura and Osamu Nishimura (*J. Mech. Lab.* (*Tokyo*), 1950, **4**, (3), 118–125).—[In Japanese]. The power necessary for drawing, the angular velocity of the drawing drum and the drum giving back tension, and the relation between the drawing force and the back tension force in a N.N.S.-type back-tension drawing machine have been calculated. Results show that: (i) the relation of the angular velocities varies with the reduction ratio, (ii) the back tension/drawing force ratio remains const., and (iii) less power is required in this type of machine than in a machine in which back tension is supplied by a brake or dead weight.—AUTHORS.

Machinability of Metals. Francis W. Boulger (*Canad. Metals*, 1952, **15**, (2), 52).—A brief discussion of the properties of metals and alloys considered to be important in machining. —W. A. M. P.

***The Size-Effect in Metal Cuttings.** W. R. Backer, E. R. Marshall, and M. C. Shaw (*Trans. Amer. Soc. Mech. Eng.*, 1952, **74**, (1), 61–71; discussion, 71–72).—A cutting process involving the formation of very small chips at high cutting speeds (micromilling) was investigated, and the results are applied to the study of the grinding operation. The shear energy involved in grinding is compared with that in turning, micromilling, and the tensile test, and a significant increase in shear energy is observed with decrease in specimen (chip) size. The shear stress involved in grinding metals under mild conditions is found to correspond to the theoretical strength, which is $\sim 1{\cdot}8 \times 10^6$ lb./in.^{2} for steel. The grit depth of cut is shown to be a more important variable in the interpretation of grinding data than the mere conventional wheel depth of cut.—H. PL.

***Cutting Temperature and Metal-Cutting Phenomena.** B. T. Chao and K. J. Trigger (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (8), 777–787; discussion, 787–793).—Many changes during the cutting of metals are due to the change in tool/chip interface temp. through its influence on the tool/chip friction. Cutting forces and cutting temp. were observed during conventional turning and orthogonal cutting under identical conditions, and the higher heat-dissipating capacity of the tool in orthogonal cutting operations was found to be fundamentally responsible for the observed differences. The tool/chip contact area is an important factor influencing interface temp. A decrease in contact area is associated with an increase in the interface temp. This area is larger in orthogonal cutting than in conventional turning at const. cutting speed. When used for cutting steel, WC tools cause higher cutting forces, higher coeff. of friction, larger tool/chip contact area, and lower interface temp. than triple carbide tools under otherwise const. cutting conditions. —H. PL.

***An Analytical Evaluation of Metal-Cutting Temperatures.** K. J. Trigger and B. T. Chao (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (1), 57–66; discussion, 66–68).—The average tool/chip interface temp. is calculated by considering the mechanism of heat generated during metal-cutting operations in which a type 2 chip is formed. The analytical results agree well with those obtained by test methods, and the theoretical analysis has yielded some important phys. quantities and measurements which affect cutting temp. 18 ref.—H. PL.

***Basic Factors in Hot Machining of Metals.** E. J. Krabacher and M. E. Merchant (*Trans. Amer. Soc. Mech. Eng.*, 1951, **73**, (8), 761–768; discussion, 768–776).—Two main factors oppose each other in their effect on tool life in hot machining. These are the increased ease with which the heat-softened tool material can be abraded away and the decreased ability of the heat-softened chip and the work-piece to produce such abrasion. It was found that these two factors can be evaluated roughly from tool-temp. data taken in combination with data on the mechanics of cutting. In the discussion some criticism is made of the use of Trigger and Chao's (*ibid.*, (1), 57; preceding abstract) equation for calculating the chip/tool interface temp., as it is argued that the carbide tool is submerged only for short intervals of time. —H. PL.

***Researches on the Lapping of Metals.** Hideo Tsuwa (*Technol. Rep. Osaka Univ.*, 1952, 2, (29), 53–60).—[In English]. The surface lapping of steel and brass, using carborundum, green carborundum, alundum, and white alundum as abrasives was studied, and the following results were obtained: in the first stage of the lapping process the grain-sizes of the abrasive are reduced, the number of effective abrasive grains increases, and lapping removal efficiency increases; in the second stage, grain-sizes in the lapping powder become equal, and lapping removal efficiency attains its max. value; in the third stage, the lapping powder is crushed down to its finest grain-size, "loading" takes place, and lapping removal efficiency is reduced. The extreme form of the pressure-lapping removal graph is interpreted in terms of "loading," which is heavier in brass than in steel. —J. S. G. T.

Electronics Applied to Machining Hard Metals. —— (*J. Metals*, 1952, 4, (8), 822–824).—The Sparcatron process of machining by elect. spark discharge is described and illustrated. —E. N.

---

24 — BIBLIOGRAPHY

(Publications marked * have been received.)

Andrade, E. N. da C. *Viscosity and Plasticity.* 6 × 9 in. Pp. 82. 1951. New York: Chemical Publishing Co., Inc. ($2.25.)

***Anglo-American Council on Productivity.** *Zinc and Aluminium Die-Casting.* Report of a Productivity Team Representing the British Zinc and Aluminium Die-Casting Industry which Visited the United States of America in 1951. 9¾ × 7¼ in. Pp. x + 98, with 31 illustrations. 1952. London: The Council, 21 Tothill Street, S.W.1; Oxford: The Zinc Alloy Die-Casters Association, Lincoln House, Turl Street; Birmingham (2): The Light Metal Founders' Association, 25 Bennetts Hill. (4*s.*, post free.)

***British Iron and Steel Research Association.** *The Fight Against Rust. The Work of the Corrosion Committee of the British Iron and Steel Research Association.* 9½ × 6¼ in. Pp. 25, with 19 illustrations. 1952. London: The Association, 11 Park Lane, W.1. (2*s.* 6*d.*, post free.)

***Delbart, Georges.** *Les Laboratoires de l'Institut de Recherches de la Sidérurgie Française.* 27 × 21 cm. Pp. 48, with 39 illustrations. 1952. St. Germain-en-Laye (S. et O.): L'Institut.

***Electrodepositors' Technical Society.** *Journal of the Electrodepositors' Technical Society.* Vol. XXVII. 1950–51. 8½ × 5½ in. Pp. [viii] + 342, illustrated. [1952.] London: The Society, 32 Great Ormond Street, W.C.1. (£2 2*s.*)

[Contents: W. de Bruijn, "Throwing and Covering Power in Electroplating Solutions" (*M.A.*, 20, 112); J. W. Cuthbertson, "Recent Developments in Tin and Tin Alloy Plating" (*M.A.*, 20, 111); H. Saenger, "The Storage and Handling of Acids and Cyanides" (*M.A.*, 20, 114); T. Dennison, "Surface Defects in Steel and Their Relationship to the Different Stages of the Manufacturing Process" (*M.A.*, 20, 119); S. C. Shome and U. R. Evans, "Studies in the Discontinuities of Electrodeposited Metallic Coatings—II, —III" (*M.A.*, 18, 715; 19, 139); W. A. Bell, G. J. Metcalfe, and A. H. Sully, "Some Factors Affecting the Rate of Diffusion of Hydrogen Through Steel During Electrolytic Pickling"; F. A. Lowenheim and R. M. MacIntosh, "The Development of Tin–Zinc Alloy Plating in the U.S.A."; N. Parkinson, "The Electrodeposition of Bright Tin–Nickel Alloy Plate" (*M.A.*, 19, 138); H. P. Rooksby, "An X-Ray Study of Tin–Nickel Electrodeposits" (*M.A.*, 19, 122); S. Tolansky, "Some Applications of Interferometry to the Examination of an Electrodeposited Film" (*M.A.*, 19, 150); W. A. Hothersall and G. E. Gardam, "Electrodeposition of Nickel in the Bores of Tubes Using Insoluble Anodes" (*M.A.*, 19, 136); A. W. Wallbank and G. W. Airey, "Hard Chromium Plating of Large Cast-Iron Drum-Drier Rolls" (*M.A.*, 18, 713); G. I. Finch and D. N. Layton, "The Crystal Structure of Metallic Electrodeposits" (*M.A.*, 18, 706); G. L. J. Bailey, "Electrodeposition on Aluminium: A Study of the Zincate Process" (*M.A.*, 19, 135); F. H. Burstall, N. F. Kembar, and R. A. Wells, "Inorganic Chromatography on Cellulose: Some Applications to the Analysis of Nickel-Plating Baths" (*M.A.*, 19, 137); A. Smart, "Factors Influencing the Design of Automatic Plating Plant"; S. C. Britton and R. M. Angles, "Corrosion-Resistance of Electrodeposited Tin–Nickel Alloy".]

Geleji, Alexander. Herausgegeben von. *Die Technologie der Halbfabrikate aus Nichteisenmetallen.* 21 × 15·5 cm. Pp. 434, with 417 illustrations. 1952. Budapest: Akadémiai Kiadó. (60 forint.)

***Henry, Norman F. M., and Kathleen Lonsdale.** *International Tables for X-Ray Crystallography.* Volume I.—*Symmetry Groups.* 11 × 8½ in. Pp. xi + 558, illustrated. 1952. Birmingham (6): The Kynoch Press (for the International Union of Crystallography). (£5 5*s.*)

***Johnson, L. W., and E. J. Bradbury.** *Corrosion-Resistant Materials in Marine Engineering.* 9¾ × 6½ in. Pp. [iii] + 39, with 8 illustrations. [1952.] London: The Mond Nickel Co., Ltd., Sunderland House, Curzon Street, W.1. (Gratis.)

[Based on a paper presented to the Institute of Marine Engineers; see *Trans. Inst. Marine Eng.*, 1951, 63, (4), 59–74; *M.A.*, 19, 728.]

***Ministry of Education.** *Metalwork in Secondary Schools.* (Pamphlet No. 22). 8½ × 5½ in. Pp. iv + 55, illustrated. 1952. London: H.M. Stationery Office. (3*s.* 6*d.*)

***Morley, Arthur.** *Strength of Materials.* Tenth edition. 8½ × 5½ in. Pp. x + 583, with 265 illustrations. 1952. London: Longmans Green and Co. (25*s.*)

***Rollefson, G. K.** Edited by. *Annual Review of Physical Chemistry.* Vol. 3. 9 × 6 in. Pp. x + 416, illustrated. 1952. Stanford, Calif.: Annual Reviews, Inc. ($6.00); London: H. K. Lewis and Co., Ltd., 136 Gower Street, W.C.1. (48*s.*)

[Contents: C. A. Coulson, "Quantum Theory, Theory of Molecular Structure and Valence"; M. G. Mayer, "Radioactivity and Nuclear Theory"; J. Bigeleisen, "Isotopes"; A. O. Allen, "Radiation Chemistry"; P. Doty and G. Ehrlich, "Polymeric Electrolytes"; W. C. Bauman, R. E. Anderson, and R. M. Wheaton, "Ion Exchange"; R. F. Newton, "Thermochemistry and the Thermodynamic Properties of Substances"; G. Glockler, "Bond Energies and Bond Distances"; C. J. Christensen and E. Roedder, "Heterogeneous Equilibria and Phase Diagrams"; G. K. Rollefson, "Photochemistry"; M. Calvin, J. A. Bassham, A. A. Benson, and P. Massini, "Photosynthesis"; N. C. Bayliss, "Spectroscopy"; C. Tanford and S. Wawzonek, "Polarography and Electrode Processes"; G. Scatchard, "Solutions of Non-Electrolytes"; T. F. Young and A. C. Jones, "Solutions of Electrolytes"; R. E. Powell, "Reaction Kinetics"; F. E. Williams, "Theory of Ionic Crystals, Semi-Conductors, and Dielectrics"; W. H. Zachariasen, "Experimental Crystallography"; L. O. Brockway, "Experimental Molecular Structure".]

Verö, J. *Vas- és fémipari anyagvizsgálat.* (*Industrial Testing of Ferrous and Non-Ferrous Metals*). 21 × 15·5 cm. Pp. 216, illustrated. 1951. Budapest: Tankönyvkiadó.

***Wulff, John, Howard F. Taylor, and Amos J. Shaler.** *Metallurgy for Engineers: Casting, Welding, and Working.* 9 × 6 in. Pp. viii + 624, illustrated. 1952. New York: John Wiley and Sons, Inc. ($6.75); London: Chapman and Hall, Ltd. (54*s.*)

NEW JOURNAL

Journal of the Mechanics and Physics of Solids. Published quarterly by Pergamon Press, Ltd., 2, 3, and 5 Studio Place, Kinnerton Street, London, S.W.1. First issue appeared in October 1952. (£4 10*s.* per annum.)

25 — BOOK REVIEWS

Studies in Large Plastic Flow and Fracture, with Special Emphasis on the Effects of Hydrostatic Pressure. By P. W. Bridgman. (Metallurgy and Metallurgical Engineering Series). 9 × 6 in. Pp. x + 362, with 169 illustrations. 1952. New York: McGraw-Hill Book Co., Inc. ($8.00); London: McGraw-Hill Publishing Co., Ltd. (76*s.* 6*d.*)

In this volume is collected the work of Professor Bridgman on plastic deformation and fracture of many kinds of materials under a wide range of complex stress conditions. It is very natural that much of the work involved a superimposed hydrostatic pressure, since Bridgman's name has become automatically associated in most scientists' minds with the investigation of high-pressure phenomena.

Much of the work has been reported elsewhere, but has not been readily available to the general public. This book, which collects the results of Bridgman's experiments over at least the last twelve years and presents them in a co-ordinated whole, is welcome for this reason alone. The value of the book, however, goes much further than this. For in submitting materials to various stress combinations and superimposing hydrostatic compressive stresses of such magnitude as only he is capable of, Bridgman has opened a new field of investigation. While his results are not such that they can be used to confirm or contradict the fundamental theory of plasticity, they suggest a very great range of profitable experimental investigations.

Bridgman has confined himself to reporting the results of his own work, and the reader must not look for a critical survey of the work of others in the field of plastic deformation.

The early chapters describe a very great number of tensile tests, both simple tension and two-dimensional, under hydrostatic pressure, and the now-famous work on necking under pressure is dealt with in detail.

A great range of compression tests are reported also, and such processes as ball hardness testing, punching, wire drawing, and extrusion, all under high pressure, form a fascinating study.

The later chapters give an account of experiments in which materials were subjected to various stress combinations after some measure of prestraining had been previously carried out. A final chapter called "Gathering up the Threads" makes remarkably good reading and shows deep insight into the problems and the fundamentals of plastic flow.

Hugh Ford.

Metallurgical Engineering. Volume I.—**Engineering Principles.** By Reinhardt Schuhmann, Jr. (Addison-Wesley Metallurgical Series). 9½ × 7¼ in. Pp. ix + 390, illustrated. 1952. Cambridge (42), Mass.: Addison-Wesley Press, Inc. ($7.50.)

In these days of international standardization one can almost envisage a conference called to discuss the various definitions of Metallurgical Engineering. The present book, based on lectures to students at the Massachusetts Institute of Technology, is the result of a recognition of the need to reorganize instruction in extractive metallurgy. This change of outlook is apparently partly due to the tendency of chemical engineers to be more useful than metallurgists even in the extractive metallurgical industry, owing to their better grasp of basic principles and their ability to apply these principles quantitatively to practical problems. The three objectives of the book are to present the subject according to principle rather than metal treated, to integrate the underlying science of physical chemistry with metallurgical engineering, and to place greater emphasis on analysis and quantitative solution of problems.

With regard to the subject matter of the book, a general discussion of the unit processes of chemical metallurgy is followed by a consideration of the application of some of the elementary physical and chemical laws to various processes to obtain a materials balance and a heat balance. Further chapters are devoted to the properties of various fuels, combustion conditions, heat utilization, and recuperator design, as well as problems of heat flow under steady and fluctuating conditions. The inclusion of a chapter on fluid flow indicates a proper appreciation of the importance of this subject in view of the number of processes where tremendous quantities of gases are involved. The properties of phases encountered in pyrometallurgical operations are discussed, including questions of gas solubility in metals and the nature and the behaviour of slags. The final chapter gives a useful account of refractory materials and the factors which influence their selection under different conditions.

On the whole, the author successfully avoids the inherent danger in a book of this kind of introducing difficult scientific ideas to explain technical matters, but it is felt that there is a certain inconsistency in the amount of knowledge which the student is expected to possess. For example, concepts such as kinetic and potential energy are explained, whereas a knowledge of the phase rule and the principles of equilibrium diagrams are assumed. The book is intended for students, and there are numerous worked examples illustrating the various principles with plenty of problems to be solved at the end of each chapter and adequate references for further reading. The topics which are included are in general well presented, and the outlook of the author is to be commended, but the scope of the book is almost inevitably not as comprehensive as the title might suggest.

N. Swindells.

Corrosion: Causes and Prevention. By Frank N. Speller. Third edition, revised. 9 × 6 in. Pp. xiii + 686, with 181 illustrations. 1951. New York: McGraw-Hill Book Co., Inc. ($10.00); London: McGraw-Hill Publishing Co., Ltd. (85*s.*)

Those familiar with the earlier editions of this book will have no difficulty in recognizing their successor. The same general treatment has been used, although chapters on biological influences and on cathodic protection have been added and an attempt has been made to bring the information up to date, or at least up to 1948. Strangers to the earlier editions need to know that the book is concerned almost wholly with ferrous metals, non-ferrous metals appearing only as coating materials for steel; that it is divided into an outline of general principles and a description of the application of knowledge to some practical problems; and also that it is well written in terms easily understood by those without specialized knowledge.

By the clear presentation, in previous editions, of material not then collected elsewhere, Dr. Speller did much to promote the vigorous growth of the subject which has made his own task in preparing a new account an extremely difficult one. The revision has not been sufficiently drastic to produce a well-balanced picture of our present knowledge of corrosion. This is partly due to an excessive reliance on American sources, though some matters commercially important on both sides of the Atlantic receive only slight attention. For instance, tinplate is mentioned only as a possible roofing material. Again, the author's evident desire to show historical developments has led to the retention of too much detail about early work to permit the clear presentation of recent findings, and sometimes the more recent ideas are not adequately related to some of the older observations. Thus, although biological influences are dealt with in a separate chapter, their probable contribution to many of the results quoted in the chapter on corrosion by soil is not mentioned.

Even if the picture is incomplete and not too well balanced, however, the general outline is good, and some of the detail is very clear, as in the sections dealing with steam-raising plant and heating systems. Although the specialist corrosion worker will not welcome this edition as warmly as he did its predecessors, the engineer, the builder, and the architect will find most of the knowledge they need about the corrosion of ferrous metals provided in terms they can understand.

S. C. Britton.

Gespritzte Metallüberzüge : Erzeugung, Gefüge, Eigenschaften und praktische Nutzanwendung. Von Hans Reininger. 21 × 14·8 cm. Pp. 246, with 151 illustrations. 1952. München : Carl Hanser-Verlag. (DM 14.80.)

This book covers very fully the whole technique of metal spraying. Its first section describes, and traces the development of, the various types of metal-spraying pistol, gives performance data for several of them, and then briefly describes auxiliary equipment such as sand-blasting apparatus. In the second section are discussed the structure of sprayed metal coatings, their adhesion to the basis metal, various methods of preparing the basis metal before spraying, and different methods of treating the coating after spraying. The third section deals with the properties of sprayed-metal coatings, and the fourth is devoted to the practical uses to which such coatings are put. The corrosion-resistance of various sprayed metals is described in detail. The book is liberally illustrated with pictures and diagrams and contains a bibliography of 223 references. It is well-produced and well-indexed and will no doubt come to be regarded as a standard work on the subject of metal spraying.

G. T. Colegate.

Electric Arc and Oxy-Acetylene Welding : A Practical Handbook for Works Managers, Welding Operators, and Students. By E. A. Atkins and A. G. Walker. Fourth edition, revised by W. A. Atkins and A. G. Walker. 8½ × 5½ in. Pp. xi + 352, with 186 illustrations. 1952. London: Sir Isaac Pitman and Sons, Ltd. (30*s.*)

For a book to reach a fourth edition after a first edition some thirty years earlier is a considerable achievement, and the new edition must be looked at in this light. The authors have taken advantage of the opportunity afforded to make some modifications, but the general layout of the book is not markedly altered. There are the same two chapters of admonitions to electric and to oxy-acetylene welders, which include the danger to the former of wearing celluloid collars, and to the latter of looking for acetylene leaks with a light. There are added chapters to cover the welding of the new metals and alloys, and also the training of welders. The list of questions at the end has been cut down to those from City and Guilds papers for welders. It is a pity that the selection given in the earlier edition has been retained in preference to ones set more recently.

Written essentially for the practical welder, the book aims to give him advice on the many practical problems he may meet, and to enable him to sit for the City and Guilds examination. The book does not follow a very clear plan except for the division between electric and oxy-acetylene welding, and the main criticism that can be made against even this new edition is that the text lacks clarity. The welder of a copper joint, for example, is likely to be a little puzzled when advised to find the correct neutral zone and only to move when it is necessary to exclude unwanted atmospheric conditions. There is much sound advice from two writers who have obviously had considerable first-hand experience, but because of the number of chapters on Do's and Don'ts and Practical Hints, the required advice may be a little difficult to find. The practical aspect is taken a little too far by the inclusion of details of a home-made welding screen using plywood, and of the series of tests to be carried out to decide on the nature of the gas contained in a cylinder which has lost its identity. Too often the words fracture and repair appear, as though the authors think of welding more for reclamation than for construction.

The metallography of welds is covered in some detail in a chapter entitled " Weld Defects and Their Testing ", but no magnifications are quoted for any of the illustrations. Some of these micrographs are rather misleading, as, for example, the structure of an oxy-acetylene weld in medium-carbon steel, which looks very like part of the section representing burning in a similar weld.

The authors are rather laboured on the advantages of using coated electrodes instead of bare wire for metallic arc welding. This may have been justified in the first edition, but not now. On the other hand, they dismiss the Heliarc process as a variation of atomic-hydrogen welding, and their only comment on the furnace brazing of steel with copper is that it has very interesting possibilities.

There is no doubt that the previous editions have given invaluable help to many welders, and it is likely that the present edition will do so too, but it is much to be regretted that the opportunity was not taken to include full details of new processes and equipment, and to omit those which are now obsolete.

F. A. Ball.

Lead in Modern Industry. 9½ × 6½ in. Pp. viii + 230, illustrated. 1952. New York (17): The Lead Industries Association, 420 Lexington Avenue. ($1.50, post paid.)

The book starts with a short history of lead and its mining, as well as extraction by smelting and refining processes. This is followed by a somewhat more detailed account of its fabrication by extrusion, rolling, and casting. Chapters are next devoted to the more important uses of lead, such as in the manufacture of storage batteries, electric cables, and the construction of chemical plant. The composition and uses of solders, bearing metals, and type metals are described, whilst space is also given to the subject of the use of lead for protection from radioactive materials.

The next section of the book deals with the use of lead compounds for purposes such as in the manufacture of paints and as an addition agent to petrol in the form of lead tetraethyl, and it also makes brief mention of newer applications, such as the use of lead compounds as stabilizers in plastics.

It ends with a list of American specifications and some detail of the properties of lead, lead alloys, and lead compounds. It is nicely printed and well illustrated with both drawings and photographs.

Quite naturally, the publication deals mainly with American practice, although there are a few references to work which has been done in this country. No claim is made that the work is complete, and there is, for instance, no mention of the use of the continuous extrusion press for sheathing cables; but it covers a very wide field, and there is some reference to most aspects of the uses of lead in industry to-day.

Altogether, the book is a useful and interesting account of the applications of lead in modern industry. It is useful because it puts together in one volume a large amount of information that hitherto has been scattered over a formidable amount of technical literature, and it is interesting because it gives a broad picture of the uses to which lead is being put to-day. It will be of general interest to those engaged in the lead industry or using its products, as well as to technical schools.

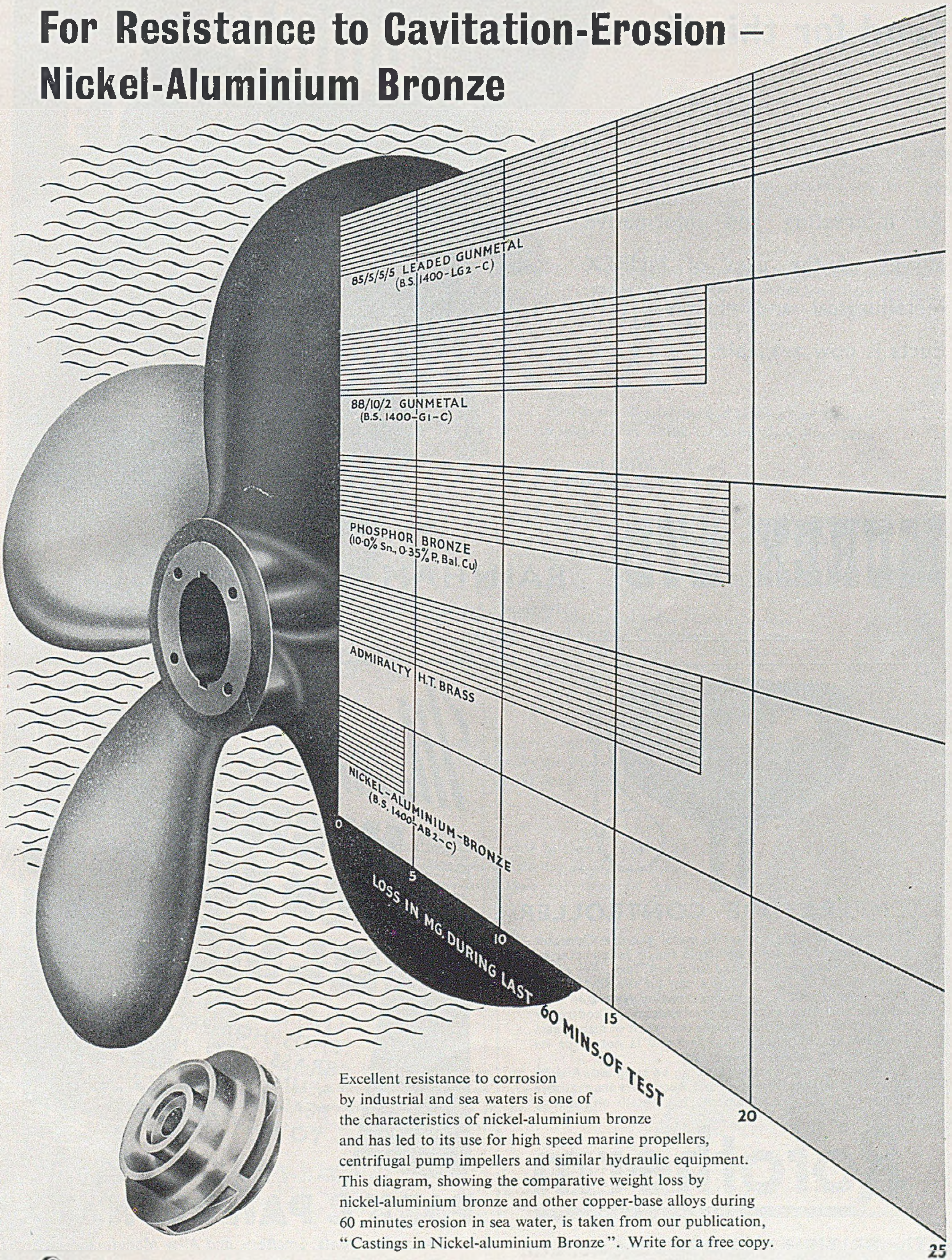
A. Makower.

METALLURGICAL ABSTRACTS

METALLURGICAL ABSTRACTS AUTHOR INDEX

12/NF/6

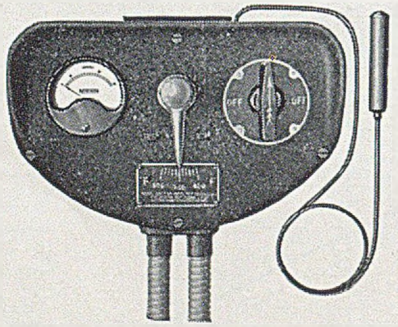
For Resistance to Cavitation-Erosion – Nickel-Aluminium Bronze

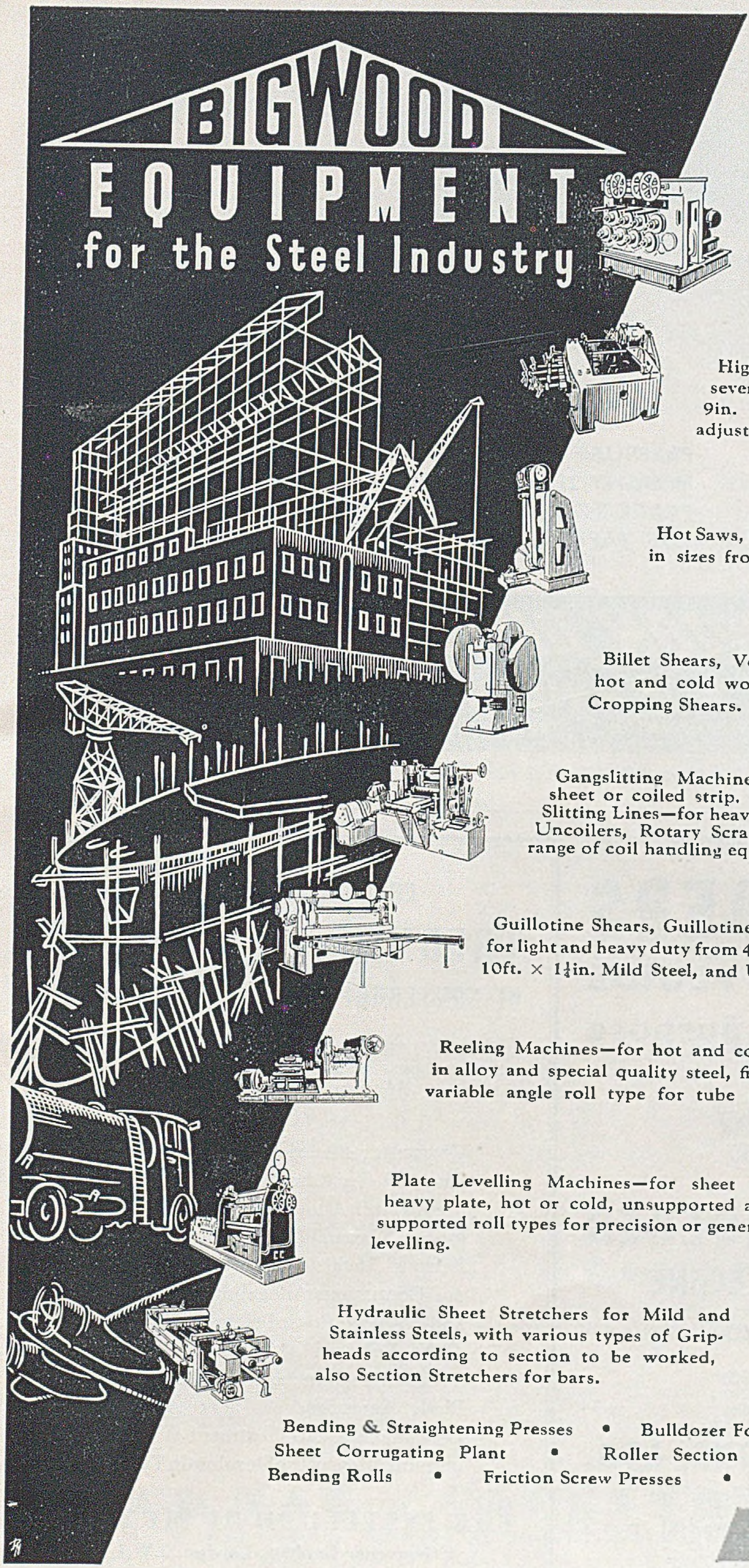


Excellent resistance to corrosion by industrial and sea waters is one of the characteristics of nickel-aluminium bronze and has led to its use for high speed marine propellers, centrifugal pump impellers and similar hydraulic equipment. This diagram, showing the comparative weight loss by nickel-aluminium bronze and other copper-base alloys during 60 minutes erosion in sea water, is taken from our publication, "Castings in Nickel-aluminium Bronze". Write for a free copy.

MOND NICKEL THE MOND NICKEL COMPANY LIMITED · SUNDERLAND HOUSE · CURZON STREET · LONDON · W1


MUREX LIMITED
RAINHAM · ESSEX
ENGLAND





BIGWOOD EQUIPMENT for the Steel Industry

Roller Section Straighteners—full range of high speed Roller Straighteners for bars and sections from 1⅛in. to 1½in. angles to 24in. × 7½in. joists.

High Speed Tube and Bar Straighteners—in several classes for bar from ¼in. diameter to 9in. diameter, with non-adjustable and adjustable angle driving rolls.

Hot Saws, Pendulum and Horizontal Types—in sizes from 36in. to 60in. Saw diameter.

Billet Shears, Vertical and Horizontal—for hot and cold work. Alligator Shears and Cropping Shears.

Gangslitting Machines,—for slitting from sheet or coiled strip. Complete Mill type Slitting Lines—for heavy coils. Coilers and Uncoilers, Rotary Scrap Cutters, and full range of coil handling equipment.

Guillotine Shears, Guillotine Plate Shears—for light and heavy duty from 4ft. × ⅛in. up to 10ft. × 1¼in. Mild Steel, and Upcut Shears.

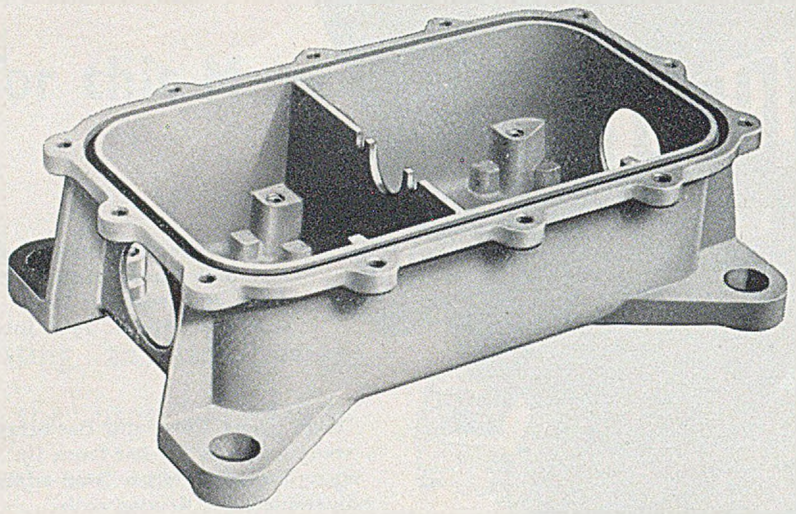
Reeling Machines—for hot and cold work, in alloy and special quality steel, fixed and variable angle roll type for tube or bar.

Plate Levelling Machines—for sheet and heavy plate, hot or cold, unsupported and supported roll types for precision or general levelling.

Hydraulic Sheet Stretchers for Mild and Stainless Steels, with various types of Grip-heads according to section to be worked, also Section Stretchers for bars.

Bending & Straightening Presses • Bulldozer Forging Machines • Drawbenches Sheet Corrugating Plant • Roller Section Forming Machines • Plate Bending Rolls • Friction Screw Presses • Centreless Bar Turning Machines.

JOSHUA BIGWOOD & SON LIMITED • WOLVERHAMPTON • ENGLAND



PRESSURE DIE CAST IN ALUMINIUM ALLOY WEIGHT 2¼ lbs. GASKET GROOVE CAST IN PLACE TOGETHER WITH ALL HOLES READY FOR TAPPING.

WOLVERHAMPTON DIE-CASTING CO. LTD.

GRAISELEY HILL • WOLVERHAMPTON

TELEPHONE: 23831/4 WOLVERHAMPTON. TELEGRAMS: DIECASTINGS, WOLVERHAMPTON

American Affiliation: PRECISION CASTINGS CO. INC. · FAYETTEVILLE · SYRACUSE · CLEVELAND · KALAMAZOO · CHICAGO

SOFT SOLDERS

AND SOLDERING FLUIDS

for every industrial purpose

Superspeed

ACTIVATED

ROSIN CORED SOLDER

“TRICENE” & “TELECENE”

Soldering Fluxes

Expert Technical Advice
on all Soldering Problems

H. J. ENTHOVEN & SONS, LTD.

ENTHOVEN HOUSE
89 UPPER THAMES STREET, LONDON, E.C.4

Telephone : MANsion House 4533

EQUIPMENT FOR THE

THERMAL TREATMENT

OF NON-FERROUS METALS AND ALLOYS

Cloth. 4to. 1952. Pp. 104, with 14 plates and numerous figures in the text. Price 15s. or $2.50, post free. (Members, one copy each at 7s. 6d. or $1.50, post free.)

CONTENTS: Electric Furnaces for the Thermal Treatment of Non-Ferrous Metals and Alloys; Gas Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys; Batch and Continuous Annealing of Copper and Copper Alloys; Bright Annealing of Nickel and its Alloys; Batch Thermal Treatment of Light Alloys; Flash Annealing of Light Alloys; Continuous Heat-Treatment of Aluminium Alloys of the Duralumin Type.

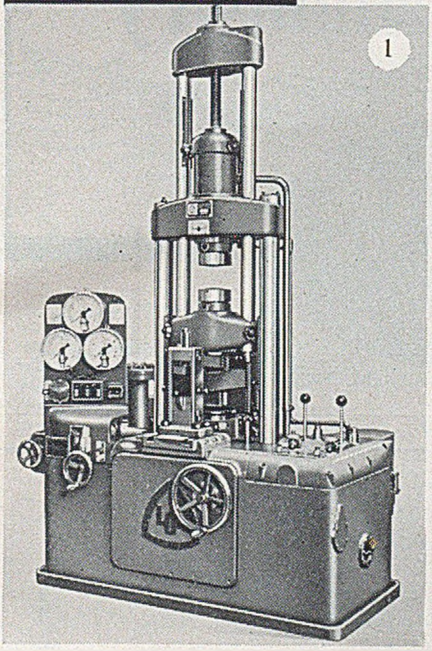
THE INSTITUTE OF METALS

4 Grosvenor Gardens, London, S.W.1

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.

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.


1

UNIVERSAL MATERIAL TESTING (1)

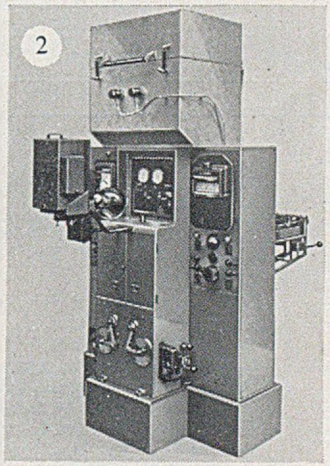
An outstanding feature of Losenhausenwerk machines for static and dynamic testing of materials is that actual components and complete assemblies can be tested under conditions as near as possible to those to which the parts are likely to be subjected while in use.

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.

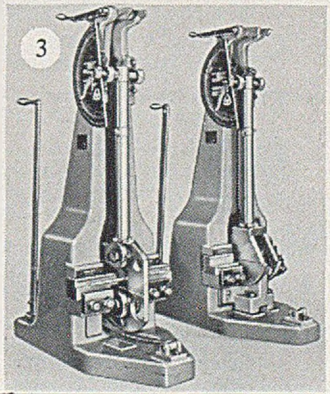
FOR STATIC AND DYNAMIC TESTS

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*.


2

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.


3

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

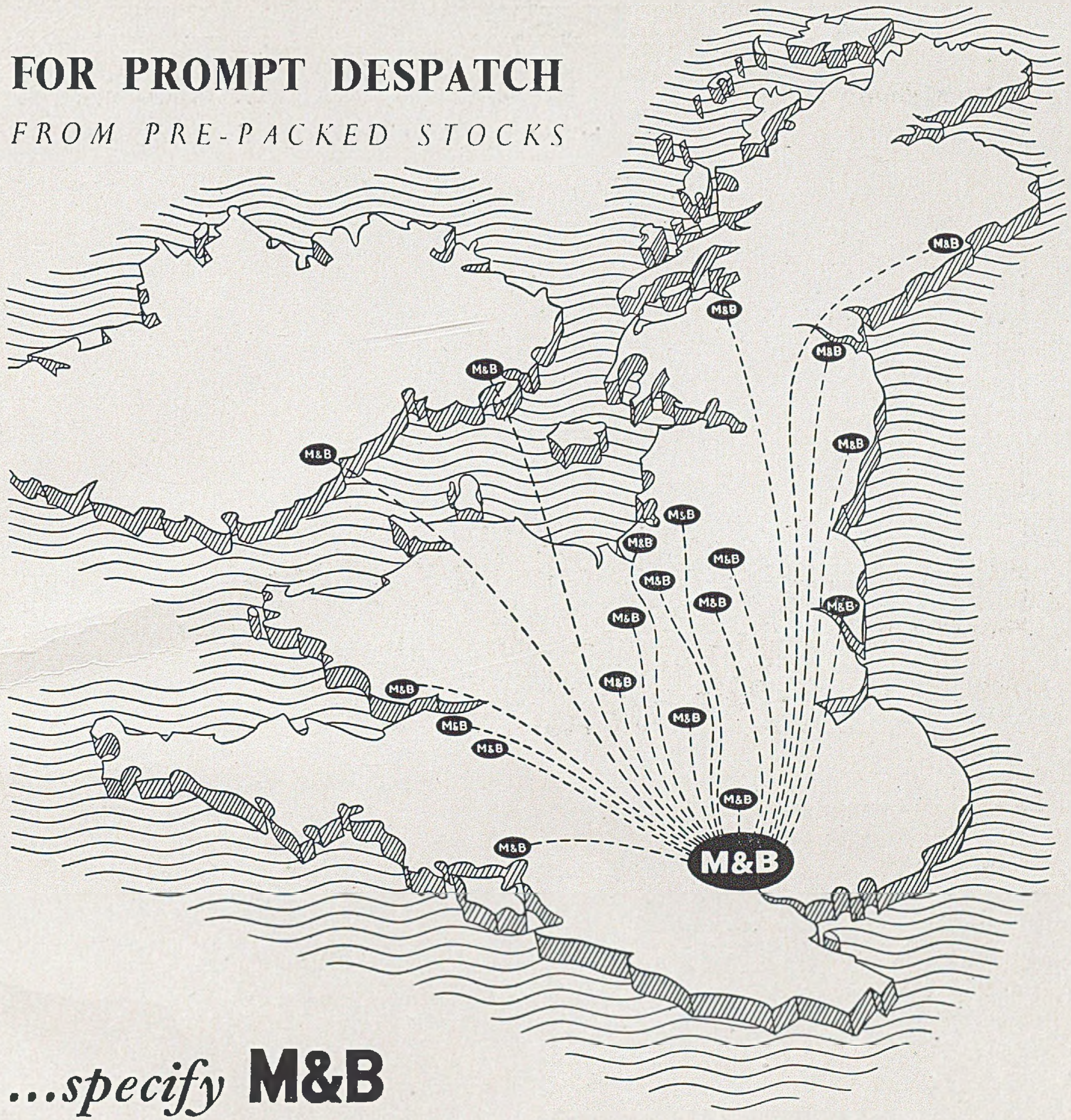
LONDON · BRISTOL · BIRMINGHAM · MANCHESTER
LEEDS · GLASGOW · NEWCASTLE · BELFAST

228F85

Index to Advertisers

FOR PROMPT DESPATCH

FROM PRE-PACKED STOCKS



...*specify* **M&B**

The containers of **M&B** Laboratory Chemicals and Reagents are standardized to those sizes in most regular demand, and pre-packed stocks of every item are held at Dagenham and by stockists throughout the country. In this way, we are able to ensure that orders are promptly dealt with.

The range comprises over 500 different specifications, each of which is carefully checked in manufacture and clearly printed on the appropriate label. Specially designed containers provide maximum protection in transit and storage, and plastic screw-caps are fitted for convenience in handling.

M&B brand

LABORATORY CHEMICALS & REAGENTS

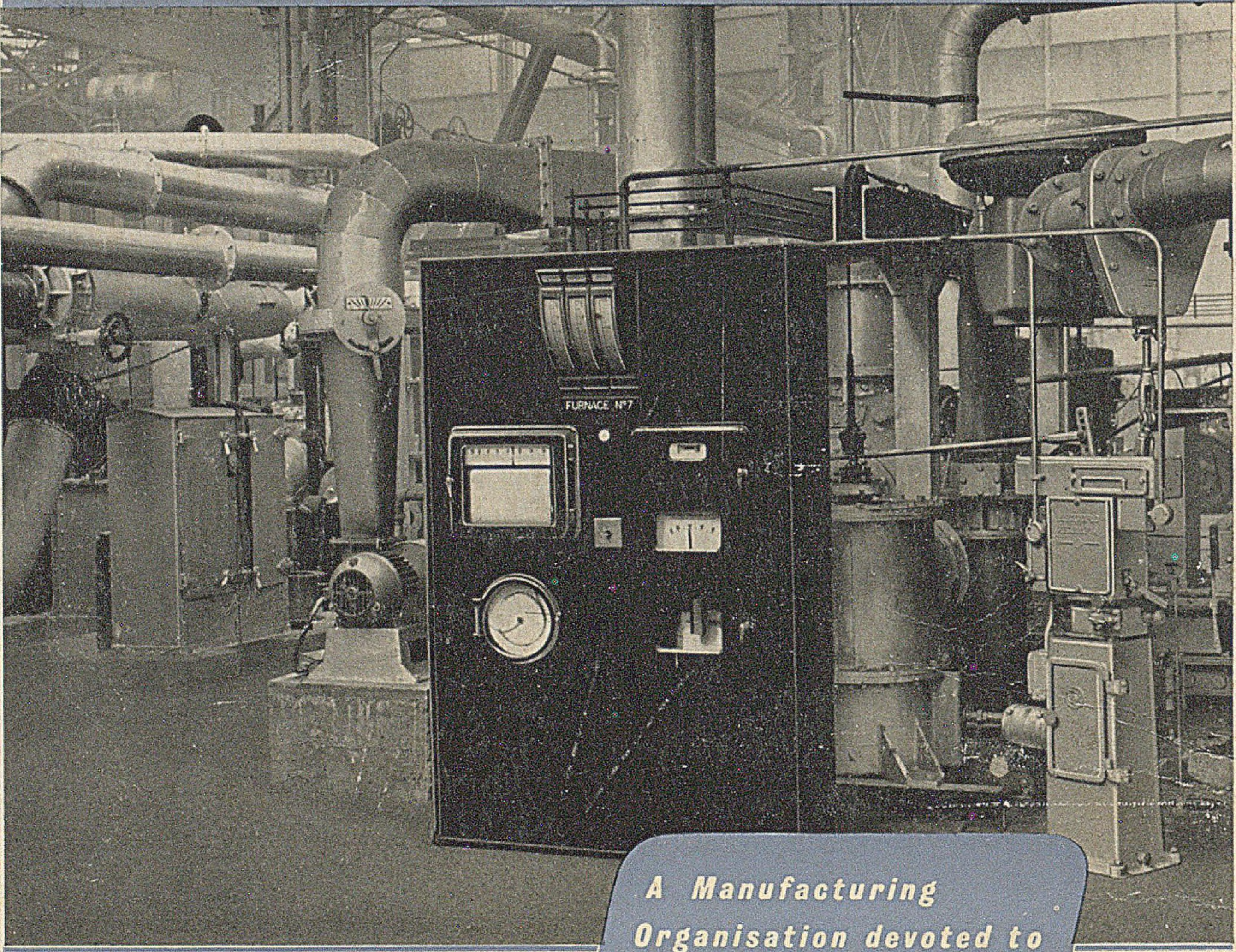
Manufactured by:

MAY & BAKER LTD
DAGENHAM · ENGLAND
Telephone: ILFord 3060 Extension 40

LA30

ASSOCIATED HOUSES: BOMBAY · LAGOS · MONTREAL · PORT ELIZABETH · SYDNEY · WELLINGTON. BRANCHES AND AGENTS THROUGHOUT THE WORLD

P



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.