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P. 99/LXXXI

THE JOURNAL OF THE INSTITUTE OF Metals

WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

211

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JUNE 1953

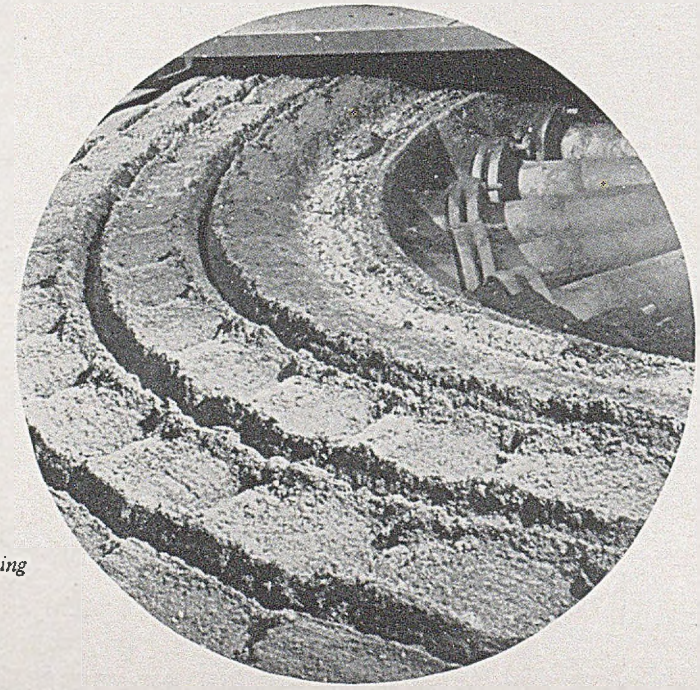
EUAINETOS *skilled in the designing of coinage*

On the earliest coins, the ruler's emblem was stamped as a guarantee of the purity of the metals. So today in the extraction of metals from complex and low grade ores and residues, the name Capper Pass is an assurance of the highest degree of purity and the most profitable rate of extraction.

You are invited to send samples or detailed analyses of your complex and low grade materials containing Tin or Lead, or tin combined with lead, copper, antimony, bismuth and silver.

This Greek artificer of the golden age of coinage worked in Syracuse at the turn of the fifth century before Christ. He created this ten-drachma silver piece to celebrate the defeat of the Athenian invasion by the Sicilians. It presents the head of Artemis Arethusa, patron goddess of the island, and on the obverse a charioteer receiving the laurels of victory.

CAPPER PASS *skilled in the extraction of metals*



Raw material after sintering

CAPPER PASS & SON LIMITED BEDMINSTER SMELTING WORKS BRISTOL

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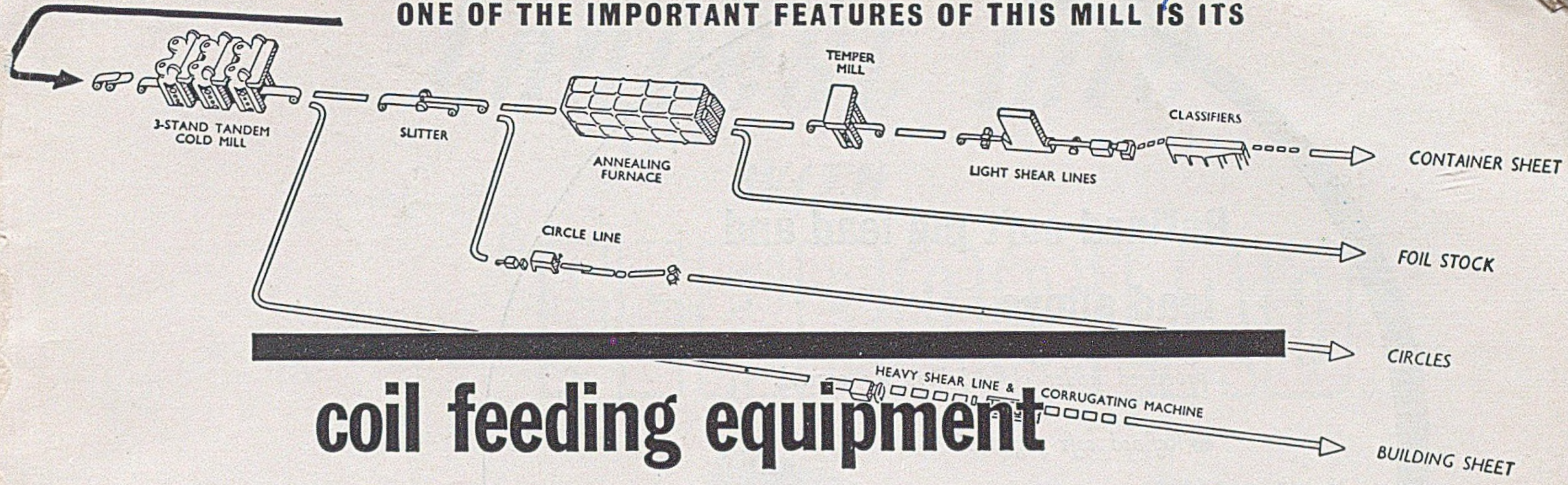
ONE OF THE IMPORTANT FEATURES OF THIS MILL IS ITS



coil feeding equipment

This 66 in. wide 3-Stand Tandem Mill installed in the Rogerstone Works of Northern Aluminium Company Limited is designed for a finishing speed of 2000 feet per minute. Production rate on this mill is also materially helped by the smooth and rapid operation of the special combined idle cone uncoiler and sticker roller bridle fitted to the entry side. This combined unit is designed to handle 5000 lb. coils of strip up to 56 in. wide. The opening and closing of the cones, the clamping action of the 7-roll bridle to the pre-determined degree of penetration and the traverse of the whole unit whereby the tail of the strip is stuck right into the bite of the first stand work rolls, are all performed hydraulically under interlock control. The motions of accurate coil entry are considerably quickened and simplified —and one operator only is required.

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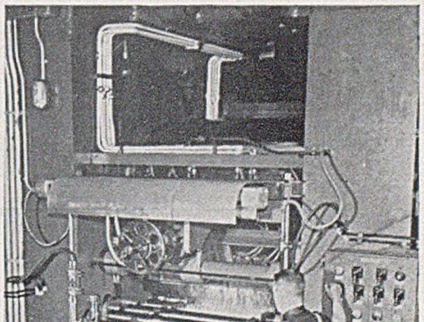
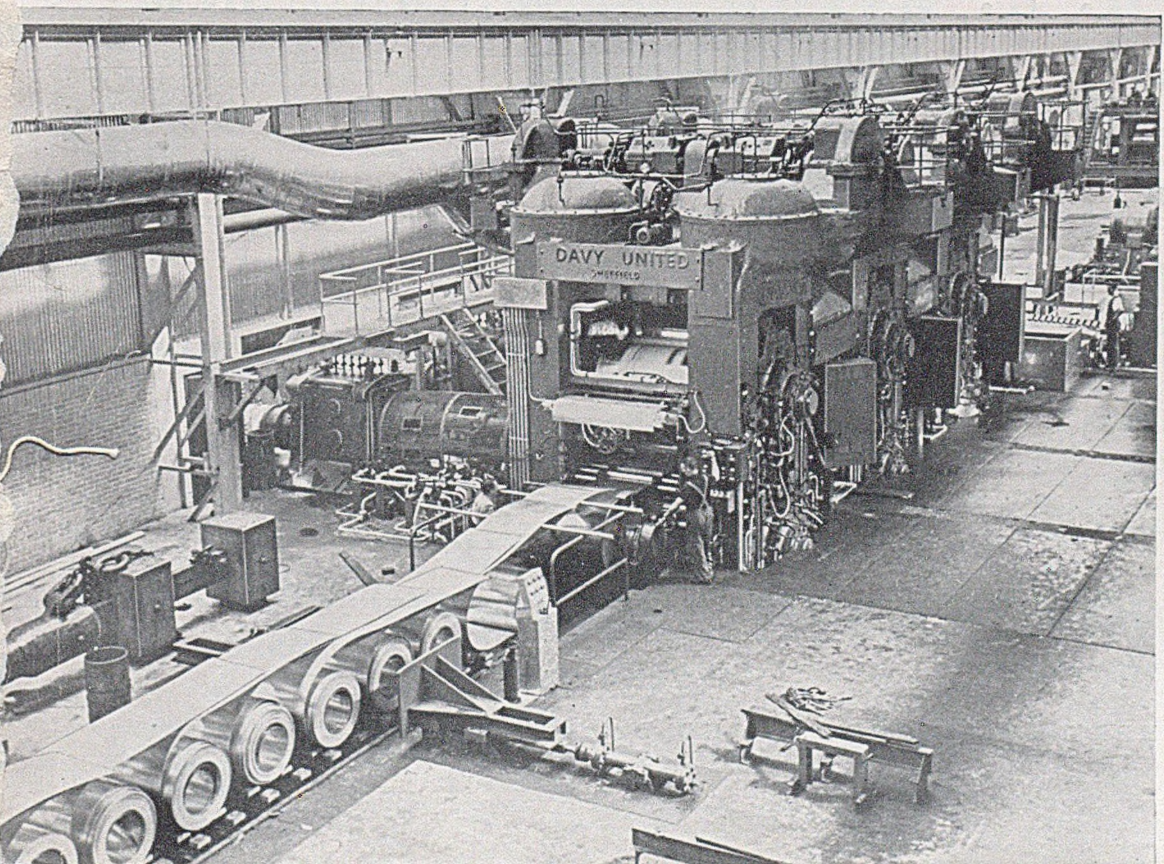


Close-up of the Combined Cone Uncoiler and Sticker Roller Bridle unit in the rolling position.



Coils queue up on the Entry Conveyor of the Tandem Mill, a time-saving advantage made possible by the smooth and rapid operation of the Combined Uncoiler and Roller Bridle unit.

Refined soft pig lead and lead alloys

processed under close laboratory control.

- *Refined soft pig lead*
- *Chemical lead*
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- *Antimonial lead*
- *Lead for equipment in radioactive work*
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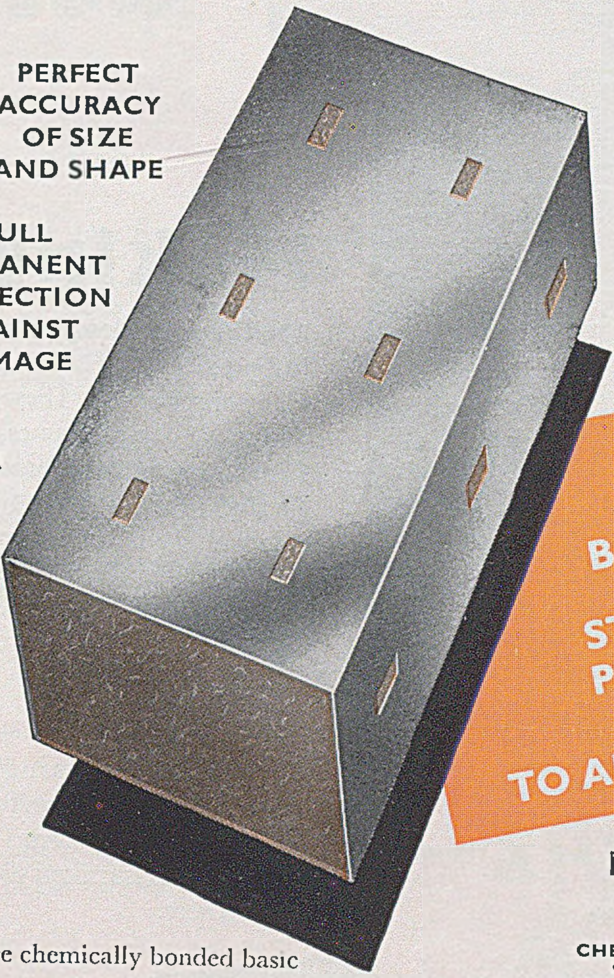

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G. R. "Ferroclad" are chemically bonded basic bricks and the only metal-cased bricks in which graded refractory material and metal casing form an integral unit by means of controlled hydraulic pressure. They are made in the normal standard sizes and are recommended for use in front walls, back walls, and ends of basic open hearth furnaces: walls and ends of copper reverberatory furnaces: in certain cases for electric furnace side walls, etc.

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'70'
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What OXYGEN Hand Cutting can do—

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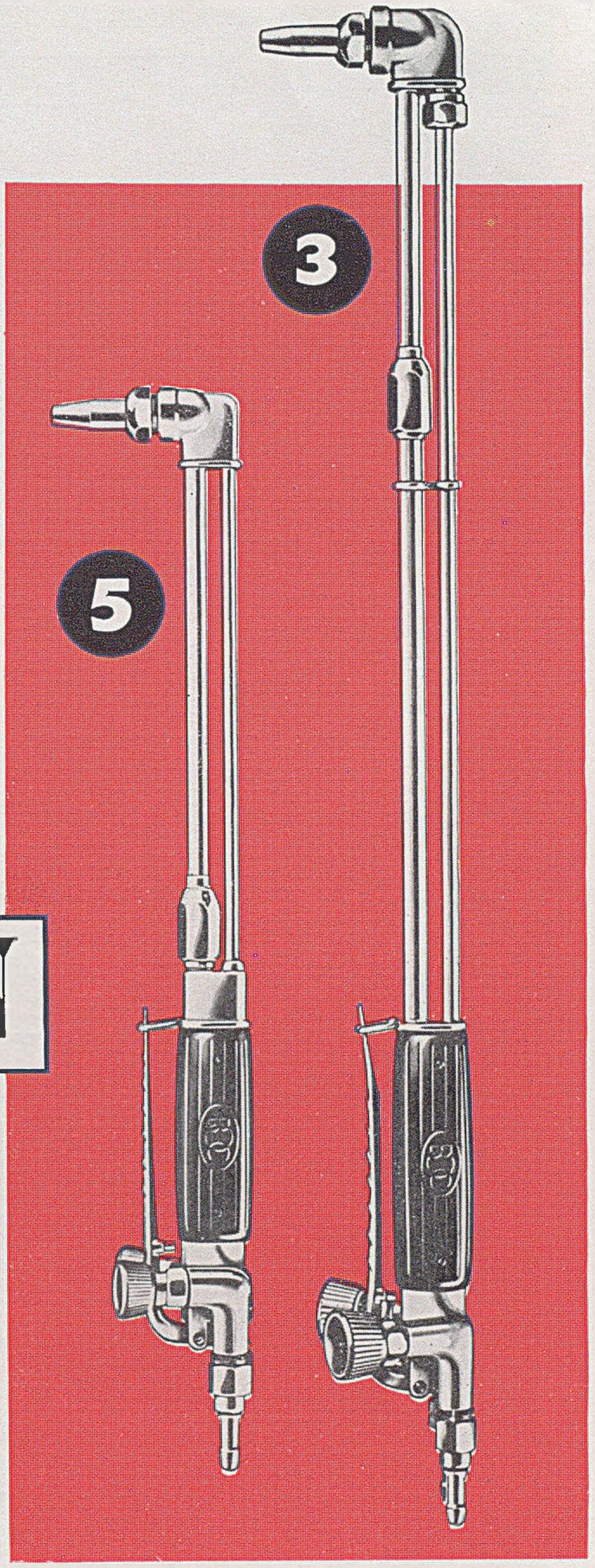
The Cutogen Blowpipe is now universally used for rapid cutting of iron and steel in a great many applications. These include shaping heavy plates for engine frames, bed plates, etc., profiling heavy sections for use in place of expensive forgings, for girder cutting, gouging and weld removal. It is an accepted tool for all constructional cutting. The clean, accurate cuts achieved make for easier fabrication.

Breaking...

The capacity and versatility of the Cutogen hand cutter makes it indispensable on all scrap cutting and breaking operations—such as the breaking-up of 10″ thick armour plate on H.M.S. Warspite shown here.

BOC DESIGNED AND MADE PERFECT BY

—and the B·O·C EQUIPMENT to do it!



Cutogen 5

1 A robust Cutter. Cuts 12 ins. mild steel and 4 in. cast iron.

2 Quick-action ball ended valves with large size control knobs.

3 Lever cutting control "off" when released.

4 Valve body and nozzle head from hot brass stampings. Nozzle head internally threaded.

5 Anti-spatter nozzles.

6 Positive colour identification and non-interchangeable threads for gas connections.

STANDARD MODELS

18″ with 90° HEAD 24″ with 75° HEAD (18″ model also available) Also available to order with longer shanks in increments of 6 ins. Chromium plated—not for eye appeal—but for sound service. The smooth, hard-plated surfaces are spatter free.

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With their accurate one-piece nozzle and finger-tip control of fuel gas and oxygen, Cutogen Blowpipes are the lightest, strongest and easiest-to-use of all oxygen cutters. Wherever there's iron or steel to be cut—remember CUTOGEN.

Cutogen 3

Similar specifications to Cutogen 5 but of larger capacity to handle the heaviest hand cutting work.

Standard length 27 ins. between centre lines of cutting nozzle and control valves. Obtainable in greater shank lengths by increments of 6 ins.

Top tube is stainless steel to give rigidity to the increased shank length necessary for comfortable operation.

Cuts 20 ins. mild steel
Cuts 15 ins. cast iron

Other Cutogen Blowpipes are available as follows:—
Cutogen 6 for Powder Cutting
Cutogen 7 for Powder Washing
Cutogen 8 for Deseaming

Write for full particulars to your B.O.C. Branch

THE BRITISH OXYGEN CO LTD LONDON & BRANCHES

Furnaces supplied and installed by The Salem Engineering Co. Ltd. Instrumentation by Honeywell-Brown Limited comprised: five Strip Chart ElectroniK Potentiometer Controllers, five Strip Chart ElectroniK Recorders, one Switch Cabinet and special thermocouples. Installation of instruments and switch cabinet jointly by The Salem Engineering Co. Ltd. and Honeywell-Brown Ltd.

BROWN INSTRUMENTATION
FOR TEMPERATURE MEASUREMENT AND CONTROL
THE STEEL COMPANY OF WALES LTD · TROSTRE WORKS

'United' Cleaning and 'Wean' Pickling Lines equipment supplied by the United Engineering & Foundry Company of the U.S.A. in collaboration with John Miles & Partners (London) Ltd. Instrumentation by Honeywell-Brown Limited as follows: two separate panels, one for the cleaning line and one for the pickling line, each similar and comprising six Circular Scale ElectroniK Controllers and one Strip Chart ElectroniK Recorder. Layout and installation of instruments and panels jointly by John Miles & Partners (London) Ltd. and Honeywell-Brown Ltd.

British Made *Patent Specification 436,098*

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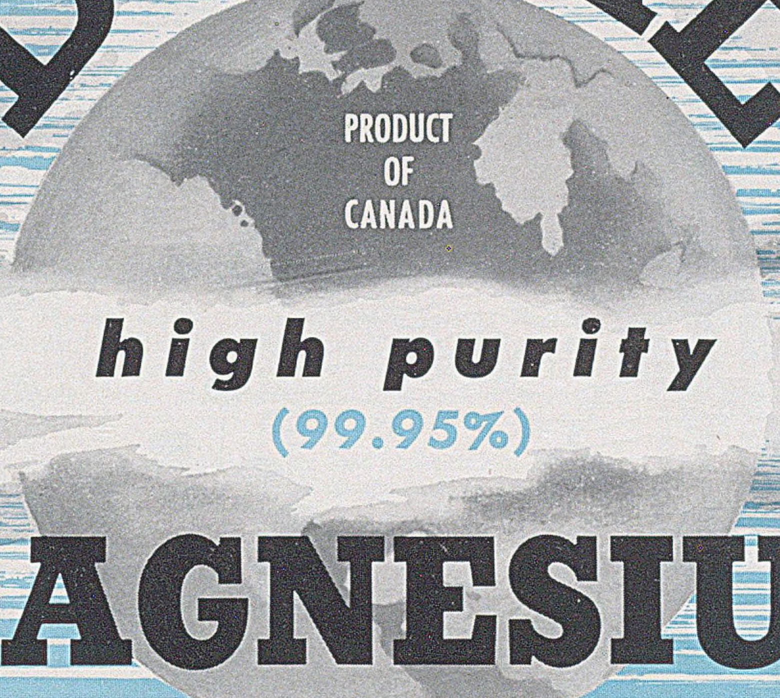

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OF DOMAL
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ELEMENT	PER CENT
Aluminium	0.004
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Copper . . . . less than	0.001
Iron	0.001
Silicon	0.006
Calcium	0.004
Lead	0.001
Zinc	0.001
Sodium . . . . less than	0.001
Total Impurities . . less than	0.022
Magnesium (by difference)	99.978

When you specify Domal Magnesium and magnesium alloys you can be sure of unvarying high purity—and of a dependable source of supply. Domal also produces high purity calcium and titanium.

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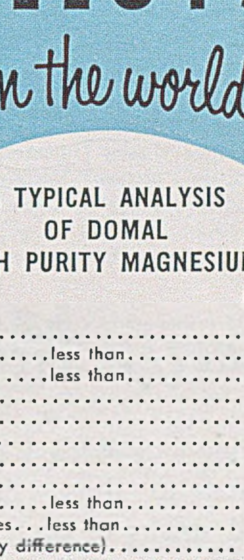

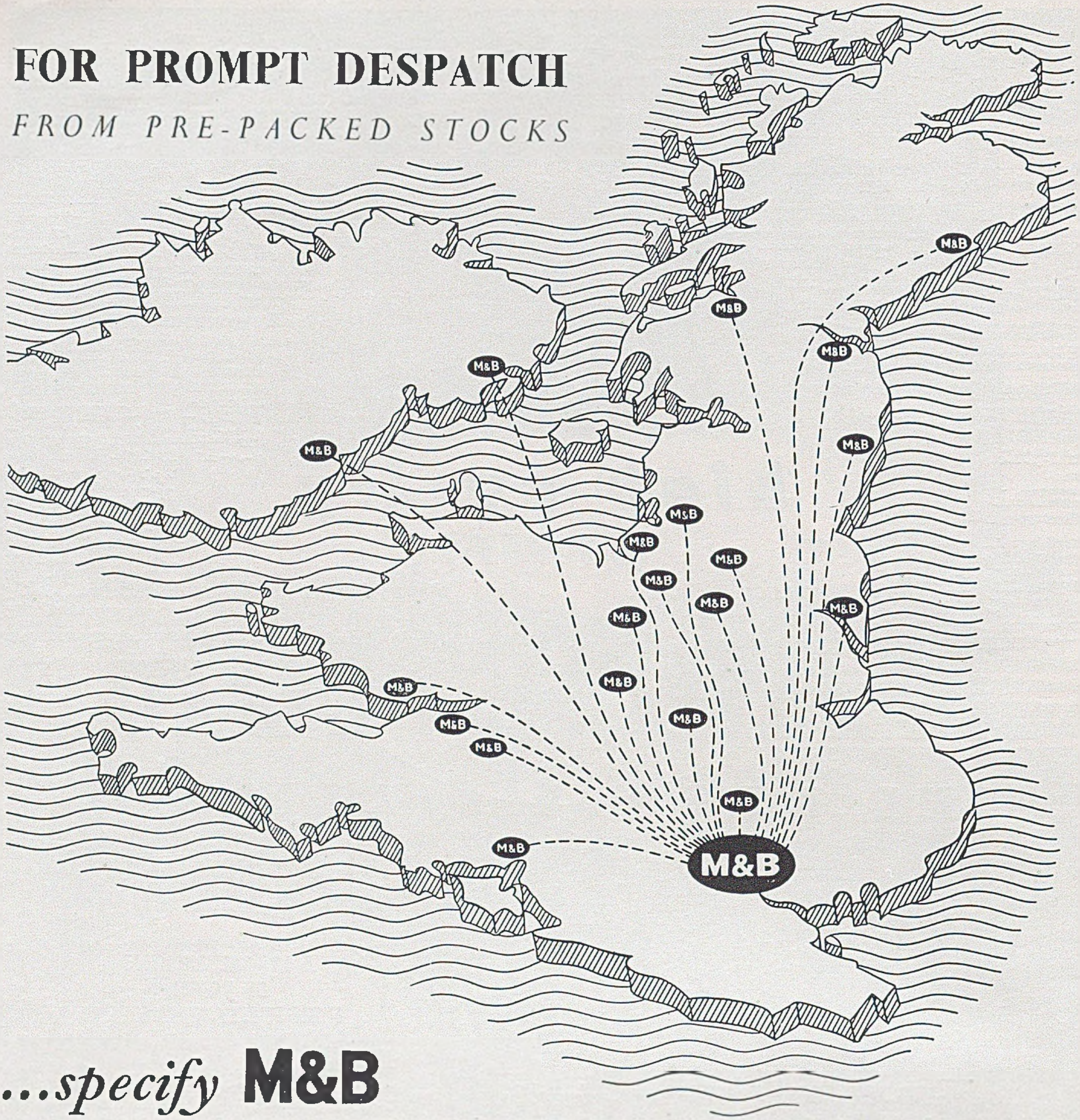

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FATIGUE TESTING EQUIPMENT

FOR MATERIALS

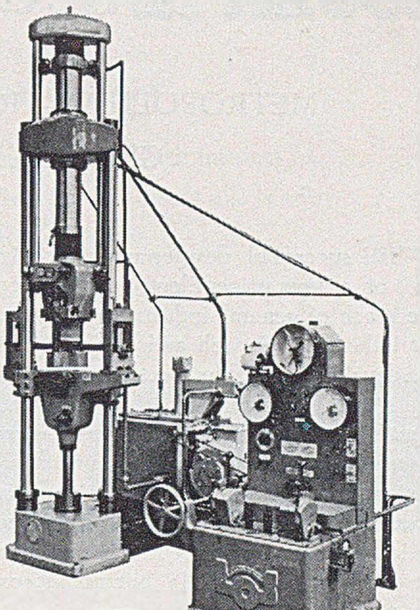
Losenhausenwerk Models UHW Universal Fatigue Testing Machines are suitable for tensile, compression and bending tests on material specimens under fluctuating loads plus/minus oscillation loads.
Manufactured in two sizes of 6 and 20 tons dynamic load capacity.

FOR COMPONENTS AND STRUCTURAL MEMBERS

Losenhausenwerk Models UHS Universal Fatigue Testing Machines are suitable for tensile, compression and bending tests on materials, components, structural members, and built-up machine assemblies under static loads, fluctuating loads, plus/minus oscillating and alternating dynamic loads.

Manufactured in sizes having maximum static load capacities of 20, 40, 60, 100 and 200 tons and maximum dynamic load capacities of 10, 20, 40, 60 and 100 tons respectively.

Losenhausenwerk Models UHP Universal Testing Machines of 6, 10, 20, 40, 60, 100 and 200 tons maximum static load capacity may be fitted with a Pulsator Unit either at the time of ordering or at a later date making them suitable for fluctuating load tests in either tension or compression up to a maximum dynamic load of approximately half the maximum static load capacity. They are not however suitable for alternating load tests from tension to compression.

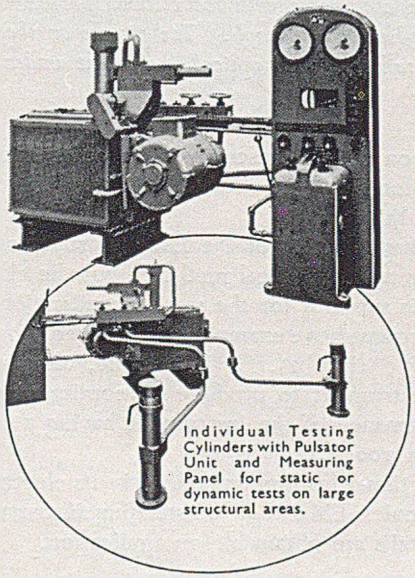


Type UHS
Universal
Fatigue Testing
Machine

FOR LARGE STRUCTURAL AREAS

When structural assemblies are of such dimensions that they cannot be accommodated in any of the machines previously described and where it is necessary to apply static or dynamic loads at points widely separated over a structural area, **Losenhausenwerk Individual Testing Cylinders** may be connected through a distributor to a Losenhausenwerk Pulsator Unit and Pump as shown in the illustration. Typical applications of this type of installation are for static and fatigue tests on aeroplane wings, tail planes, etc., and in building research.

These individual testing cylinders are manufactured for maximum static loads of 2, 6, 10, 20, 40, 60 or 100 tons with maximum dynamic load capacities equal to half the static load capacity. Losenhausenwerk Pulsator Units are manufactured in sizes of 50, 100, 200, 300, 800 cubic cm./volume. Models 50 to 300 with 330, 500, 600, 660 and 1000 strokes per minute and model 800 with 200, 300, 400 and 600 strokes per minute.



Individual Testing Cylinders with Pulsator Unit and Measuring Panel for static or dynamic tests on large structural areas.

The complete Losenhausenwerk range in addition to the equipment described includes Vibration Machines, Vibration Tables, Universal Testing Machines, Building Material Testing Machines, Creep Testing Machines, Chain Testing Machines, Spring Testing Machines, Pendulum Impact Testing Machines and Balancing Machines.

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Remarkable Advance in Electron Microscopy

METROPOLITAN-VICKERS DEVELOP LOW PRICE INSTRUMENT FOR GENERAL INDUSTRIAL RESEARCH AND TESTING

THE successful development by Metropolitan-Vickers of a low price Electron Microscope brings within the reach of general industry a remarkable scientific tool for industrial research and testing. Metrovick pioneered the development of the electron microscope in this country, their first instrument being completed 1935/36 and remaining the sole British electron microscope until 1940. Metrovick post-war research has resulted in two successive designs, EM2 and EM3, each being a remarkable technical advance upon its predecessor and now comes the EM4 which has a performance in most respects comparable with its predecessor the EM3 but at approximately one-half the price.

In general terms, the electron microscope is capable of a performance one-hundred times better than that of the best optical microscope. In practical service in the laboratory it may be assumed that the EM4 enables detail twenty times finer than is observable by normal optical methods to be viewed and recorded. In the Metrovick microscope the final image is formed on a fluorescent screen viewed through a single port of wide aperture. The microscope design includes a camera fitting and no adjustment need be made to the normal viewing image.

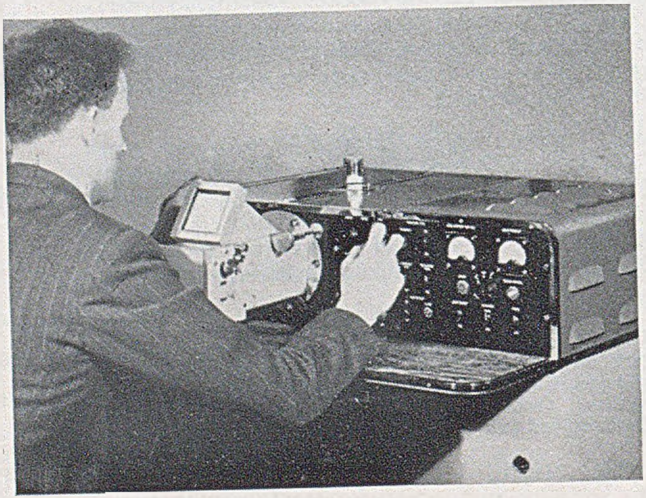


Fig. 1. Close-up view of the control desk and camera of the new Metrovick electron microscope Type EM4.

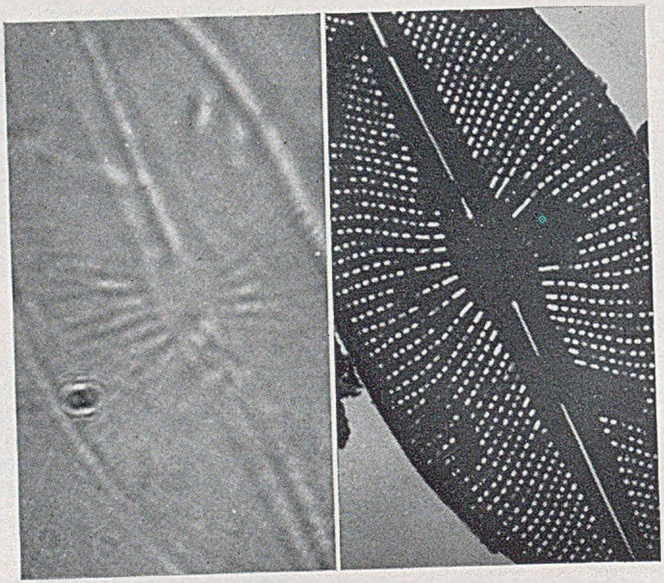


Fig. 2. In these microphotographs of diatoms (x 2375) the clarity of definition obtainable with the electron microscope is shown by the right-hand photograph. The picture on the left shows definition obtainable with optical instrument under the same conditions.

Sharp definition on the fluorescent screen automatically means that sharp focus will be obtained on the photograph without further adjustment.

Not only is the magnification of the EM4 far superior to that obtainable by normal optical methods, but the clarity of the definition is superior for the same magnification, as is seen in the two comparative examples of microphotographs of diatoms x 2375 (Fig. 2). It is possible, also, with the Metrovick instrument to produce stereomicrographs, the overall performance of the instrument having a resolution better than 100 A.U.

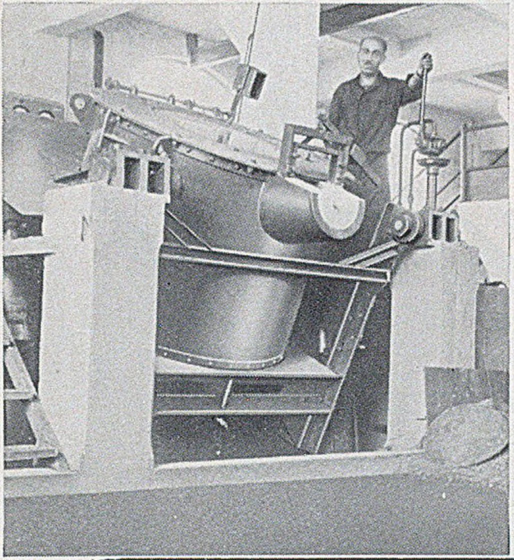
In all its constructional features as well as in its electronic design, the Metrovick EM4 is an outstanding instrument. Full technical details are obtainable on application.

METROPOLITAN-VICKERS ELECTRICAL COMPANY LTD.
TRAFFORD PARK, MANCHESTER, 17. *Member of the A.E.I. group of companies*

METROVICK *Applying Electronics to Industry*

C/E 201

Latest designs of mains frequency core type



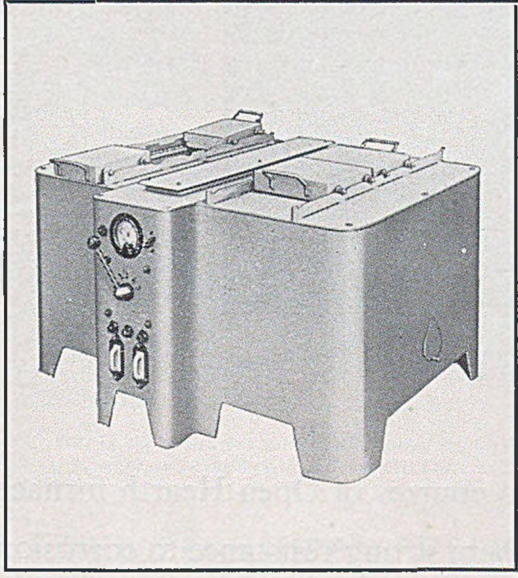
Induction Melting Furnaces

now available in Great Britain

for aluminium and other light metal alloys with outputs of 100 to 1,750 lbs. per hour and for copper and its alloys with outputs of 50 to 4,000 lbs. per hour.

Among the many advantages of the

GWB-A.TAGLIAFERRI furnaces

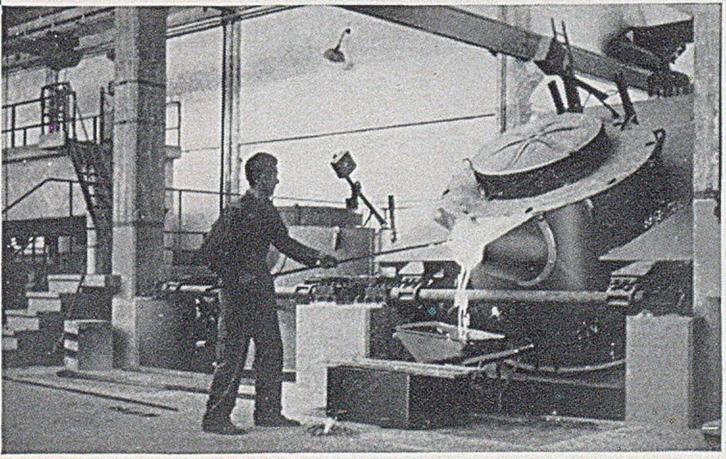


are the following:-

1. Initial starting without using hot metal.
2. Positive metal circulation without turbulence gives maximum refractory life.
3. Special channel design on furnaces for light metals facilitates cleaning without the need for completely emptying the furnace.
4. The refractory lined casing is easily exchanged with the relined spare without removing electrical connections or inductors.
5. Clear indication is given when end of lining life is approaching.
6. Refractory lining of melting duct can be repaired without dismantling the furnace.

Full details of standard units suitable for all normal foundry requirements may be had on request.

TOP: This unit has hydraulic tilting and is rated at 210 kW.
ABOVE: An aluminium bale-out type furnace with two crucibles having control panel incorporated.
RIGHT: A 300 kVA unit, to melt brass at 2,200 lbs. per hour, bronze at 1,760 lbs. per hour.



Proprietors: Gibbons Bros. Ltd.,
& Wild-Barfield Electric Furnaces Limited.

G.W.B. ELECTRIC FURNACES LTD.
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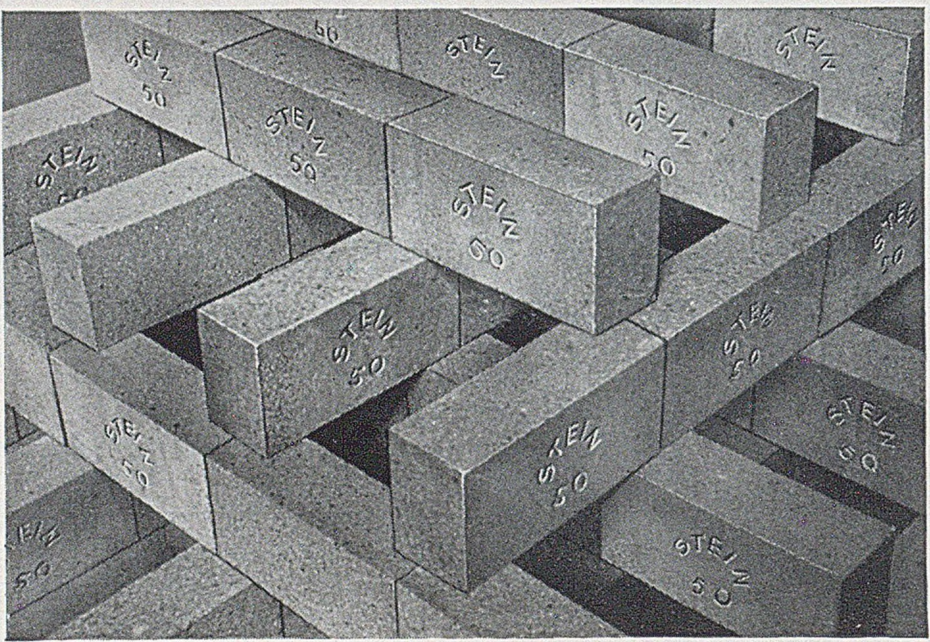
STEIN *Refractories*



STEIN 50

APPROXIMATE TECHNICAL DATA	
Alumina	51-52%
Refractoriness	Cone 35/36 1780°C
Refractoriness-under-load 28 lbs./in²	5% Deformation at 1595°C
After-Contraction 2 hrs. at 1500°C	Nil

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

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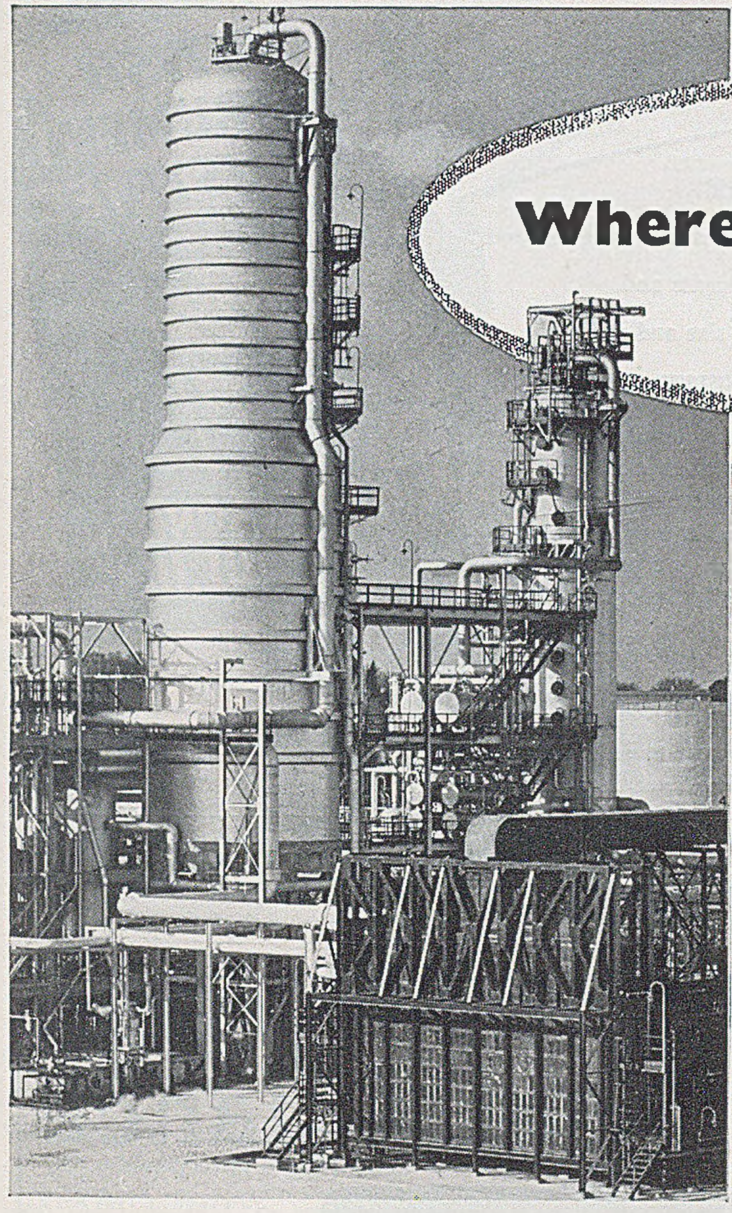
36/CR/18

Where Chlorides attack MONEL Defends



One of the distillation units at the Fawley Refinery.
(By courtesy of the Esso Petroleum Co. Ltd.)

Standard petroleum refinery practice is to use Monel in those sections of the distillation towers where chloride attack has to be withstood. The inset shows a Monel section in course of construction at the Fawley Refinery, where production will eventually reach 6,000,000 tons of petroleum products annually.

In this great refinery, Monel is used for appropriate sections of the linings of the fractionating towers and for ancillary piping, bubble caps and trays in both atmospheric and vacuum distillation units. Monel has high mechanical strength combined with outstanding resistance to the attack of a wide range of corrosive media.

OUR TECHNICAL PUBLICATIONS

Wiggin Nickel Alloys — No. 19 contains a full description of the Fawley refinery with other articles on typical uses of the high-nickel alloys in industry.

MONEL* and Some Other High-Nickel Alloys *versus* SULPHURIC ACID — a 72-page technical publication, containing a wealth of data based on research and plant corrosion tests.

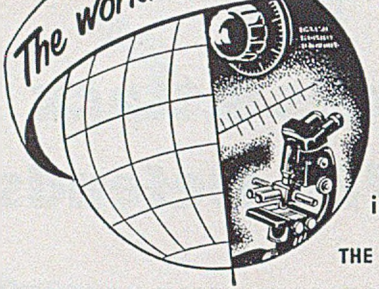
* *Monel is a registered trade mark.*

We shall be glad to send copies of either or both of these publications, free of charge.

Fitting Monel tubing and bubble cap trays.
(By courtesy of Messrs. Foster Wheeler Ltd.)

HENRY WIGGIN & COMPANY LIMITED · WIGGIN STREET · BIRMINGHAM 16

The world's most comprehensive display of Instruments



Nowhere in the world will you find an exhibition, specialising in instruments and instrumentation, of the size and scope of THE BRITISH INSTRUMENT INDUSTRIES EXHIBITION

Every field of instrumentation in Industry, Medicine, Research and Education is covered. You will be able to see all classes of temperature, pressure and flow controllers for industrial use, complete equipment for sea and air navigation and land surveying, British equipment for the research laboratory in pure and applied science, and precision equipment in such diverse fields as metrology and photography.

BRITISH INSTRUMENT INDUSTRIES EXHIBITION

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JUN. 30 - JUL. 11 '53
OPEN DAILY (EXCEPT SUNDAY) 10 A.M.-6.30 P.M.

IT'S CORONATION YEAR!

Accommodation should be booked as early as possible. If you require any assistance, or an official catalogue of the exhibition if you are unable to come, write to the Organizers:—

The full scope of the products of the British Instrument Industry will be revealed at this exhibition. Present users of instruments and industrialists, whose problems might be solved by appropriate instrumental control, will find a visit to this exhibition extremely rewarding.

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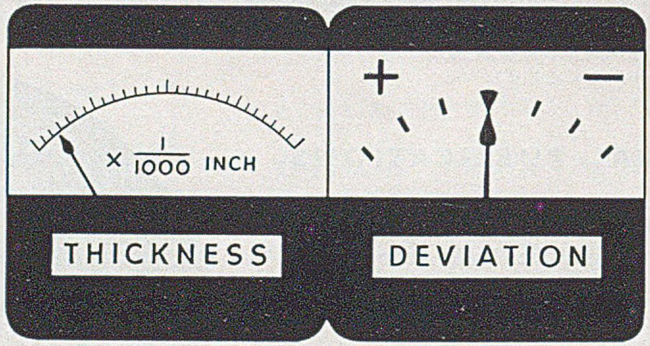
Thickness without contact — at a glance

ALUMINIUM
BRASS
STEEL



COPPER
and other
ALLOYS

with the **ATOMAT** beta-ray thickness gauge

The **BALDWIN ATOMAT THICKNESS GAUGE** gives an accurate direct reading (and recording if required) of both actual thickness and deviation from the nominal.

The continuous control of thickness ensures production of uniform material.

For further details see Leaflet V124.

See Stand No. 83 at the British Instrument Industries Exhibition (June 30th—July 11th).

BALDWIN
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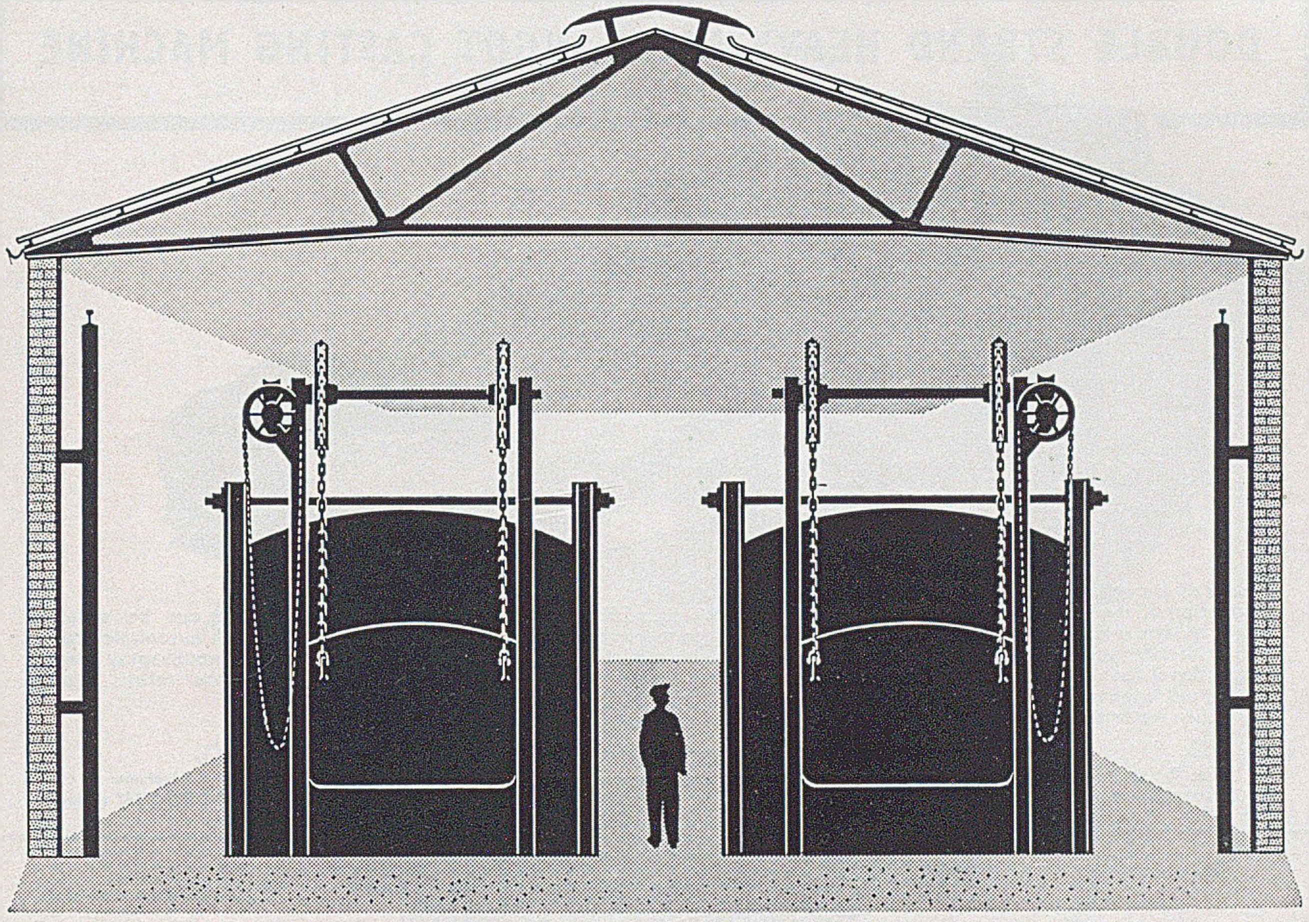

How many furnaces can you count — two or three?

"*You needed three to keep pace with the new production line. . . . Where's the third?*"

"In the other two. We've increased their capacity by fifty per cent. We get six charges per shift instead of four."

"*They look about the same.*"

"As a matter of fact, externally, they are exactly the same. A bit larger inside—thinner walls. But that's not the point; we've cut down the heat the walls soak up on every charge, it's that which has made the difference. They're built of a light-weight brick with a very low thermal capacity—and it's a good insulator into the bargain. We use no more fuel for six charges than we used for four."

"*The old hot face insulation, eh? Shouldn't have thought it would stand the temperature.*"

"The old stuff wouldn't. This is something new: stands 1540°C or 2800°F if you prefer it that way hence its name M.I.28. M. for Morgans of course."

MORGAN Refractories

ARE WORTH FAR MORE THAN THEY COST

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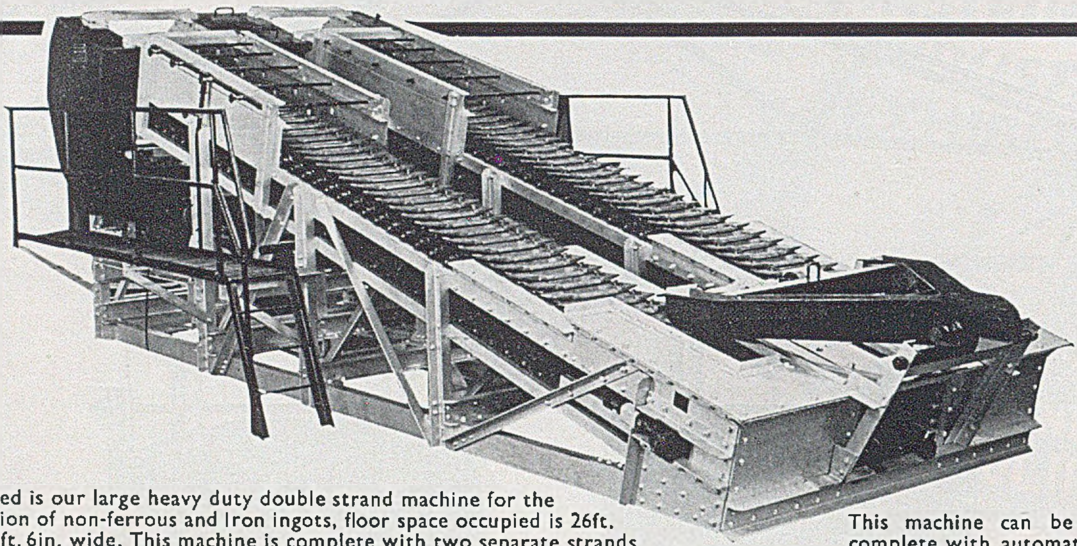

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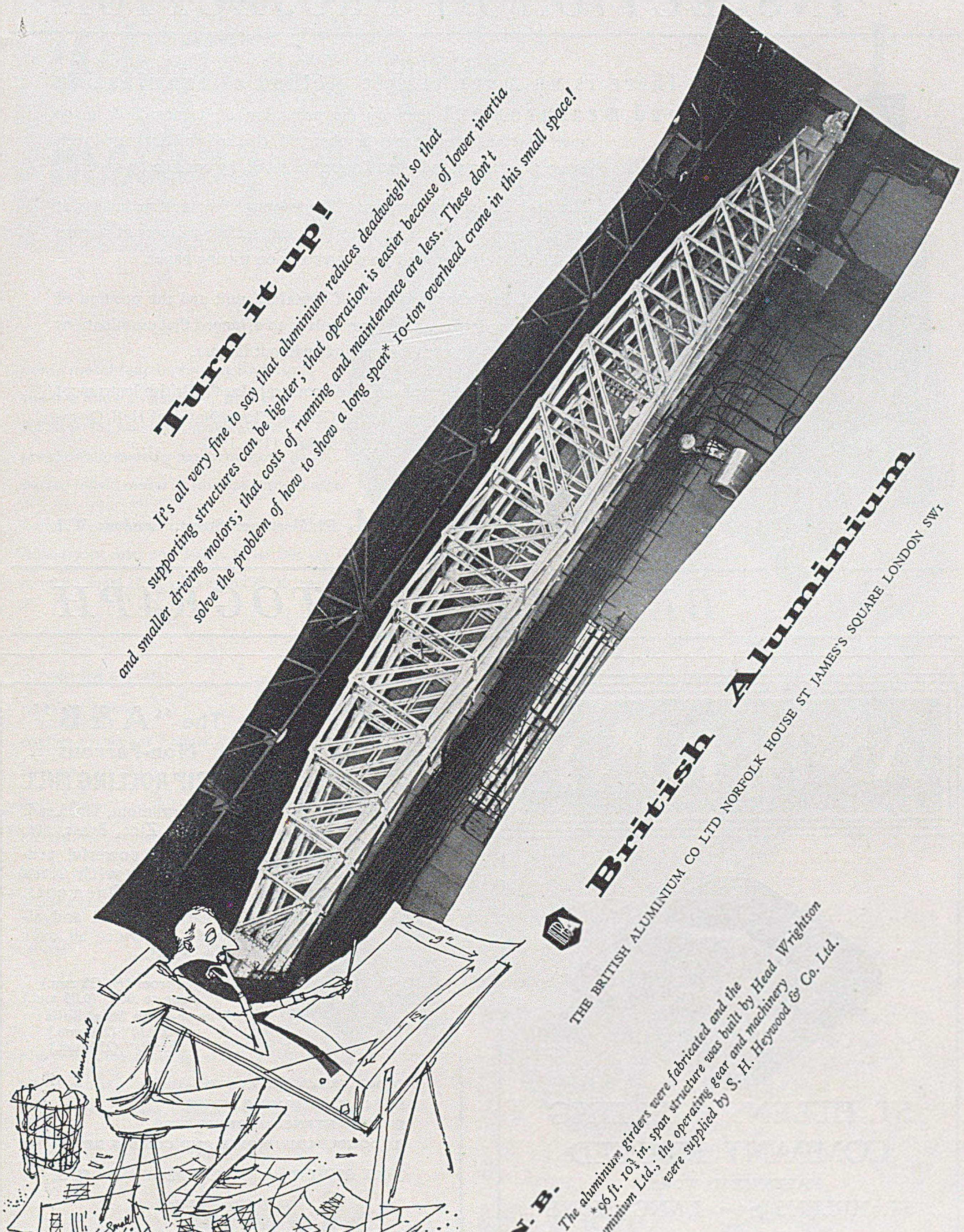

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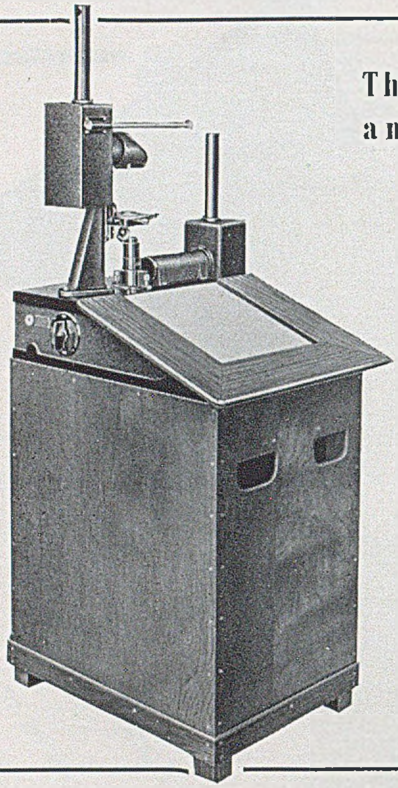

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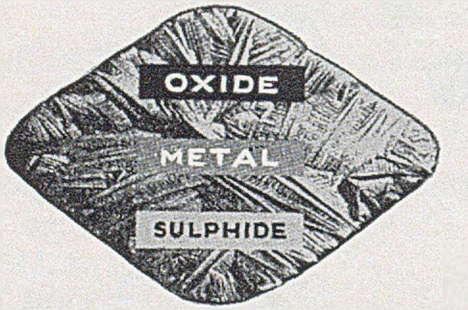

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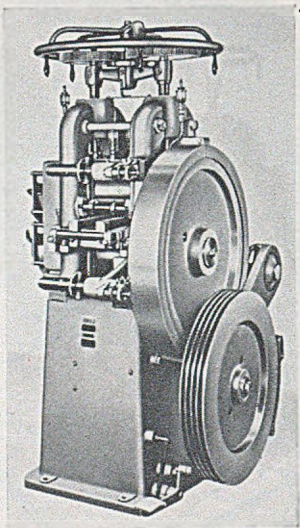

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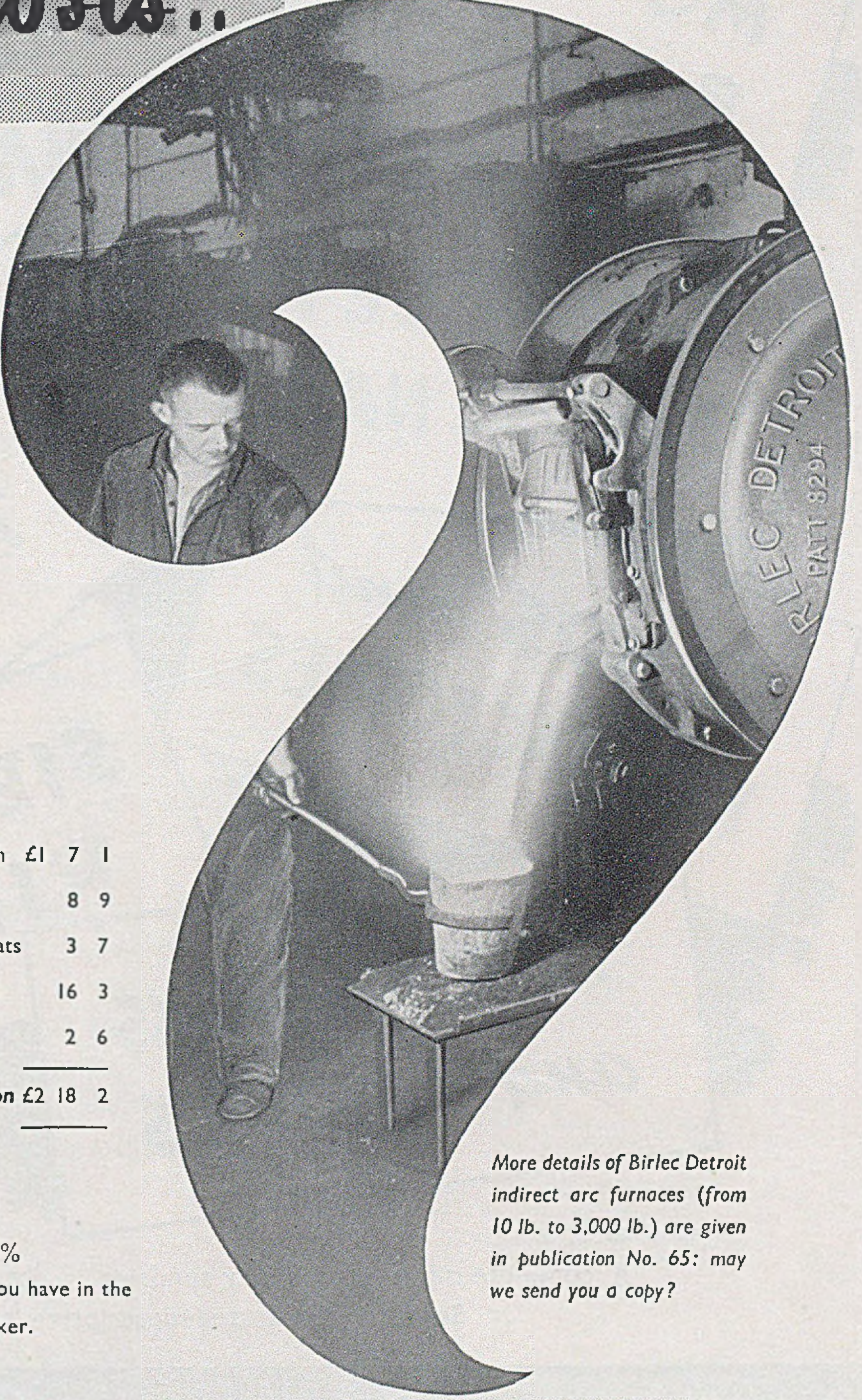

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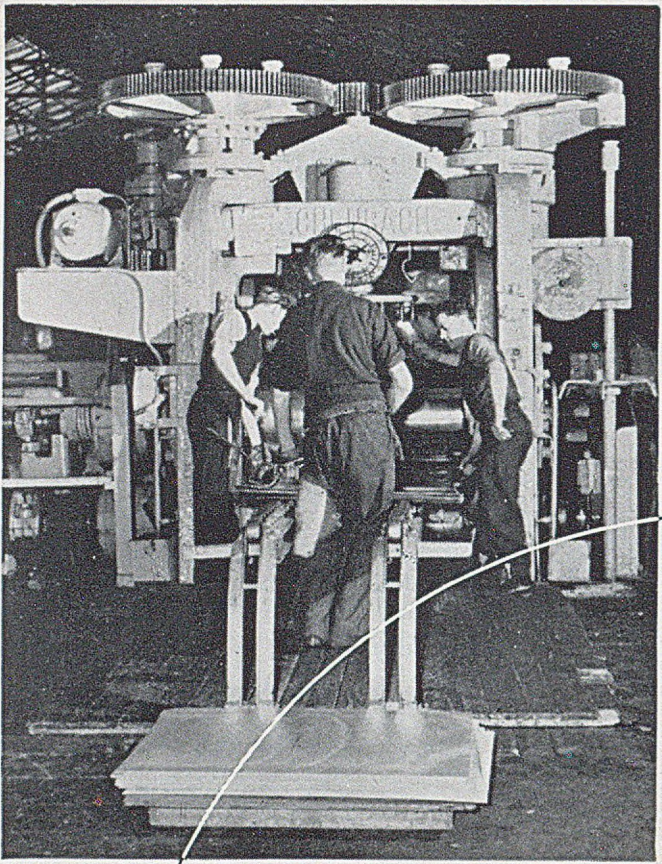

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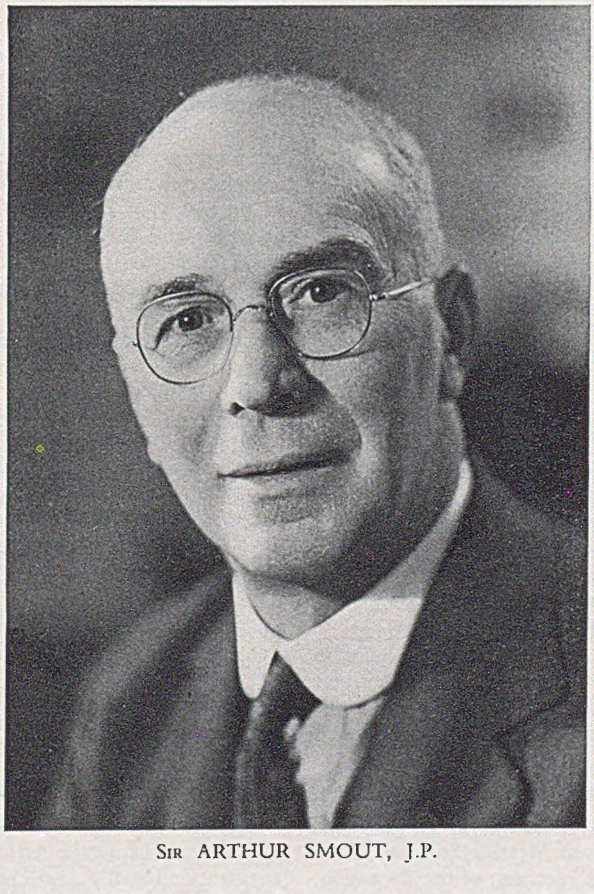
OF THE INSTITUTE OF METALS

VOLUME 1 JUNE 1953 PART 22

INSTITUTE NEWS

Election of Fellow

The Council has elected Sir Arthur Smout (Past-President) a Fellow of the Institute in recognition of his long and distinguished services. The number of Fellows is limited to twelve, and Sir Arthur's election brings the present total to nine.



Sir ARTHUR SMOUT, J.P.

Arthur John Griffiths Smout was born at Birmingham on 18 November 1888, and was educated at King Edward's School, Birmingham, and the College of Technology, Birmingham. In 1905 he joined the Elliott Group of metal companies as a student apprentice.

From 1920 to 1924 he was Works Manager of Elliott's Metal Company and from 1924 to 1934 Production Director to the Elliott Group, which was merged into Imperial Chemical Industries, Ltd., in 1927. He became Managing Director in 1934 and was Chairman from 1934 to 1942 of what is now known as Imperial Chemical Industries, Ltd., Metals Division; he was appointed to the Main Board of Imperial Chemical Industries, Ltd., in 1944, with responsibility for the Company's metal, ammunition, and explosives interests. He retired from the Board earlier this year. He is Chairman of Murex, Ltd., and Murex Welding Processes, Ltd., a Director of Pyrotenax, Ltd., and a part-time member of the West Midlands Gas Board. He was Director-General of Ammunition Production at the Ministry of Supply from 1942 to 1945, and was knighted for his services in July 1946.

Sir Arthur has devoted much time to public service and to the advancement of science and industry. He is a Justice of the Peace for the City of Birmingham and Member of the Magistrates' Courts Committee; a Life Member of the Court of Governors and Member of Council of the University of Birmingham; Member of the West Midlands Advisory Council for Further Education; Past-President of the Birmingham Chamber of Commerce; Chairman of the B.I.F. Management Committee; Vice-President of the Institution of Mining and Metallurgy; and a Member of Council of the British Non-Ferrous Metals Research Association and of the Copper Development Association.

He is a Fellow of the Royal Institute of Chemistry, a Fellow of the Institution of Metallurgists, and an Associate of the College of Technology, Birmingham. He holds the Insignia Award of the City and Guilds of London Institute, which was conferred upon him by H.R.H. The Duke of Edinburgh, in January, 1953.

Sir Arthur was elected a member of the Institute of Metals in 1917, since when he has given long and devoted service to the Institute. He was a Member of Council from 1924 to 1936 and from 1937 to 1940; a Vice-President from 1940 to 1943; a Member of Council from 1945 to 1946; a Vice-President from 1946 to 1948; President 1948–1950; and Past-President 1950–1953. During the course of these years he has served on many Committees of the Institute, notably in the important role of Chairman of the Finance and General Purposes Committee from 1946 until he assumed the Presidency.

Autumn Meeting, Southport, Monday to Friday, 21–25 September

Since going to Press with the May issue of the "Bulletin", Professor E. C. Rollason (University of Liverpool) has accepted an invitation to join the Reception Committee in connection with this meeting.

A change has also been made to the Provisional Programme. The all-day tour on Friday, 25 September will now be to North Wales and not to the Lake District.

It is hoped to distribute a full programme of the meeting late in June or early July.

Faraday Society Jubilee Celebrations

Dr. L. B. Pfeil (Member of Council) represented the Institute at the Jubilee Celebrations of the Faraday Society in London on 16 April 1953.

Engineering, Marine, and Welding Exhibition and the Chemical Plant Exhibition, Olympia, London, 3–17 September 1953

The Council has accepted an invitation from the Organizers, for members of the Institute of Metals to pay an official visit to this Exhibition on Thursday, 10 September, next.

Tickets of invitation will be distributed to members in due course.

Binding Cases

The binding cases for the *Journal*, Vol. 80, and *Metallurgical Abstracts*, Vol. 19, 1951–52, are now available and are being sent to all those members who in the past have intimated their desire to receive the cases regularly. Any other members or non-member subscribers who wish to have the cases, which are available free, should apply to the Secretary without delay.

Members are reminded that, when having their issues for 1951–52 bound, they should retain the *Bulletin* sections, as the first volume of the *Bulletin* will not be completed until August 1953. A binding case for this volume will be made available subsequently.

Decennial Index to Metallurgical Abstracts

The Name Index volume of the decennial index to *Metallurgical Abstracts*, Vols. 1–10 (1934–43) has recently been published, and the Subject Index volume is in the press. The two volumes, which are not sold separately, may be obtained by members at a price of £4 4*s*., post free, and by non-members at £5 5*s*., post free. An order form was sent to all members with the April issue of the *Journal*.

The decennial index will prove invaluable to all those who have complete sets of *Metallurgical Abstracts*. Any members with gaps in their sets are invited to communicate with the Secretary, who may be able to supply the missing volumes.

"Properties of Metallic Surfaces"

The monograph containing the thirteen papers presented at the Symposium on "Properties of Metallic Surfaces" held last November, together with a full account of the discussion, is now in the press, and is due for publication in June.

The volume, which is No. 13 in the Institute's Monograph and Report Series, contains more than 350 pages and 150 illustrations. The published price is 35*s*. ($5.50), post free, but each member can obtain one copy for 17*s*. 6*d*. ($3.00), post free. Orders should be sent to the Secretary.

Summer School in the Use of Electrons in the Examination of Metals

Arranged by The Institute of Metals, in Co-operation with the Board of Extra-Mural Studies, University of Cambridge

A Summer School in "The Use of Electrons in the Examination of Metals" will be held in Cambridge this year. By courtesy of Professor Sir Lawrence Bragg, O.B.E., F.R.S., Professor G. W. Austin, O.B.E., and Dr. F. P. Bowden, F.R.S., the course will be held in the Cavendish and Goldsmiths' Laboratories and in the Laboratory for the Physics and Chemistry of Surfaces. The administration will be under the control of the Board of Extra-Mural Studies of the University of Cambridge. The teaching will be carried out by the following lecturers: Dr. A. F. Brown (Edinburgh University), Dr. V. E. Cosslett (Cavendish Laboratory, Cambridge), Mr. M. E. Haine (Associated Electrical Industries Research Laboratory, Aldermaston), Dr. J. W. Menter (Department of Physical Chemistry, Cambridge), and Dr. J. Nutting (Department of Metallurgy, Cambridge).

Purpose and Scope

In recent years summer schools in electron microscopy have been held in Cambridge, which have provided training in the theory of the electron microscope and its practical application to biological, physical, and metallurgical problems. The subject has now become so diverse that, in co-operation with the Institute, it has been decided to arrange a more specialized course on "The Use of Electrons in the Examination of Metals". This will consist of lectures, demonstrations, and practical classes in the principles of the electron microscope and electron-diffraction camera and their practical applications to the study of metals. Opportunity will be provided for operating and discussing the maintenance of the Metropolitan-Vickers, Siemens, and R.C.A. electron microscopes and the Edwards electron-diffraction camera.

Practice will be provided in the standard methods of specimen and replica preparation, and the new technique of reflection electron microscopy for the direct examination of metals will be discussed and demonstrated. A detailed syllabus can be obtained from the Secretary of the Institute of Metals. The opening lecture will be given by Professor A. G. Quarrell (Sheffield University) at 9.30 a.m. on Monday, 20 July. The school will end at 5 p.m. on Friday, 31 July.

Admission

It is necessary to limit the size of the Summer School, and it may not be possible to enrol all those who wish to attend. In order to assist the organizers, applicants are asked to complete a form of application which should be returned by 13 June. Applicants will be notified as soon as possible, and in any case not later than 26 June, whether it has been possible to accept their application or not.

Fees

The fee for attendance at the Summer School will be £20. The fee should not be sent with the original application. A registration fee of £2 will be due from those selected to attend as soon as they are informed to that effect, and the balance will be payable before the Summer School begins. If anyone has to withdraw after having been selected to attend, the return of the registration fee cannot be guaranteed, though a refund would be considered if the vacancy could be filled by another person.

Accommodation

A certain number of places for men have been reserved in Pembroke College, which is near to the Laboratories. The charge will be approximately 25*s.* per day. It is expected that some may be able to make their own arrangements for accommodation, and in other cases advice can be given about hotels, boarding-houses, &c. Unfortunately it is impossible to make any definite reservations in hotels at this stage on behalf of prospective members of the Summer School.

Details of the course and forms of application may be obtained from the Secretary of the Institute of Metals, and those interested should apply without delay.

"Acta Metallurgica": Privileged Rate of Subscription to Members of the Institute

As previously announced, each member of the Institute of Metals may subscribe to *Acta Metallurgica* at the privileged rate of £3 6*s.* per annum. Applications by members to subscribe at this rate must be sent to the Secretary of the Institute, accompanied by the appropriate remittance.

It is desired to make it clear that *all* members of the Institute, whether resident in the United Kingdom or not, are entitled to submit their subscriptions at this rate, through the Institute, if it is convenient to them to pay in sterling rather than in U.S. dollars.

Companies, libraries, &c., must subscribe, at the full rate of $12.00 direct to the publishers; the Institute cannot receive and transmit such subscriptions.

Election of Members

The following 14 Ordinary Members and 5 Student Members were elected on 14 April 1953:

As Ordinary Members

ALLEN, John, Metallurgist, Metals Division (IA), Royal Arsenal, Woolwich, London, S.E.18.

ALLEN, Thomas, General Manager, G.W.B. Electric Furnaces, Ltd., Dibdale Works, Dudley, Worcs.

ATKINSON, John Barnes, B.Sc., A.R.I.C., F.I.M., Senior Lecturer, Department of Metallurgy, College of Technology, Birmingham 1.

BENNISON, Peter, A.R.Ae.S., Deputy Materials Engineer, Research Laboratory, Rolls Royce, Ltd., Derby.

CLARK, Peter Howard, Technical Manager, McKechnie Bros., Ltd., Stratford, London, E.15.

GLASER, Frank W., M.S., Physicist, American Electro Metal Corporation, 320 Yonkers Avenue, Yonkers 2, N.Y., U.S.A.

HOCH, Alois, Dr.rer.nat., Metal Finishing Consultant, Radevormwald-Bergerhof, (Rhld.), Germany.

MCPHERSON, Donald James, B.Met.E., M.Sc., Ph.D., Supervisor, Physical Metallurgy Research, Armour Research Foundation, 35 West 33rd Street, Chicago 16, Ill., U.S.A.

MEIJERS, Cornelius Johannes, Chief of the Metallurgical and Technical Department, H. Drijfhout & Zoon's Edelmetaalbedrijven N.V., Amsterdam, Netherlands.

van OUWERKERK, *Ir.* L., JMLzn., Manager, Röntgen Technische Dienst N.V., Overschie, Netherlands.

van REENEN, *Ir.* J. C., Junior Managing Director, Rotterdamsche Droogdok Maatschappij N.V., P.O. Box 913, Rotterdam, Netherlands.

SAĞISMAN, Muzaffer, Dr.-Ing., Istanbul Teknik Üniversitesi, Makina Fakültesi Teknoloji Kürsüsü, Istanbul, Turkey.

STAP, *Ir.* M., Engineer, Nederlands Koperenbuinen Fabriek, Leiderdorp, Netherlands.

WHINFREY, Charles George, B.Sc., Research Chemist and X-Ray Crystallographer, Research Centre, Johns-Manville Corporation, Manville, N.J., U.S.A.

As Student Members

BUCKLEY, S. N., B.Sc., Research Student, Department of Metallurgy, University of Manchester.

HAHN, Henry, M.S., Graduate Student and Research and Teaching Associate, Department of Metallurgy, Rensselaer Polytechnic Institute, Troy, N.Y., U.S.A.

JONES, Richard Lee, M.S., Graduate Student of Metallurgical Engineering, University of Michigan, Ann Arbor, Mich., U.S.A.

KING, Alan John, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

MIKUS, Emil Bill, M.S., Graduate Student of Metallurgy, University of Michigan, Ann Arbor, Mich., U.S.A.

The following 8 Ordinary Members, 1 Junior Member, and 4 Student Members were elected on 30 April 1953:

As Ordinary Members

BLAKE, Albert Leonard, General Manager, Cable Makers Australia Pty., Ltd., Liverpool, N.S.W., Australia.

CALVO, Felipe Angel Calvo, Chemist, Instituto de la Soldadura, Goya 58, Madrid, Spain (temporarily at the Department of Metallurgy, University of Cambridge).

DIRKSE, Philippe Marie Jean Louis, Dipl. Ing., Chief, Central Repair Department, N.V. Oranje Nassau Mijnen, Heerlen, Netherlands.

FRANKEL, Herbert A., B.S., Secretary, Foundry Services, Inc., 280 Madison Avenue, New York 16, N.Y., U.S.A.

JOLLIE, Andrew, M.Eng., B.Sc., M.I.Mech.E., M.Inst.F., Director and General Manager, Steel, Peech and Tozer (Branch of The United Steel Companies, Ltd.), The Ickles, Sheffield 1.

REEVE, *Major* Reginald William, T.D., Technical Director, South African Light Metal Works (Pty.), Ltd., P.O. Box 4305, Cape Town, South Africa.

SCHWOPE, Arthur D., B.Ch.E., M.S., Supervisor of Mechanical Metallurgy, Battelle Memorial Institute, Columbus, O., U.S.A.

YOUNG, Melvin H., B.S., Director, Engineering Production Services, Wright Aeronautical Division, Wood Ridge, N.J., U.S.A.

As Junior Member

MUKHERJEE, Prabhat Charan, B.E., 16 Hilgrove Road, London, N.W.6.

As Student Members

EVANS, Dain Stedman, Student of Metallurgy, University College, Swansea.

GLOVER, Stanley George, B.Sc., Research Student, Department of Metallurgy, University of Birmingham.

STOBO, James Jack, B.Sc., Research Student, Metallurgy Department, Royal Technical College, Glasgow.

THORNEWELL, Gordon Leonard, Laboratory Assistant, The Mond Nickel Co., Ltd., Development and Research Laboratory, Wiggin Street, Birmingham 16.

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Bailey, Mr. G. L.
Graham, Mr. A. B.
Murphy, Professor A. J.
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Smithells, Dr. C. J.
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Cunniffe, Mr. P. W. A. (*Honorary Secretary, South Wales Local Section*).

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Hanstock, Dr. R. F.
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King, Mr. R.
McLean, Mr. D.
Nutting, Dr. J.
Oliver, Mr. D. A. (representing the Iron and Steel Institute and the British Iron and Steel Research Association).
Richards, Dr. T. Ll.
Sully, Dr. A. H.
Wakeman, Dr. D. W.

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Inglis, Dr. N. P.
Jenkins, Dr. Ivor.
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Phillips, Mr. H. W. L.
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Baer, Mr. A. (*Chairman, Finance and General Purposes Committee*).
Jones, Mr. E. H. (*Honorary Treasurer*).
Raynor, Professor G. V. (*Chairman, Metal Physics Committee*).
Thomas, Mr. W. J. (*Chairman, Metallurgical Engineering Committee*).

The constitution of certain Committees is fixed by standing orders of the Council, as follows :

Medal Committee

President (*Chairman*).
Senior Vice-President

and

Not more than four Institute of Metals (Platinum) Medallists who are, or have been, members of the Council (to be selected by the President), with power to the President to co-opt not more than two other persons.

Nominations Committee

President.
Two Immediate Past-Presidents.
Senior Vice-President.

PERSONAL NOTES

Mr. J. R. Abraham has left for New Plymouth, New Zealand, to take up the appointment of Managing Director of McKechnie Bros. (N.Z.), Ltd.

Professor Sir Lawrence Bragg has been appointed Fullerian Professor of Chemistry at the Royal Institution and Resident Professor and Director of the Davy Faraday Research Laboratory. He will take up the duties of the Fullerian professorship at once and the laboratory and resident duties on 1 January 1954.

Mr. A. R. G. Brown has left the British Iron and Steel Research Association and joined the staff of the Royal Aircraft Establishment, Farnborough, as a Senior Scientific Officer.

Dr. G. P. Contractor, who has been Acting Director of the National Metallurgical Laboratory of India for three and a half years, relinquished his position with the Indian Council of Scientific and Industrial Research at the end of April. He is leaving India in August, and will visit Great Britain on his way to Canada.

Dr. F. A. Fox delivered the guest lecture at the Annual General Meeting of the Australian Institute of Metals held last month in Brisbane.

Professor André Guinier delivered the Robert S. Williams lectures in the Department of Metallurgy, Massachusetts Institute of Technology, on 10–12 March. He spoke on "Precipitation Phenomena in Supersaturated Alloys" and "Recent Progress in X-Ray Crystallography".

Mr. M. M. Hallett has been appointed Director of Sales to Sheepbridge Engineering, Ltd., Chesterfield. He was previously Research and Development Manager to the Company.

Mr. H. E. Jackson has relinquished the Chairmanship of the Metals Division of Imperial Chemical Industries, Ltd. He will retire from the Metals Division at the end of June when his period of office as President of the British Non-Ferrous Metals Federation comes to an end.

Mr. J. F. B. Jackson, who has been Director of Research to the British Steel Founders' Association since the formation of its Research and Development Division in 1949, has been appointed Director of the recently formed British Steel Castings Research Association.

Mr. F. A. Lytton has taken up a post with William Jessop and Sons, Ltd. He was previously with The Morgan Crucible Co., Ltd.

Mr. J. A. Matthews has been made a Liveryman of the Worshipful Company of Farriers and a Freeman of the City of London.

Mr. W. J. Neagle-Jones has been appointed Chief Metallurgist to Brockworth Engineering Co., Ltd., Hucclecote, Gloucester.

Mr. C. E. Prosser has been appointed Chairman of the Metals Division of Imperial Chemical Industries, Ltd., in succession to Mr. H. E. Jackson.

Mr. K. Sachs has been awarded the Ph.D. degree of London University.

Mr. A. W. Trash, Chief Chemist to H. J. Enthoven and Sons, Ltd., has retired after 44 years' service with the Company. He will, however, continue to act as Consultant to the Enthoven Organization.

OBITUARY

Mr. Douglas Jepson

The death of Mr. Douglas Jepson, at the age of 51, occurred on 24 February 1953 after a short illness.

Mr. Jepson was born at Glossop, Derbyshire, and passed from the Grammar School there to the University of Manchester, where he graduated in 1922 with 1st Class Honours in Metallurgy. After a period of research work, the degree of Master of Science was conferred on him in 1923. From 1924 to 1927 he held the post of Assistant Lecturer in Electrochemistry at the University of Manchester. In 1927 he entered industry, and from then until 1940 he held a number of appointments with I.C.I., Ltd. (General Chemical Group) at Middlewick, Runcorn, and Billingham, where he gained valuable experience as a metallurgist, in plant management, and in special problems arising in industry. In 1940 he returned to the teaching profession as Lecturer in Metallurgy at Bradford Technical College and went from there to Birmingham, where he was appointed Senior Assistant in the Department of Metallurgy at the then Central Technical College, in 1945. Two years later he was appointed Head of the Department of Metallurgy at the College of Technology in succession to the late Dr. F. Johnson.

In this capacity he had been much occupied with the planning and equipment of the metallurgy laboratories in the new College now in course of erection. In addition, he was an active member of a number of Committees, including the Service and Publicity Committee and the Standing Committee on Matters Concerning Metallurgical Education and Training, both of the Institution of Metallurgists; the Joint Committee on Metallurgical Education of the principal metallurgical institutions in the country; the Education Committee of the Institute of British Foundrymen; the West Midlands Advisory Council for Technical, Commercial, and Art Education; and the Council of the Birmingham Metallurgical Society. In addition, he was a member of the Iron and Steel Institute, the Institution of Metal Finishing, and the American Society for Metals. He was elected a member of the Institute of Metals in 1920. Mr. Jepson also took a keen interest in the affairs of the Parish Church at Erdington, Birmingham, and was one of the Managers of the School associated with that Church.

He is survived by a widow, two sons, and a daughter.

W. E. Thorneycroft.

LETTERS TO THE EDITOR

The Development of a High-Strength Non-Magnetic Alloy

The magnetic measurements made by the Danish deep-sea expedition in the ship *Galathea* in 1950–52 have been reported recently.[1]

To contain the magnetometers used in these measurements, spheres were required which would withstand the water pressure even at the greatest depths. Calculation showed that the material for the sphere must have a yield strength in compression of at least 28 kg./mm.^{2}, and further it had to be completely non-magnetic. As the spheres were very heavy, with a wall thickness of 4 in. and a finished weight of 2500 lb., it was decided to cast them.

None of the existing casting alloys proved suitable, however. A special austenitic steel that was tested had adequate strength, but failed in the magnetic tests under stresses in the region of the yield point. The only light alloy which was sufficiently strong had to be forged and heat-treated, and the cost of the spheres would have been very high. Non-magnetic aluminium bronzes were considered to be too difficult to cast, and their yield strength too low. Beryllium bronze might have been sufficiently strong, but experience in the casting and heat-treatment of such large objects was lacking. Manganese bronze without iron has normally too low a yield point, and with iron it is slightly magnetic.

It was necessary, therefore, to develop a new alloy, and in the spring of 1950 the writer undertook this task, in the laboratories of Paul Bergsøe and Son, Copenhagen. Some iron-free manganese bronzes were first tried, and it was confirmed that iron was a necessary constituent if a sufficiently high yield point was to be attained. It was then decided to replace iron by another, non-magnetic, element. In manganese bronzes iron is to be found as a "blue phase", which consists of iron saturated with copper, and which seems to influence the grain-size of the material considerably. It appeared to be necessary to find an element which would behave similarly to iron in copper alloys, and chromium was accordingly tried. It showed the same effect as iron, and finally an alloy was made which fulfilled all requirements.

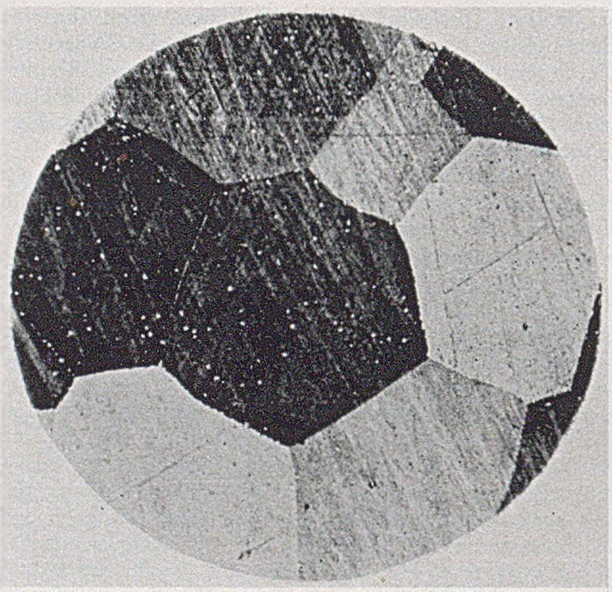

FIG. 1.—Structure of Galathea Bronze. × 150.

The composition chosen was quite unorthodox: copper 62, aluminium 5, manganese 2, chromium 0·5%, balance zinc. Even with 0·1% iron, it was completely non-magnetic. Mechanical tests showed that the yield point in compression was very similar to that in tension, being 38 kg./mm.^{2} (0·2% permanent set) and the alloy appeared to be fully elastic up to 28 kg./mm.^{2}. The tensile strength was 70–80 kg./mm.^{2}, with an elongation of 10–20%. The alloy was no more difficult to cast than the ordinary manganese bronzes and was not liable to absorb gases, owing to the presence of zinc. The castings were completely pressure-tight.

The microstructure contained β crystals and small particles of "blue phase" (Fig. 1). The alloy was very susceptible to stress-corrosion, but as the spheres were to be exposed to sea-water only intermittently and for a short time only on each occasion, this was considered unimportant. No trouble was experienced in this respect even in the most highly stressed parts of the sphere, namely, the bolts with which the two halves were connected (Fig. 2).

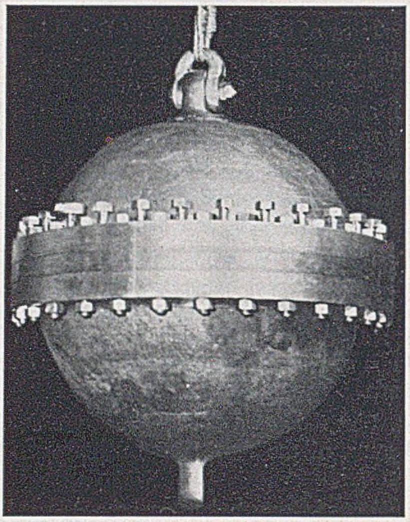

FIG. 2.—Sphere Used by Danish Deep-Sea Expedition.

The alloy has been named "Galathea bronze", and the spheres have been tested several times in the Philippines trench, where one sphere touched bottom at about 10,000 m.

BØRGE LUNN

Copenhagen.

REFERENCE

1. *Nature*, 1953, **171**, (4348), 383.

Crystallization of β-Brass

The dendrite is, as it were, the metallurgist's *pièce de résistance*, and recently the subject of dendritic crystallization has been receiving further attention. β-brass is interesting in this connection because it is representative of a large class of alloy constituents. Normally, no evidence of dendritic growth is obtained from the microstructure; the grains are

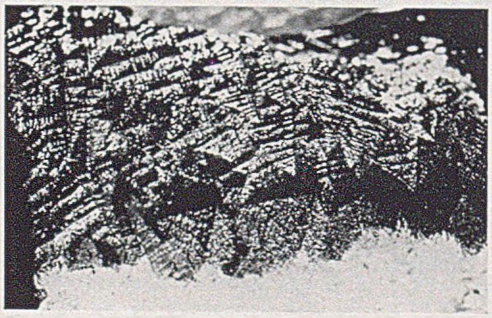

FIG. 1.—Dendrites in the Secondary Pipe of Sand-Cast β-Brass. × 5.

homogeneous in appearance and the smooth boundaries rather belie dendritic formation.

However, it has been possible to show that dendritic crystallization can occur in β-brass (made from cathode copper and 99·99% zinc). Clear evidence has been found in the secondary pipe of sand castings made in this alloy (Fig. 1). Further, dendrites have been detected in samples chill cast in contact with molten lead (Fig. 2).

From consideration of the equilibrium diagram, selective crystallization is to be expected in β brass, resulting in coring with practical rates of cooling, as occurs in the α phase. The writer has examined microscopically many samples of cast β-brass and in none detected the slightest signs of coring. In addition, analysis of samples taken from the outside and centre of cast material, even when quenched from a high temperature, has as yet failed to indicate any segregation which might arise from selective crystallization. The apparent explanation would be that the atomic movements accompanying the subsequent ordering process are sufficient to even out the composition, rather than that the rate of diffusion of zinc in the β structure is markedly greater than that in the α structure.

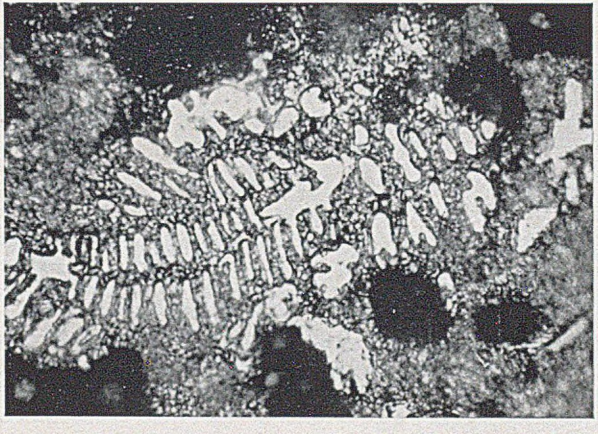

FIG. 2.—Dendrites of β-Brass (light) in Matrix of Lead. × 600.

A. R. BAILEY

Constantine Technical College,
Middlesbrough.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

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Asst. Hon. Secretary : P. Hutchinson.

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London Local Section

At a meeting of the Section held at 4 Grosvenor Gardens, London, S.W.1, on 5 March, Dr. B. A. Bilby gave a lecture on:

Dislocations in Crystals

This lecture, which was arranged by the Metal Physics Committee, has been given before a number of Local Sections and Associated Societies, and a summary of it will be found on p. 179 of the *Bulletin* for March 1953.

JOINT ACTIVITIES

Sir George Beilby Memorial Award, 1952

The Administrators of the Sir George Beilby Memorial Fund, representing the Institute of Metals, the Royal Institute of Chemistry, and the Society of Chemical Industry, have made an award from the Fund for 1952 of one hundred and fifty guineas to

Thomas Victor Arden, B.Sc., F.R.I.C.,

in recognition of his experimental work on the hydrometallurgy of uranium, with particular application to the separation of uranium from low-grade ores. This work has been carried out at the Chemical Research Laboratory, Teddington, since 1946, though little of it has been published so far. Mr. Arden has also taken a leading part in the development and operation of treatment plants for uranium ores in various parts of the world. He has recently left the Chemical Research Laboratory to join the Permutit Co., Ltd., as head of the Applied Ion-Exchange Department.

Awards from the Fund are made to British investigators in science as a mark of appreciation of distinguished work, particularly in such fields as fuel economy, chemical engineering, and metallurgy in which Sir George Beilby's special interests lay. In general, the awards are not applicable to more senior investigators, but are granted as an encouragement to relatively young men who have done independent work of exceptional merit over a period of years.

NEWS OF KINDRED SOCIETIES

Deutsche Gesellschaft für Metallkunde: General Meeting, Nürnberg, 4–6 September 1953

The General Meeting of the Deutsche Gesellschaft für Metallkunde will be held in Nürnberg from Friday to Sunday, 4–6 September, inclusive. The main subject for discussion will be "The Hot Working of Non-Ferrous Metals". Further details will be published in due course.

Members of the Institute of Metals will be welcome at this meeting, the full programme of which can be obtained on application to The Secretary, Deutsche Gesellschaft für Metallkunde E.V., Alteburger Str. 402, Köln/Rh., Germany.

Société Française de Métallurgie: Autumn Meeting, Paris, 19–23 October 1953

The Journées Métallurgiques d'Automne, organized by the Société Française de Métallurgie, will take place in Paris from Monday to Friday, 19–23 October.

A cordial invitation for members of the Institute of Metals to attend this meeting is extended by the Société.

Particulars of the meeting, and of membership of the Société, may be obtained from the Secrétaire-Général, Société Française de Métallurgie, 25 rue de Clichy, Paris (9e).

OTHER NEWS

Groupement pour l'Avancement des Méthodes Spectrographiques

The sixteenth congress of the Groupement pour l'Avancement des Méthodes Spectrographiques will be held on 24–26 June 1953 at the headquarters of the Association des Ingénieurs des Arts et Métiers, 9 bis Avenue d'Iéna, Paris. Technical sessions will take place on the afternoon of 24 June and the mornings of 25 and 26 June. There will be a visit to the Renault laboratory and works on the afternoon of 25 June, and an excursion is being arranged for the afternoon of 26 June. Full details may be obtained from G.A.M.S., 1 Place Saint-Thomas d'Aquin, Paris (7e).

Exhibition of X-Ray Crystallographic Equipment

The X-Ray Analysis Group of The Institute of Physics announces that its autumn conference will be held in London on 20 and 21 November, and an exhibition of X-ray-diffraction equipment will be an important feature of it. Offers of exhibits are invited under two headings: (*a*) apparatus commercially available in this country and (*b*) examples of recent developments in X-ray crystallographic equipment that have been made in universities and other research centres in this country. Examples of the kind of exhibit envisaged are: X-ray tubes, diffraction cameras of all types, microbeam techniques, counters and counter-spectrometers for diffraction work, any aids to interpretation, monochromators, microdensitometers, travelling microscopes, X-ray films and photographic accessories, and spectrometers for fluorescent analysis. Owing to the limited space available, a selection will no doubt have to be made from the offers submitted.

Offers of exhibits, giving details of approximate bench and floor space required and of any services needed, should be submitted before 1 September to the Conference Secretary, Mr. H. J. Goldschmidt, F.Inst.P., c/o The Institute of Physics, 47 Belgrave Square, London, S.W.1.

Appointments Vacant will be found on p. (a), *at the end of "Metallurgical Abstracts".*

43rd ANNUAL MAY LECTURE

THE PRESENT AND FUTURE METALLURGICAL REQUIREMENTS OF THE CHEMICAL ENGINEER *

By SIR CHRISTOPHER HINTON,† M.A., M.I.C.E., M.I.Mech.E.

SYNOPSIS

Up till about 1925 the designer of chemical plant made use only of the range of constructional materials that had been developed for other purposes. In the last twenty-five years, however, the chemical engineer has made increasing demands on the metallurgist for new materials to meet his special requirements. In particular, the plant needed for operations in connection with the exploitation of atomic energy has presented a number of metallurgical problems. A brief survey of some of these is given.

It is suggested that a better range of halide-resistant materials is necessary, and that greater attention should be paid to methods of fabricating new materials.

I.—INTRODUCTION

Engineering, like metallurgy, is not an exact science; indeed, engineering design is best thought of as an art with a scientific basis, and therefore the final solution to a design problem is a compromise between the conflicting advantages and disadvantages of many alternative solutions which present themselves. It follows that, although in all design work there are certain fundamental rules of soundness and purity which must normally be respected, the application of those rules and the finished form of the design will depend on the taste and training of the designer. There are usually alternative solutions to every design problem, all equally acceptable; which of these is adopted depends on the background against which the designers work; some of this background is peculiar to the individual, but much of it is common to all those engaged within a particular field. Because of this, we should be wise, before we survey the materials of construction which the chemical-plant designer needs or will need, to consider the tradition of design which forms the background against which he works.

The foundations of the modern heavy-chemical industry lie in the second half of the 18th century, when the manufacture of sulphuric acid, bleaching powder, and black ash began on an industrial scale. Throughout the Industrial Revolution there was a steady expansion. The manufacture of heavy alkalies was put on its present basis by the introduction of the ammonia soda process in 1873 and the invention of the electrolytic soda process in 1890; the contact process for the manufacture of sulphuric acid was made practicable in about 1870. Meanwhile similar developments were taking place in the industries centred around coal tar. By 1802 Boulton and Watt were lighting their factory at Birmingham by coal gas, and from the growth of the coal-distillation industry arose the developments of synthetic dyestuffs which continued throughout the 19th century, though the initiative was largely lost to Germany before the First World War.

That war called for an expansion of the chemical industry on a scale which was without precedent; later, it will be noted that the last war stimulated the beginnings of atomic energy in a similar way. There had been great developments in the explosives industry under Nobel, but in 1914 T.N.T. had only just been approved as a high explosive, and the demand for it and for cordite led to a great programme of factory building which was carried out by the Government under Lord Moulton and Quinan. At the end of the war Britain, having learned the seriousness of lacking supplies of ammonium nitrate and the danger of not having a well-developed coal-tar industry, started the fixation of atmospheric nitrogen and encouraged the growth of the dyestuffs industry. The formation of great combines of chemical manufacturers, which followed in the late 1920's, left the business organization of the industry roughly in the form in which we know it today.

II.—MATERIALS IN USE UP TO 1925

Until 1925 the chemical engineer had never demanded from the metallurgical industry the materials which he preferred or required for the manufacture of his plants; he had taken the materials which had been developed for other industries, and had devised processes which could be worked in vessels made of the available materials. The foundry industry had developed rapidly during the Napoleonic wars to

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* Delivered at the Annual General Meeting, London, 23 March 1953.

† Deputy Controller of Atomic Energy (Production), Ministry of Supply.

supply the Navy and Army with guns and shot, and it had grown still further to meet the requirements of the engineering industry in the Industrial Revolution. The steel industry had grown largely to meet the needs of transport—the railways and shipping companies; the lead industry largely to meet the needs of the building industry; and the copper industry those of the domestic user, the builder, and the shipwright.

The chemical engineer seized on these and on ceramic materials, all of which had been intended for other industries, and used them skilfully to build his plants. He did little to demand that materials should be invented to meet his requirements, and so his vessels were made of cast iron or mild steel, lined with blue brick if necessary; they were of timber lined with lead, or of copper. The chemical processes had to be so devised that they could be worked in these vessels. The engineering and metallurgy of chemical plants followed the growth of technique in the general engineering industry—it did not lead it.

III.—PRESENT POSITION REGARDING MATERIALS FOR THE CHEMICAL ENGINEER

The late 1920's marked a turning point, and from then onwards chemical engineers began to demand the materials which they require for the construction of their plants. It is interesting to speculate on how the change came about; perhaps it was due to the general development and growth of the chemical industry which was then taking place in Great Britain, the construction of dyestuffs plants, plants for cellulose chemistry, and plants for high-pressure reactions; perhaps the formation of the large combines placed chemical firms in a stronger position to state their needs. Possibly the new outlook did not arise entirely from changed conditions in this country; in the late 1920's the American oil industry was expanding rapidly, it was young and energetic, the efficiency of many of its processes could be improved by using higher pressures and temperatures, and it could afford to take risks. It did not hesitate to specify its metallurgical requirements and to demand improvements, and the line it adopted may have influenced us.

Perhaps all these factors and others acted together to bring about the change, but in the metallurgical industry the conditions were favourable. Most of the demands of the chemical industry are for materials which are corrosion- and erosion-resistant, and in 1913 Brearley had invented stainless steel. The original material, containing 12–14% chromium and 0·3% carbon, could be hardened by heat-treatment and was suitable for the cutlery industry; it was of little use to the chemical engineer, however, as the range of chemicals to which it was resistant was limited and it was not easily fabricated. Development was retarded by the First World War, but in 1923 it was found that the corrosion-resistance could be increased by higher chromium contents, and in 1925 the 18% chromium, 2% nickel steel, now covered by British Standard S80. was evolved. This martensitic steel was still of only limited interest to the chemical engineer, but the later development of the austenitic, and to a lesser extent the ferritic, stainless steels went far to meet his requirements.

Simultaneously with these advances in ferrous metallurgy there had been a corresponding widening in the range of non-ferrous alloys available for chemical plant construction. The manufacture of nickel had been put on an industrial basis at the beginning of the century, and at the end of the 1914–18 war it was important as an alloying material in ferrous metallurgy and as a plating material. In the period of which we are now talking its merits and the merits of non-ferrous nickel alloys as corrosion- and heat-resisting materials were realized. To complete our outline picture of the second quarter of the century, we should call to mind the progress which took place in the non-metallic protection of plant by rubber and glass covering, and in the use of plastics.

It is reasonable to ask ourselves what the chemical engineer has succeeded in doing with these great new facilities. It is doubtful whether he can claim correspondingly to have broadened the number of generic varieties of chemical products which are marketed, though, within a rather wider generic field, he can claim considerably to have increased the number of species. But he can certainly claim that he has made good use of the new materials to evolve more elegant and more economical plants. Possibly the best single example of this is in the manufacture of nitric acid. In our first period it was made in large quantities from Chile nitrate in plants built mainly of cast iron and acid-resisting brick; but in the second period the use of the cheaper ammonia-oxidation process was made possible only by developments in the austenitic stainless steels.

IV.—PROBLEMS IN FABRICATING CHEMICAL PLANT

1. Conditions of Oxidizing Corrosion

With this background we can ask ourselves what further developments may reasonably be asked for by the chemical engineer. In answer we can say that not only must he have satisfactory materials but he must also have the techniques necessary for their use; too often these techniques are not immediately available and have to be developed with some loss of time, to meet specific problems. The austenitic steels provide an example.

Modern austenitic stainless steels give very good resistance to corrosion under oxidizing conditions, but the chemical engineer is not concerned with the properties of the plate so much as with the properties of the vessels and plants which he can fabricate from it, and it is not possible to have the same confidence in the welded joint of a stainless-steel vessel as in the parent metal. Under conditions which are not too rigorous it is normal to use a steel containing 18% chromium and 8% nickel, with a stabilizer chosen to suit the conditions; although this steel is commonly called austenitic, it has in reality a duplex structure containing both austenite and ferrite. Under severe corrosion conditions, such a steel is often found to be more readily attacked than one free from ferrite. To eliminate the duplex structure and to secure a completely austenitic matrix, the nickel content must be increased and on an average the safe minimum nickel content for a fully austenitic steel is 13%. To avoid the weld-decay type of corrosion, additions of strong carbide-forming elements are necessary, and where a great deal of cross-welding is expected niobium is generally preferred to titanium. A steel of this type containing 19% chromium, 13% nickel, and 1% niobium can have many uses in chemical engineering; in the form of plate it also has a high resistance to nitric-acid corrosion, but the conditions in welded joints require further consideration. If the electrode deposit is fully austenitic, there is grave danger of producing fine cracks in the weld. In the adjoining zone the liability to cracking in the plate owing to faulty welding technique is accentuated.

Faced with this problem in our plants, we have adopted the compromise solution of using an 18 : 8 : 1 welding electrode which gives a certain amount of ferrite in the weld metal, though this is not a solution about which one can feel entirely happy. In the chemical plants associated with atomic-energy processes it often happens that, because of the radioactivity of the chemicals handled, it is impossible to approach vessels to carry out examinations or repairs, and in these conditions we have been particularly alive to problems of fabrication. We should feel safer if we could guarantee both freedom from cracks and freedom from ferrite.

We have met similar problems in making welded joints in stainless-steel pipe lines. In plants where conditions are not too severe and where repairs are possible, it is satisfactory to make such joints by metallic-arc welding, but this gives rise to dangers, either of incomplete penetration or of too large and irregular an internal bead; both conditions can encourage corrosion. We have evolved a reasonable solution by making the first run of weld metal by the argon-arc process. The ends of the pipes are machined to a modified J-end having a fairly thin root face. This enables a small weld pool to be maintained during the argon-arc run, making possible adequate control of penetration, and the edges are fused together giving a fully austenitic weld on the inside of the pipe. The remainder of the weld is filled by metallic-arc welding with metal with a low ferrite content. Thus the weld metal exposed to the corrosive liquors is of the same composition as the body of the pipe. To prevent the under side of the weld from becoming heavily oxidized, it is necessary to provide argon backing inside the tube during welding. This is done by inserting removable sponge-rubber plugs in the pipe on each side of the weld and admitting argon into the space between them. For closing welds when sponge-rubber plugs cannot be withdrawn, balloons are inserted on either side of the joint and inflated to provide the seal; these balloons can be withdrawn when welding is complete.

The welded seams on all vessels which have to contain highly radioactive material are X-rayed, and all welded joints in pipes which work under similar conditions are either internally examined with an Introscope or, where this is not possible, are X-rayed.

A major difficulty has been to define the standard of acceptance; X-ray examination of welds has been developed mainly for the examination of Class A pressure vessels, and the standard of acceptance is based not on any absolute or scientific level of perfection but on a comparison of X-ray plates with similar photographs of welds which have or have not proved satisfactory. It did not follow that the same criteria would be applicable to our requirements, and a great deal of experimental work was necessary to determine which characteristics on an X-ray plate should lead to rejection and which were tolerable. For the development of such new techniques as these, we were mainly dependent on our own resources, though we had most willing and sympathetic collaboration from the fabricators.

But the picture that I want to present is that, when faced with a new problem in chemical-plant design and construction, it is comparatively easy to select a material which will reasonably meet our requirements, but it is much more difficult to devise fabrication techniques for this material. It seems to me that this is a common difficulty in the development of chemical plant, and it has certainly been met with in the history of stainless steel since the early days of its use in the chemical industry. The problem is not confined to stainless steels, and one can think of a number of cases of materials which in billet or plate form exhibit most valuable properties of corrosion-resistance, but where the technique of fabrication still presents serious difficulties and limitations.

This forms the basis of the first plea that I should like to make on behalf of the chemical engineer—that methods of fabrication should keep pace with developments in basic materials of construction. I suggest that there is too great a tendency to carry out laboratory corrosion tests on plate or billet samples when, in fact, the important tests are those on welded joints or on samples cut from castings. It is wrong and uneconomical that the cure for weld decay should be found only after extensive trouble has been experienced on full-scale plant.

The difficulty is not an easy one to solve, since collaborative effort is necessary. Some responsibility must unquestionably rest with the user, who, I think, must state his requirements more clearly, not merely in respect of the properties that he requires in his materials of construction, but also in respect of the sort of plant into which he expects to fabricate these materials. But I think that the main responsibility must inevitably fall on the metal manufacturer and that he ought to develop a greater consciousness of the problems of fabrication. Many of the large metallurgical firms have contracting engineering workshops of their own. I suggest that these workshops might be encouraged to take an interest in fabrication problems of new materials that are being developed. Where the metallurgical manufacturer does not possess such facilities, he might make a collaborative arrangement with one of the manufacturers of chemical-plant vessels to achieve the same advantages.

2. CONDITIONS OF HALIDE ATTACK

We have said that, provided suitable fabrication techniques are worked out, the austenic steels meet most of the reasonable requirements of the chemical engineer in oxidizing corrosion conditions. But there are limitations to their use; firstly, their resistance results from the formation of a protective surface film, and for this reason they are not fully satisfactory where there is erosion; and secondly, their resistance to corrosion is destroyed in the presence of halides. In regard to the first of these limitations, almost all the corrosion-resistant materials owe that property to the formation of a protective film, and the film on stainless steels and on nickel and its alloys is so thin, tough, and adherent that it would, in my opinion, be unrealistic to expect much improvement in its mechanical properties. I think that the engineer must expect to have to deal with the erosion element of the erosion-corrosion problem by skilful design, rather than look to the metallurgist to provide a solution.

The position in the field of halide attack is far less satisfactory than in the field of oxidizing attack. In the case of dry halides, mild steel can be satisfactorily used provided the temperature is not too high; even fluorine can be handled in the dry state in mild-steel plant, provided the temperature is below 200° C. Above this temperature violent attack can take place. The temperature limitation can be raised by using nickel or Monel, but even these are not safe above 500° C.

Where hydroxyl ions are present, the problem becomes very much more difficult. In the case of hydrochloric acid and its corrosive compounds chemical engineers usually take refuge in the use of glass- or rubber-lined vessels. Glass-lined vessels have limitations in size and complexity, while rubber-lining cannot be employed at temperatures much above 90° C. if it is to give a reasonable life. The only metals which give good corrosion-resisting properties in this field are the nickel–molybdenum alloys in the Hastelloy group, but these are extremely difficult to fabricate.

The position in regard to bromine is still less satisfactory, since rubber can be less readily utilized, but one meets the greatest difficulty in dealing with the fluorine compounds. Here not only rubber-lining but also glass-lining and ceramics are normally excluded. Mild steel, nickel and nickel alloys, copper, and aluminium can be used with suitable temperature limitations and in "dry" conditions, but these are very severe limitations to place on the plant designer. Thus the second reasonable request that the chemical engineer may make to the metallurgist is, for a better range of materials to resist halogen action, and within this field I stress the need for materials to withstand fluorine attack. I believe that this is of peculiar importance, as there is likely to be great growth in the status of fluorine chemistry in the next ten years. The only substance that can at present be recommended for use under severe conditions of fluorine attack is graphite, and even with modern developments there are limitations to the plants and vessels which can be built of this material. I think that the chemical industry itself has not sufficiently realized its responsibility in regard to developments in this field; for the construction of chemical plants one needs not only metals or other materials of which to build plant vessels, but also such auxiliaries as jointing material, gland packing, and gland seals. The efforts made to produce such materials have not been sufficiently determined, and unless steps are taken to provide better materials for the construction of plants subject to fluorine corrosion, I think that this country may find that it is left behind in a new and important field of development.

3. OTHER PROBLEMS

We have now examined the position in two broad generic groups which we have called the oxidizing-corrosion problem and the problem of halide attack. There remain two other groups: those of high-temperature and of high-pressure. In the first of these I include processes where the type of corrosion of which we have already spoken does not occur, but where the temperature is beyond that normal in chemical-plant practice and the creep of metals becomes a dominant design feature. I would suggest that in this field the requirements of the chemical engineer are likely to be covered by those of the gas-turbine industry. If that is the case, the chemical engineer does not need to formulate his own demands, but can follow in the wake of the developments being made for the aircraft-engine and power-plant industries, watching only the problem of fabrication to ensure that a reasonable number of the heat-resisting alloys produced can be formed into the shapes in which he needs to use them. The problem of high-pressure chemical plant is very specialized, and it is not a field in which I have experience.

Before we leave orthodox chemical engineering there is one other problem which we ought to consider. A great deal of corrosion-resistant material is used for chemical-plant construction where conditions are not strictly corrosive. Stainless steel, nickel, and nickel alloys, are extensively used in plants for the food and pharmaceutical industries to ensure the biological cleanliness necessary there; in yet other processes the chemical industry is no longer content with the dark and dirty plants so common twenty years ago, and is rightly aiming at a far higher standard of amenity. The plant designer has found it perhaps too easy to provide the higher standard demanded of him by using expensive corrosion-resisting alloys merely to secure cleanliness. This is a question to which the metallurgical industry might wisely give attention. Unless they are able to provide a cheaper means of providing materials which give cleanliness, there may be a swing towards the use of inexpensive plastic materials in these conditions. There is already some indication that this is taking place in the food industries.

With this question of cleanliness and ease of maintenance of plant one ought to associate the problem of building maintenance which is often so heavy in the chemical industry. In a great many cases the chemical engineer should eliminate this problem, as has been done in the oil industry, by centralizing the controls and instruments and placing the plant in the open. But there are plants where this is not possible and where corrosion of the building is severe. Reinforced concrete construction is not a satisfactory solution, because of the fear that corrosive liquids may penetrate to the reinforcing rods and for other reasons. In these cases a structural steel which could be kept in good condition at reasonable cost would be very helpful.

V.—MATERIAL REQUIREMENTS FOR ATOMIC ENERGY

I should like now to speak of some of the metallurgical problems which arise in connection with atomic-energy development. These can be divided into two groups : those associated with the atomic piles and those associated with the ancillary chemical plants. The chain of processes involved in the manufacture of fissile material starts with the extraction of uranium from its ore, followed by its purification and reduction to the metallic form. In this form it is made into rods which are enclosed in aluminium cans. The canned uranium rods are placed in the atomic-energy pile, where part of the uranium is converted into plutonium. The irradiated rods containing this plutonium and the associated fission products are discharged from the pile; they are then extremely radioactive and can only be handled remotely. The irradiated rods go to the chemical-separation plant, where they are dissolved in nitric acid and processed to separate an impure plutonium, impure uranium, and the fission products. The plutonium goes through a purification process and is reduced to metallic form; the fission-products solution, which contains most of the radioactivity, is stored indefinitely.

It is not widely realized that in these early days of atomic energy, more than half the industrial effort and technical difficulty lies, not in the piles but in the chemical plants, and these problems are engineering, chemical, and metallurgical in nature. This will probably not always be the case and, as the use of atomic energy for power production develops, the volume of effort will probably swing in the other direction, but we shall consider first of all the present and future metallurgical requirements of the chemical plants. The processes for the manufacture of uranium from ore are fairly conventional; they involve corrosion problems throughout, but these are well covered by the suggestions already put forward in connection with the development of materials to withstand oxidizing corrosion and halide attack.

The chemical plants in which the separation of the plutonium and fission products is carried out, after the uranium has been irradiated, are unconventional in that they are handling materials of such high radioactivity that the plants must be remotely controlled, and many of these plants cannot be approached for repair when they have once operated. Basically, therefore, the problem of materials of construction is the same as for other chemical plants which are subject to oxidizing corrosion, but the problem is made more difficult and interesting by the consideration that repairs cannot be carried out, and failure must therefore be avoided. Very careful consideration of methods of fabrication has therefore been necessary. We believe that the solution to many of our problems can be found with fairly conventional materials, providing use is made of the very finest techniques in design and fabrication.

The metallurgical problems in connection with the atomic piles are quite different and have nothing in common with those of the chemical plants. Piles (or as they are now more often called reactors) can be divided into two broad classes, thermal reactors and fast reactors. The thermal reactor uses as its fuel either natural uranium which contains 0·7% of the fissionable U_{235} isotope, or uranium which has been slightly enriched in this isotope by a physical process of isotope separation. Fission of a U_{235} atom produces, on an average, rather more than two neutrons; we require that approximately one of these should be captured by an atom of U_{238} to form U_{239}, which breaks down first to neptunium and then to plutonium, while one other neutron is used to cause fission of a further atom of U_{235} which will release more neutrons to continue the chain of reaction. Since our uranium contains 140 atoms of U_{238} for every atom of fissionable U_{235}, it is obvious that our neutrons are much more likely to strike atoms of U_{238} and to be absorbed than they are to strike atoms of U_{235} and cause fission; if this happens, we stop the further release of neutrons and our chain of reaction dies out.

We must therefore increase the probability that our neutrons will cause fissions of U_{235}; this is done by slowing them down in a moderator to energies which correspond with the temperature of that moderator; they are then called thermal neutrons. A neutron is a valuable article; we need all we have got in the pile, either to form plutonium or to cause further fissions of U_{235}; we cannot afford to waste them by having them uselessly absorbed by other materials. Our problem therefore in a thermal reactor is to select metals for use in the reacting core which will absorb the minimum number of neutrons and yet will meet our other requirements; those most readily available are aluminium and magnesium.

Until now aluminium has been used, particularly for the manufacture of the cans in which the uranium is enclosed. And here we have met the problem, so common in atomic-energy experience, that for the conditions which have to be met the material has defects. Commercial aluminium is most satisfactory for the uses to which it is currently put, but it contains non-metallic inclusions, and to obtain cans free from defects we had to do quite a lot of development in collaboration with the manufacturers. But if we are to use our reactor to generate industrial power we must release the heat at a high temperature; aluminium and magnesium, with their comparatively low softening and melting points, set a limit on the permissible temperature, and there is reason for interest in beryllium and zirconium which also have low neutron-capture cross-sections and which have higher melting points.

The position in the fast reactor is different; here the fuel element contains a high proportion of fissile material; there is no need to maintain a balance between the neutrons which are captured to form plutonium and those which cause further fissions because the quantity of U_{238} in the fuel element is small enough to take care of this. It is therefore unnecessary to slow our neutrons down by using a moderator and we make use of the fast neutrons. They are less likely than thermal neutrons to be captured by metals used in the construction of the reacting core, and in the design of a fast reactor nuclear physics gives us a greater choice of materials. But we run into many other difficulties. Because our fissile material is less diluted than it was in the thermal reactor, the volume and therefore the surface area of our fuel elements is far less, and for this reason we are faced with very high rates of heat transfer and very high temperature gradients. Heat can be removed at the necessary rate only by using liquid-metal coolants, such as mercury, sodium, potassium, lead, bismuth, or alloys of these materials, each having its own advantages and disadvantages. The heat flow from the fuel element is so great as to cause heavy temperature stresses in the container, and in selecting a material for this, the problem is to choose a material which has high heat conductivity, so as to keep these stresses to a minimum, and high creep strength, and which is not corroded either by the liquid-metal coolant or by the fuel, which may be either uranium or plutonium. The problem is further complicated by the fact that the material is subjected to intense neutron bombardment; this bombardment is liable to knock atoms out of their normal position in the crystal-lattice structure, and it may therefore alter the physical and chemical properties of the material. The effect is likely to be least marked on pure metals and on alloys which are simple solid solutions; more complicated alloys may be materially changed.

There is one other metal which I should mention as being of interest in atomic-energy work. Uranium is our only source in nature of fissionable atoms; we can use neutrons from these fissionable atoms to convert part of our uranium into plutonium which is also fissionable. But we can also use the neutrons which are produced from fission of U_{235} to bombard atoms of thorium and so produce another isotope of uranium with atomic mass 233 which is fissionable and which has many desirable properties. It is from this that our interest in the metallurgy of thorium arises.

VI.—CONCLUSION

We have now briefly surveyed the present and future metallurgical requirements of the engineer in the atomic-energy field; neither time nor security considerations would allow us to go into them in much greater detail. We have looked at the shortcomings of the metals of construction available to the more orthodox chemical engineer, and I asked that greater attention should be paid to problems of fabrication when new metals were being developed and proposed that there should be closer collaboration between the metal manufacturers, the plant fabricator, and the user at all stages of development. I also suggested that the range of materials available for the construction of plants that are subject to halide (and particularly fluorine) attack is unsatisfactory.

Even without these additional materials the chemical engineer has at his disposal today far more, far better, and far more expensive materials of construction than were available to him twenty-five years ago, and I would like to conclude with a plea for moderation in their use.

In the old days research chemists in devising new processes paid very close attention to the metallurgical problems which would arise. Some of the solutions provided have great simplicity and cheapness; for instance, in the ammonia soda process, soda ash of high commercial purity can be produced in cast-iron and mild-steel plant, not because these materials are inherently resistant to the chemicals used, but because suitable chemical conditions have been devised for the formation of protective scales. Chemists ought still to be alive to such possibilities and to avoid assuming that the metallurgist can and should solve for him the problems of plant materials to which his processes give rise. I hold that chemical engineers should be discriminating in their choice of materials and avoid "getting the stainless-steel habit"; perhaps lastly I might even venture to suggest that the metal manufacturer should avoid overselling his corrosion-resisting materials; it is neither in his own interest nor in the interest of the chemical industry to do so.

THE USE OF DIAMOND ABRASIVES FOR A UNIVERSAL SYSTEM OF METALLOGRAPHIC POLISHING *

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By L. E. SAMUELS,† B.Met.E., MEMBER

SYNOPSIS

An investigation has been carried out to determine the most efficient and economical methods of using diamond abrasives for metallographic polishing. It has been found that the abrasive is most efficiently used when dispersed in a carrier paste, the polishing rate then being 4–5 times that of dry application, and that it can be used economically only in the finer grades (0–10 μ particle-size range). The polishing rates are then extremely high; even with hand polishing, the rates are frequently higher than that of electrolytic polishing.

The results of the work are applied to the development of a system of metallographic polishing, the costs of abrasive for which are sufficiently low to permit its use for general, and even routine, polishing. To bridge the gap over which diamond abrasives cannot be used economically, a cast wax-abrasive lap has been developed, the characteristics of which are such as utilize the full potential of the subsequent diamond stages. Consideration has also been given to final polishing treatments necessary to give truly scratch-free surfaces. The system finally developed is very rapid and produces a polish of very high quality, with a surface free from objectionable deformation.

I.—INTRODUCTION

THE advantages of the use of diamond abrasives for metallographic polishing have been discussed in a number of recent papers, but in most cases the authors have concentrated upon the preparation of samples that are difficult, or impossible, to prepare by classical methods. Woodside and Blackett [1] described processes for refractory materials, such as molybdenum carbide and arc-cast molybdenum, and for other difficult specimens, such as grey cast iron and specimens of large surface area. Tarasov and Lundberg [2] developed a system for polishing harder materials and specimens containing constituents likely to polish in relief. Perryman's [3] process was also intended primarily for non-ferrous specimens containing constituents or areas differing greatly in polishing characteristics; Davies and Hoare [4] and Greenfield and Davies [5] have applied Perryman's [3] methods to tin-coated materials. Tottle [26] has recently described a modification of Perryman's system.

The above processes are of two general types: first, those in which the diamond abrasive is added to the polishing pad as a dry powder [2–5] and, secondly, those in which it is added by means of an impregnated carrier paste.[1] Impregnated carrier pastes would appear to have considerable advantages both in ease of handling of the diamond dust and in ensuring a more even distribution of the abrasive particles on the polishing pad. All processes so far described are based on rotating pads, and in that employed by Perryman [3] the specimen was also made to rotate and to traverse the pad simultaneously. In most processes a napped cloth is recommended for the polishing pad, although Tarasov and Lundberg [2] recommend silk. Water,[2] carbon tetrachloride,[1] and white spirits (light paraffin oil) [3–5] have been suggested as suitable lubricants. The preliminary polishing stages recommended for general specimens also differ appreciably in detail, but are essentially of the following basic types:

(*a*) Grade 000 or 0000 emery paper,[3–5] immediately preceding the fine diamond stage.

(*b*) A coarse diamond pad, (10–35) μ grade, is used between the abrasive papers and the fine diamond finishing stages.[6]

(*c*) A fixed abrasive lap is used as an intermediate stage between the abrasive papers and the diamond pads. Lead laps, lead-foil laps,[4] diamond grinding wheels, and a vitrified diamond hand hone [2] have been suggested for this purpose.

Although the emphasis has been on the use of diamond abrasives for special purposes, all authors recognize that the process is very rapid and would give very high-quality results with the more usual types of metallographic specimen. The application of diamond abrasives to general metallographic polishing, however, must depend essentially on economic considerations of abrasive costs. The purpose of the present investigation, therefore, was to establish the most efficient methods of using diamond abrasives with a view to developing an economically feasible system of routine polishing.

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Furthermore, the high polishing rates obtained with diamond abrasives suggest that the abrasive would be suitable for hand-polishing processes, and attention was particularly directed to this end.

II—INVESTIGATION OF CHARACTERISTICS OF DIAMOND ABRASIVES

A systematic investigation has been made of the relative merits of the various methods of using the abrasive by determining the polishing rates of pads charged under a range of conditions, by different methods, and with different grades and weights of abrasive. Since this included an assessment of the value of impregnated carrier pastes, it was first necessary to develop a suitable paste of this type.

1. Diamond-Impregnated Carrier Paste

The essential requirement of a carrier paste is that it should be a permanent emulsion prepared from liquids. The abrasive may then be suspended in one of the liquids before mixing and obtained in uniform dispersion throughout the final paste. It is desirable that the paste should be miscible with water to permit water-washing of the specimens if desired, and that it should be readily absorbed by all types of polishing pads. Grodzinski[7] has published the formula of a patented paste, and several types are available commercially.

A simple stearic acid–triethanolamine soap emulsion has been developed and found to be quite satisfactory for metallographic purposes. The components of a batch of paste of convenient size are as follows:

Stearic acid	12·5 g.
Triethanolamine . . .	6·0 ml.
Water	25 ml.
Diamond abrasive * . .	0·5 g.

* Ex Tricfus and Co., Ltd.

The stearic acid is melted and heated to 80°–90° C. The triethanolamine and most of the water are mixed and heated to the same temperature range, a small amount of wetting agent and the diamond dust are added, and the abrasive shaken into a uniform suspension. The molten stearic acid is stirred vigorously with a mechanical stirrer and the abrasive suspension introduced rapidly; the water not used in the original suspension can then be used to wash in any abrasive remaining in the container. The mixture emulsifies immediately, but stirring should be continued until the emulsion cools and thickens. The paste can be conveniently stored in, and dispensed from, tin-lined lead collapsible tubes.

In normal use, a 3–4-in. ribbon of paste (containing 15–20 mg. of diamond) is rubbed well into the polishing pad. Almost any napped polishing cloth, including "Selvyt" cloth, is suitable, but a proprietary synthetic suede cloth ("Microcloth") has been found particularly satisfactory because of its high wear-resistance. For reasons discussed below, kerosene (light paraffin oil) has been adopted as the most suitable lubricant.

2. Polishing-Rate Tests

(a) *Method of Determining Polishing Rate*

The polishing rates (μ/min.) were calculated from the loss of weight of a clean unmounted specimen during a definite polishing period. It is difficult to standardize completely the numerous variables of a polishing process, and no attempt was made to substitute mechanical devices for the normal hand-polishing procedure, particularly as it was found that the condition of the pad was a major variable. Furthermore, it was desired to obtain some indication of the polishing rates to be expected in normal usage. All polishing was done by two experienced operators, polishing alternately for 1-min. periods, and the loss in weight was measured after a total polishing time of 5 min. Except in the first series of tests (see Section II, 2 (b)), the results of 4–6 such polishing periods were averaged. The results are thought to be comparative and reproducible in any one series to within approximately $\pm 5\%$. The synthetic suede polishing cloth and kerosene lubricant were used in all these tests.

(b) *Method of Adding Abrasive*

A comparison was made between two pads, the first charged with diamond dust shaken into a suspension in carbon tetrachloride, and the second charged by means of the impregnated carrier paste. In the former case an effort was made to distribute the abrasive particles as uniformly as possible. Both pads were charged with a total weight of 20 mg. of an (8–20) μ grade of abrasive, and the polishing rate of an annealed copper specimen (hardness 45 D.P.N.) was determined for a succession of 5-min. polishing periods.

The polishing rate of the suspension-charged pad remained substantially constant at 0·8 μ/min. The polishing rate of the paste-charged pad, however, increased rapidly with time until, after approximately 30 minutes' total polishing time, the polishing rate reached a stable value of 4·0 μ/min.—five times that of the suspension-charged pad (see Fig. 1, Plate LXXIII).

The paste-charged pads retained a high polishing rate for long periods. For example, a record was kept of the polishing rate of the paste-charged (4–8) μ-grade pad used for the tests described in Sections II, 2 (c) and II, 2 (d). It was found that the maximum polishing rate (6–7 μ/min.) was maintained on the standard annealed copper specimen for approximately 350 minutes' total polishing time; thereafter, the polishing rate decreased slowly, eventually reaching 2–3 μ/min. after a total pad life of 700 min. The fall in polishing rate is presumably due to drag-off losses of abrasive. Paste-charged pads, therefore, do not have to be regularly cleaned and recharged, as in normal practice. Provided precautions are taken against gross contamination, the pads in the author's laboratory can be used continuously without attention for several weeks, in which period several hundred specimens are polished.

The higher polishing rate of the paste-charged pads is thought to result primarily from a more even distribution of the abrasive particles. When the abrasive is added dry, or as a liquid suspension, groups of abrasive particles settle into that part of the pad which they strike first and do not appear to spread during subsequent use of the pad. With the paste the individual abrasive particles must spread with the paste; the abrasive distribution of a paste-charged pad can therefore be expected to be initially more uniform and to improve as the paste is spread during use.

In any event, it was apparent that the abrasive is most efficiently used when applied by means of a carrier paste, and all further attention was confined to such pastes. The results also indicated that it is necessary to work a polishing pad to a stable condition before any reliable polishing-rate tests can be made.

(c) *Weight of Abrasive on Polishing Pad*

The polishing rates of two samples, an annealed copper (hardness 45 D.P.N.) and a fully heat-treated aluminium–magnesium–chromium alloy (hardness 105 D.P.N.), were determined on pads charged with progressively increasing amounts of paste such that the weight of added abrasive increased from 10 to 120 mg. Three grades of abrasive were tested, namely (0–1) μ, (4–8) μ, and (20–30) μ particle size. The results of these tests are shown in Fig. 2 (Plate LXXIII).

The polishing rates of the (0–1) μ and (4–8) μ grades were, in the range tested, independent of the weight of abrasive. As expected, the polishing rate of the (4–8) μ grade was greater than that of the (0–1) μ grade. At low abrasive weights the polishing rate of the (20–30) μ grade was considerably less than either of the other two grades, but the polishing rate gradually increased as further abrasive was added. These observations can be interpreted by assuming that the maximum polishing rate is attained when the number of abrasive particles per unit area is not less than a certain critical value, and that the critical value is substantially independent of particle size. For the pad area used in these tests, the critical number is apparently contained in less than 10 mg. of the (0–1) μ and (4–8) μ grades, but more than 120 mg. would be required to attain the critical number in the (20–30) μ grade. It is to be noted that the number of particles per unit weight decreases as the third power of particle size.

(d) *Grade of Abrasive*

The results of the tests described in the preceding section indicated that, if the weight of abrasive added to the polishing pad was to be kept within reasonably economic limits (say 10–40 mg.), only the finer grades of abrasive could be used effectively. Further quantitative information as to the most suitable grades was obtained from comparative polishing-rate tests made with the annealed copper and the heat-treated aluminium alloy specimens referred to previously, and with a B.S.970 : En 9 steel heat-treated to a hardness of 250 D.P.N. The following particle-size grades were tested: (0–$\frac{1}{2}$) μ, (0–1) μ, (4–8) μ, (20–30) μ, and 300 mesh. With all grades 40 mg. of abrasive was added to the polishing pad by means of an impregnated paste. In Fig. 3 (Plate LXXIII) the polishing rates have been plotted against the maximum particle size of each abrasive grade.

It will be noted that the copper and aluminium alloy showed a maximum polishing rate at approximately the (4–8) μ grade, and steel a small maximum at the (0–1) μ grade. This information indicates that only grades in the (0–10) μ particle-size range can be used effectively in general polishing. Since the polishing rates of the finer grades of diamond abrasive do not differ greatly, it is for consideration whether the diamond abrasive pads should not be confined to one of the finer grades, say (0–1) μ. Two diamond stages are considered desirable, however, even if only to ensure that the final pad is protected from contamination. For the first or roughing pad, a coarser grade (e.g. (4–8) μ) is the most suitable in order to take advantage of the maximum polishing rate that some metals show in this range. As for the second or finishing pad, it has been found that the quality of finish obtained with (0–$\frac{1}{2}$) μ and (0–1) μ grades of abrasive is substantially the same. In both cases the maximum scratch depth appears to be largely determined by the type and condition of the polishing cloth, and it is thought, therefore, that no advantage would result from the use of even finer grades of abrasive. The (0–1) μ grade is clearly to be preferred for the finishing pad because of its higher polishing rate.

The results so far discussed were obtained with hand polishing; some consideration was also given to the use of mechanical wheels for the diamond abrasive pads. A number of tests indicated that, under otherwise similar conditions, the polishing rate of a mechanical pad rotating at 250 r.p.m. was of the order of 2–3 times that of a stationary hand pad. Although hand-operated (4–8) μ-grade diamond pads can readily cope with the surfaces produced on the wax lap described in Section III, the use of a mechanical wheel for this roughing pad greatly speeds the operation and ensures a more thorough treatment on the pad.

(e) *Lubricant*

No systematic survey has been made of polishing lubricants. It has been repeatedly found, however, that the pads must be maintained in a reasonably moist condition in order to achieve the maximum polishing rate. For example, the polishing rate of the aluminium alloy on a (4–8) μ-grade pad falls as low as 2–3 μ/min. when the pad is excessively dry; the addition of a small amount of kerosene restores the polishing rate to the normal value of 7–8 μ/min. Volatile lubricants, such as carbon tetrachloride, are therefore considered unsuitable, since it is virtually impossible to maintain the polishing pad in a satisfactorily moist condition.

(f) *Material of Specimen*

It will be noted that the polishing rate of the aluminium-alloy specimen in the tests reported in Section II, 2 (d) was consistently higher than that of the copper specimen, although the hardness was over twice that of the copper. On the other hand, the polishing rate of the steel specimen of approximately twice the hardness again was only a fraction of that of the aluminium alloy. These features are in conformity with the theory advanced by Bowden and Hughes [8] that the process of polishing is greatly influenced by the relative melting point of the abrasive and the material being polished and that the relative hardness is comparatively unimportant. To obtain further information on this point, polishing-rate tests were carried out on a series of samples representing a range of melting points and hardness values. The specimens were tested in groups of similar melting point, and each group was compared with the annealed copper specimen, so that, by correcting the results to a polishing rate for copper of 7·0 μ/min., the effect of variation in the condition of the polishing pad could be eliminated. The results are set out in Table I, the materials being arranged in order of increasing melting point.

Table I.—*Polishing Rates of Metals in Relation to Melting Point.*

Specimen				Polishing Rate, μ/min.		
Alloy	Condition	Melting Range, °C.	Hardness, D.P.N.	Alumina Wax Lap, (10–30) μ Grade	Diamond Pad, (4–8) μ Grade	Electrolytic Polishing
Bismuth-tin-lead eutectic	As-Cast	95	11	50	16·8	...
Tin	As-Cast	231	9	30	5·9	5[9]
Cadmium	Annealed	321	22	25	8·8	...
Lead	Annealed	327	5	...	4·7	...
Zinc–0·25% cadmium alloy	Annealed	420	50	17	7·5	0·2[9]
D.T.D. 289 cast magnesium alloy	Heat-Treated	455–610	65	16	10·5	...
Aluminium alloy	Heat-Treated	580–650	105	15	8·4	0·25[10]
Aluminium alloy	Annealed	580–610	40	...	5·3	...
Aluminium	Annealed	660	25	22	5·7	3[10]
Brass (40% Zn)	Extruded	900–905	155	17	9·7	...
Brass (30% Zn)	Annealed	915–955	95	13	9·4	...
Silver	Annealed	960	30	...	7·6	...
Copper (Tough-pitch)	Annealed	1083	45	10	7·0	0·9[9]
Austenitic 18 : 8 stainless steel	Annealed	1400–1425	170	5	2·2	0·6[9]
B.S. 970 : En 9 steel	Heat-Treated	1425–1450	250	...	0·8	0·75[11]
B.S. 970 : En 9 steel	As-Hardened	1425–1450	800	...	0·7	...
B.S. 970 : En 2A steel	Annealed	1450–1500	150	8	0·8	1·0[9]
Nickel	Annealed	1452	125	8	0·2	60[9]
Titanium	Annealed	1725	275	...	0·1	...

The results suggest there is a general relationship between the polishing rate of the diamond abrasive pad and melting point of the specimen being polished, although, when specific comparisons are made, there are a number of departures from a strict relationship. It would appear, therefore, that the polishing rate is partly dependent on features other than the melting point. Futher support for this conclusion is found in the observation that the individual grains of many metals polish to slightly but noticeably different levels on these pads; the grain relief shown under phase-contrast illumination in the super-purity aluminium specimen of Fig. 9 (Plate LXXIV) entirely originated on the diamond abrasive pads. The effect is most noticeable in coarse-grained and non-cubic metals. It is suggested, therefore, that polishing rate is an anisotropic property. There is clearly no relationship between polishing rate and hardness.

In the last column of Table I previously published values of the electrolytic polishing rate of similar metals are also given. It will be noted that the polishing rate of metals of low and intermediate melting point on the diamond pad is usually at least of the same order as, and in most cases considerably greater than, the rate of the standard electrolytic polishing process. If, as suggested, polishing rate is related inversely to melting point, the electrolytic polishing rate of high-melting-point metals can be expected to exceed that of diamond polishing, since the former is unrelated to the melting point of the specimen.

3. Summary of Optimum Conditions of Use of Diamond Abrasives

The important features of the above experiments, so far as the practical use of diamond abrasives in general metallographic polishing is concerned, may be summarized as follows:

(*a*) Diamond abrasives are most efficiently used when dispersed in a carrier paste which ensures even distribution of the abrasive on the polishing pad.

(*b*) If the amount of abrasive added to the polishing pad is to be kept within reasonably economic limits, abrasive grades must be restricted to the (0–10) μ particle-size range. The most suitable abrasive grades in this range are a grade of approximately (5–10) μ particle size for a roughing pad and a (0–1) μ grade for a finishing pad. An addition of 10–20 mg. of these grades is sufficient to attain the maximum polishing rate.

(*c*) Polishing pads charged as above do not require constant cleaning and recharging, as in normal practice, but can be used continuously over considerable periods. Abrasive costs are, therefore, nominal.

(*d*) The polishing rates even of hand-operated diamond abrasive pads used in this manner are very high. The polishing rate would appear to depend partly on the melting point of the specimen and partly on other factors, as yet unidentified.

III.—DEVELOPMENT OF A PRACTICAL POLISHING SYSTEM

The most serious limitation of diamond abrasives for general metallographic polishing is the restriction, on economic grounds, to relatively fine grades. There is also a limitation to the fineness of finish obtainable both because of the limited range of grades at present available and because of the influence of the polishing cloth itself on the finish. It would therefore seem that, although diamond abrasives are eminently suitable as the basis of a polishing procedure, it may be necessary to develop auxiliary stages to cover the ranges over which diamond abrasives are not suitable. The development of such stages is considered in this section of the paper.

1. Preliminary Preparation

(a) *Requirements*

In spite of their high polishing rates, the coarsest practicable diamond abrasive pads cannot remove in a conveniently short period of hand polishing the scratches produced by the usual preliminary abrasive papers; even more difficulty would be experienced in ensuring removal of the deformed layer associated with such surfaces. Although it would be possible to follow the abrasive papers with a relatively long machine operation, experience has shown that it is much more satisfactory to introduce an intermediate stage of polishing, using commoner abrasives, such as alumina, between the abrasive papers and diamond pads. These remarks apply particularly to silicon carbide waterproof papers, which are available only in relatively coarse grades, but which are nevertheless preferred to the finer emery papers as being more uniform and consistent in use.

It was first recognized by Vilella,[12, 13] Morrogh,[14] and Amberg [15] that an intermediate stage of the type contemplated must be of the fixed-abrasive lap type if non-metallic inclusions are to be retained, edges preserved, and excessive relief between constituents avoided. A lap of this type is essential to ensure that the full potential of the diamond pads can be utilized.

(b) *Development of Intermediate Lap*

A large number of different types of fixed-abrasive laps has been developed in recent years. The earlier types consisted of a surfaced block of cast iron, lead, or paraffin wax which was charged by rolling abrasive into the surface. Laps of this type, which have been described by Amberg,[15] Ellinger and Acken,[16] Jarrett,[17] and Ferguson,[18] are satisfactory only when freshly surfaced and, consequently, are uncertain in use and require constant attention. To overcome these difficulties, Cohen and Maker [19] developed a process in which the abrasive was rubbed into the surface of a strip of lead foil. The foil can be used for a short period only, and must then be discarded. Vilella [13] has described a wax-impregnated cloth lap which is charged by rubbing abrasive into the surface; this lap can be easily cleaned and recharged. Similar pads have been in use in the author's laboratory for a number of years, and are known to operate satisfactorily when correctly prepared. The charging of such laps is, however, very critical and somewhat uncertain, and they require frequent cleaning and recharging.

An entirely different type of lap has been suggested by Dowdell and Wahll [20] and by Dauber,[21] in which a mixture of molten wax and abrasive is cast on to a polishing head to form a wax-abrasive slab. This type is considered to be the most satisfactory of all tried, but the type of wax and the proportions of wax to abrasive previously suggested [20, 21] have been found unsuitable for hand polishing. For this purpose, the wax must have a high melting point, so that it will not " pick up " during polishing, but it must also be sufficiently ductile to withstand the casting contraction stresses. A compounded wax with a melting range of 70°–85° C. has been found to be suitable. The abrasive : wax ratio is also somewhat critical. Although, in general, this ratio should be as high as possible, an excessive abrasive concentration results in a hard lap which does not cut freely and evenly. Once the optimum ratio has been established for a particular wax and abrasive grade, the laps are quite reproducible. For the wax and abrasive used in the present work, the optimum concentration is 150 g. abrasive : 50 g. wax; this charge size being suitable for a 20-cm.-dia. lap.

The method of preparing a lap is as follows: A network of saw cuts is made in the face of a standard metal polishing head, a paper dam is fixed around the periphery of the head, and the head is mildly preheated. The wax is melted and the abrasive added and stirred in well to a uniform mixture. The mixture is then cast on to the head and allowed to cool slowly in still air. Finally, the working face is dressed flat in a lathe.

It has been found absolutely essential to use an elutriated grade of abrasive in the lap. If the finer particles of the usual bulk abrasive are not removed, the lap clogs very rapidly and does not cut freely. The elimination of the very coarse particles is also desirable in order to obtain the maximum uniformity of finish. The most suitable grade of alumina for a lap to follow silicon carbide waterproof papers is an elutriated "20 sec." grade ((10–30) μ nominal particle size), as prepared by the method described by Rodda.[22]

(c) *Operation of the Intermediate Lap*

The laps must be used dry, since lubricants, even water, loosen the surface abrasive and the fixed-abrasive characteristic is lost. The working surface consequently clogs with debris and must be cleaned after being used for several specimens. The cleaning is normally carried out simply by rubbing the surface with cotton wool moistened with alcohol. This treatment also removes the superficial layer of worn abrasive, and the cleaned lap is ready for immediate use in the optimum cutting condition. After extensive use the simple treatment may fail to clean the surface effectively; in addition, it may become excessively undulating. The surface should then be flooded with benzol, the surface layers softened by rubbing with the fingers and scraped off with a straight-edge, thus forming a new flat working surface. The main advantage of the lap is that it can be kept in the optimum cutting condition by these simple cleaning treatments, without the need for re-charging.

This lap is intended for hand operation while stationary; it is not suitable for use as a rotating wheel. The cutting rate by hand, however, is very high, being of the order of that attained with abrasive papers. The lap can readily deal with the finish produced on 400-mesh silicon carbide waterproof paper. For comparison with the diamond pads, the results of a number of polishing-rate tests are included in Table I; these rates were determined from the loss of weight for several 15-sec. polishing periods on freshly cleaned laps, and thus represent the maximum polishing rates. It will be noted that this series is in good agreement with Bowden and Hughes's theory of polishing.[8] The finish produced by the lap is fine, uniform, and clean, and is a marked improvement on that produced by the finest emery papers (see Fig. 5, Plate LXXIV); the lap scratches can usually be removed in less than 1 min. of hand polishing on a (4–8) μ-grade diamond abrasive pad. This, combined with the retention of inclusions and the high degree of flatness of the surface produced, makes the lap an ideal preparation for the subsequent diamond stages.

On the basis of the correlation between polishing rate and the relative melting points of abrasive and specimen, an alumina-abrasive lap can be expected to have some limitations. It has been found in practice that materials with melting points much above 1500° C., such as titanium, molybdenum, and tungsten, cannot be satisfactorily polished with this lap; for such materials an intermediate stage using an abrasive of higher melting point is probably necessary. The present lap is intended, therefore, only for the more usual lower-melting-point metals.

2. Final Polishing

The finish produced by the finer diamond abrasive pads has been reported to be satisfactory as a final polish for many ferrous materials.[2, 6] As noted by Perryman,[3] however, the finest diamond finish on the usual non-ferrous materials shows noticeable scratches, and a further final polishing stage is desirable for photomicrography. A Selvyt cloth pad charged with a fine grade of alumina is usually recommended for this purpose.[3–5] The author's preference is for calcined magnesium oxide for the final polishing stage; it has been found, moreover, that this treatment also results in an appreciable improvement in the finish of many ferrous materials.

It is doubtful, however, whether surfaces finished by either of these methods are truly scratch-free. The as-polished surface may appear to be satisfactory when examined in normal bright-field illumination, but more critical examination under oblique illumination or, more particularly, under phase-contrast illumination usually reveals a maze of fine background scratches. For example, the specimen of as-cast super-purity aluminium shown under phase-contrast illumination in Fig. 8 (Plate LXXIV) was finished by the usual technique on a hand-operated magnesium oxide pad and appeared to be substantially free from scratches under bright-field illumination, as does the matrix of Fig. 7 (Plate LXXIV). Whether or not an etched specimen appears to be scratched depends on the characteristics of both the etching reagent and the material of the specimen. When the etch is light and the matrix still highly reflecting, the surface will appear unscratched in bright-field illumination; if, however, the etch removes little material and is of the colouring or staining type, the scratches may be made quite obvious. A heavy etch frequently removes completely the scratched surface layer.

It is thought that the major scratches produced on a magnesium-oxide pad result from the polishing cloth rather than the abrasive. On this basis, the following final polishing technique was evolved. The magnesium oxide is mixed into a very thick paste with water and pushed through a fine-mesh sieve on to the polishing pad; the sieving treatment serves both to remove extraneous grit and to break up the paste into a smooth thick cream. The paste is spread over the surface of the polishing cloth and the pad operated with a light pressure so that the specimen skids over a packed bed of paste without touching the nap of the cloth. This technique is capable of producing surfaces which, even in the softest low-melting-point metals, appear scratch-free under phase-contrast illumination. The super-purity aluminium specimen finished by the "skidding" technique is shown in Fig. 9 (Plate LXXIV) for comparison with Fig. 8 (Plate LXXIV); some shallow undulations are present in the surface, apparently because the major grooves produced on the (0–1) μ-grade diamond pad are partly reproduced even after extensive periods of polishing, but there is no evidence of scratches as such. The specimens of tin and zinc shown in Figs. 10 and 11 (Plate LXXIV) were prepared by this method, and apparently are also completely free from scratches.

The polishing rate of such a pad is of the same order as that of normal magnesium-oxide pads and, if the pad is operated on a low-speed mechanical wheel, the polishing is quite adequate for finishing the lower-melting-point non-ferrous metals; this applies to metals with a melting point at least as high as that of aluminium. It is usually even possible to remove (0–1) μ diamond scratches on higher-melting-point metals in a reasonable time (less than 5 min.), but it may not be practicable to remove completely the deformation beneath these scratches. In such cases, it has been found necessary to increase the polishing rate by the use of a "polish-attack" technique. It is possible to control such a technique to a high degree in the present process by using a definite concentration of the required etching reagent in the solution used for mixing the magnesium-oxide paste.

Among the commoner alloys only copper and copper alloys, in the author's experience, require a polish-attack technique. The usual etching reagents for these alloys seem to be particularly sensitive in revealing the deformation zones associated with the previous system of diamond scratches. Ammonium persulphate has been found to be the most suitable reagent, the concentration required in the paste solution differing somewhat for the various alloys and final etching methods; a concentration of the order of 10–15 g./l. is suitable for copper and 15–30 g./l. for alloys such as high-zinc α-brasses.

One further precaution is necessary when this technique is used with metals, such as aluminium and zinc, which are strongly electropositive with respect to the copper alloys normally used for polishing heads. Severe etch-pitting may develop in these cases as a result of electrolytic effects between the specimen and the polishing head. The effect can be completely prevented, however, by fitting a thin disc of an insulating material between the polishing cloth and head. This precaution was taken during the final polishing of the aluminium and zinc specimens shown in Figs. 9 and 11 (Plate LXXIV), respectively.

IV.—SUMMARY AND DISCUSSION

The main results of the investigation have been, first, that the methods of obtaining the maximum polishing rates with diamond abrasives have been established and, secondly, that it has been possible to evolve a system of metallographic polishing based on diamond abrasives, the abrasive costs for which are so low as to permit its use for general, and even routine, polishing. The stages of this system are summarized below:

(1) *Abrasive Papers.*

Silicon carbide waterproof papers used with flowing water lubricant.

- (*a*) 220-mesh grade.
- (*b*) 400-mesh grade.

(2) *Intermediate Lap.*

Cast wax-abrasive lap charged with (10–30) μ-grade elutriated alumina. Used dry; hand-operated.

(3) *Diamond Abrasive Pads.*

Impregnated paste on napped cloth. Kerosene lubricant; hand or preferably machine operation.

- (*a*) Roughing stage; (4–8) μ grade (optional).
- (*b*) Finishing stage; (0–1) μ grade.

(4) *Finish Polishing.*

Magnesium oxide used on Selyvt cloth in one of the following manners. Hand or machine operation.

- (*a*) As a slurry.
- (*b*) As a thick paste with water, so that the specimen skids over a bed of paste.
- (*c*) As in (*b*); paste made with a dilute solution of a suitable etching reagent.

The treatment of all the usual specimens is the same up to the finish-polishing stage; the most suitable method of finish polishing for a particular type of specimen must be determined by trial. The major portion of the polishing takes place on diamond pads, so that the overall process is extremely rapid. The average preparation time to the end of the diamond stages never exceeds 5 min.; the time required for final polishing depends on the specimen, but, even with the skidding techniques, it rarely exceeds 5 min., and is usually of the order of 1–2 min.

As a typical example, the surface condition of an unmodified aluminium–13% silicon alloy at the major stages of the process is illustrated in Figs. 4–7 (Plate LXXIV); the specimen was finished by Method (4*a*) above. This series illustrates particularly the uniform clean finish, free from relief, obtained on the wax lap, the improvement resulting from the magnesium-oxide final polish, and the high quality of the final result.

Although the rapidity of the process is an important advantage, an even more important feature associated with the use of diamond abrasives is the marked improvement in the quality of the polish produced. This results from low relief between constituents and dissimilar zones, the high degree of flatness of the surface, the sharp retention of edges, and the complete retention of non-metallic inclusions —features which have been fully discussed by previous authors.[1, 3, 4, 5] Furthermore, considerably less skill is required on the part of the operator to obtain such results. It is considered that in these respects the use of diamond abrasives represents a major advance in metallographic technique.

It has also been noted that surfaces produced by this process are invariably free from polishing artifacts, i.e. deformation of the surface during polishing which is detectable during the microscopic examination. Two metals that are notoriously difficult to prepare to a truly representative surface by mechanical polishing have been selected as examples, namely, tin and zinc. Mechanically polished specimens of tin frequently contain artifacts in the form of a fine-grained recrystallized layer or mechanical twins,[5, 23] which, in previous polishing systems, required removal by a laborious series of alternate polishing and etching treatments.[5] Tin specimens prepared by the present method are invariably free from such defects on the first etch (see Fig. 10, Plate LXXIV); if present, the depth of the deformed layer on the as-polished surface must therefore be less than the thickness removed by etching.

Specimens of zinc when polished by the present method show no evidence of the mechanical twins that are so characteristic of specimens prepared by earlier polishing methods.[23, 24] The specimen shown in Fig. 11 (Plate LXXIV) was photographed in the unetched condition in polarized light; in this case, therefore, the as-polished surface itself is established to be true. The high polarization contrast in Fig. 11 is also of interest, as it has previously been suggested that it is not possible to obtain optimum grain contrast under polarized light on mechanically polished surfaces of zinc.[23, 25] It is to be emphasized that both these specimens were polished by the standard method and were not etched between stages; the tin specimen was given only the one final etch after polishing, whereas the zinc specimen was not etched at any stage.

It has been noted that the only precaution necessary to produce artifact-free surfaces in such sensitive specimens is to ensure that the specimen is given a thorough treatment on the (4–8) μ-grade diamond pad. It is thought that the artifacts normally found on mechanically polished surfaces largely originate in the early abrasion stages. The high cutting rate of the diamond abrasive pad apparently effects the complete removal of the deformed layer produced in these stages without itself seriously deforming the surface. It is clear, however, that this aspect requires further investigation, and such an investigation is in hand.

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THE SOLUBILITY OF INDIUM IN COPPER *

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SYNOPSIS

The X-ray measurements that have been made to determine the solubility of indium in copper, and thus the α-phase boundary of the copper–indium system over the range of temperature from 710° to 470° C., are recorded. The maximum solubility of indium in copper is found to be 10·9$_0$ at.-% at 575° C.; the solubility at the peritectic temperature (710° C.) is 10·0$_5$ and at 500° C., 8·1$_0$ at.-%.

I.—INTRODUCTION

In a recent paper by Owen and Morris [1] some unpublished results on the solubility of indium in copper in the temperature range between about 470° and 710° C., were quoted in support of the view that X-ray measurements are capable of yielding results which fix the positions of phase boundaries in equilibrium diagrams of alloy systems as accurately at high as at low temperatures. The purpose of the present paper is to place on record the measurements made in arriving at the solubility values quoted in the earlier paper.

The technique developed to produce ingots of alloys in a state of equilibrium, homogeneous and of known composition, has been described in some detail in a paper by the present authors on the solubility of certain metals in gold,[2] and has been further explained and amplified in the discussion [3] on the paper by Owen and Morris. There is therefore no need to go into further detail here, and only the briefest account of essential experimental procedure will be given.

II.—EXPERIMENTAL TECHNIQUE

The alloys were prepared from copper and indium of purities 99·95 and 99·98%, respectively, in ½-g. ingots, by fusion in quartz tubes. A preliminary investigation was carried out to find how quenching in oil compared with quenching in water as far as the production of homogeneous alloys was concerned. It was definitely demonstrated that for these alloys quenching in water was by far the more satisfactory method. Several alloys were prepared and quenched in water from 20°–50° C. above the liquidus temperatures. Weight losses in the alloys after annealing amounted to less than 1 part in 3000 in five of the alloys, and to 1 part in 2000 in two. Before examination, alloys containing 2·55, 5·08, 9·12, and 10·84 at.-% indium were lump-annealed at 600° C. for 19 days and alloys containing 1·38, 3·78, and 5·83 at.-% indium at 700° C. for 14 days.

III.—RELATION BETWEEN LATTICE PARAMETER AND COMPOSITION

The lattice-parameter measurements showed that the alloys, prepared in the form of small cylinders, were fairly free from segregation. The figures given in Table I show the variation in lattice parameter observed in samples taken from different parts of the ingot. Samples A, C, and B were taken respectively from the outside surface, central longitudinal section, and a longitudinal section at a distance about half the radius of the ingot from the central section; samples D and E were taken from the two ends of the ingot. It will be observed that the greatest difference between the lattice-parameter values is found in the end samples of alloy containing 1·38 at.-% indium; this amounts to 0·0008 kX, which corresponds to less than 0·1 at.-% in composition.

Table I.—*α-Phase Lattice Parameters of Copper–Indium Alloys at 18° C.*

Indium Content, at.-%	Specimen and Lattice Parameter, kX					Estimated Lattice Parameter, kX
	A	B	C	D	E	
1·38	3·6204	3·6202$_5$	3·6201	3·6198	3·6206	3·6201
2·55	3·6316	3·6311	3·6310	3·6313	3·6310	3·6310
3·78	3·6428	3·6426$_5$	3·6427$_5$	3·6432	3·6432	3·6427$_5$
5·08	3·6546	3·6547	3·6550	3·6547	3·6550	3·6550
5·83	3·6615	3·6618	3·6620	3·6620	3·6623	3·6620
9·12	3·6927	3·6926	3·6922$_5$	3·6917	3·6922$_5$	3·6922$_5$
10·84	3·7074	3·7072	3·7072$_4$	3·7069	3·7072	3·7072$_4$

The values shown in the last column, which are the same as those in column C, were considered to be the most representative and were used to draw the curve showing the relation between lattice parameter and composition across the α-phase region. These are included in Fig. 1 with those of Weibke and Eggers,[4] who were the first to investigate this system of alloys. On the whole there is fair agreement between the two results. Most of the points obtained in the present

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investigation lie within 0·0002 kX of the curve drawn in the diagram.

IV.—α-PHASE BOUNDARY BETWEEN 710° AND 470° C.

Several ingots of composition within the adjacent two-phase regions were prepared, from which powder specimens were taken, annealed at different temperatures, and quenched in cold water. Nickel radiation was used to obtain the photographs. Table II contains the results of measurements made on samples of alloys annealed between 704° and 469° C., alloys which were considered to have reached the equilibrium state. The temperature range is that shown in Fig. 6 (p. 160) of the paper by Owen and Morris.[1]

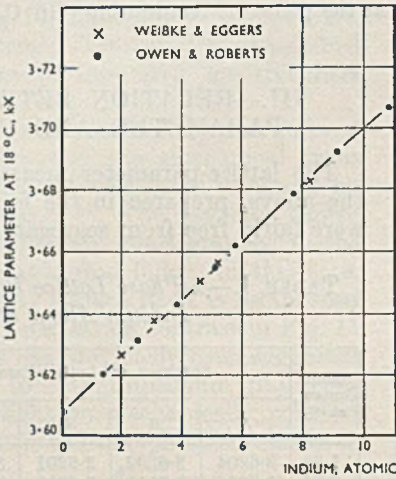

FIG. 1.—Lattice Parameter/Composition Curve for Copper–Indium α-Phase Alloys.

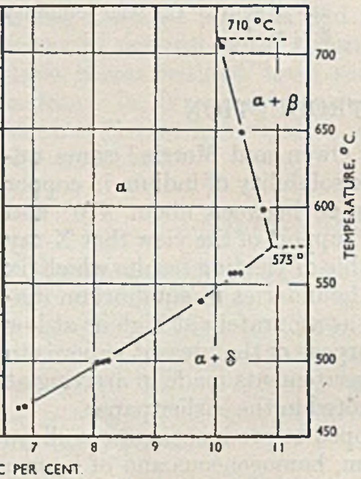

FIG. 2.—α-Phase Boundary of Copper–Indium Alloy System Between 715° and 470° C.

TABLE II.—*Data to Determine the α-Phase Boundary.*

Annealing Treatment		Lattice Parameters at 18° C., kX				Boundary Point, at.-% In
		Indium Content, at.-%				
Time, hr.	Temp., °C.	7·66	9·12	10·84	11·89	
0·25	704	...	...	$3{\cdot}7011_5$	3·7011	$10{\cdot}1_1$
1	649	...	...	3·7040	$3{\cdot}7041_5$	$10{\cdot}4_3$
0·75	599	...	...	3·7069		$10{\cdot}7_5$
2·5	599	...	...	$3{\cdot}7069_5$		
2	598	...	...	...	3·7070	$10{\cdot}7_6$
2	557	...	...	3·7036	...	$10{\cdot}3_8$
2	557	...	...	...	3·7025	$10{\cdot}2_6$
2·5	558	...	...	...	3·7030	$10{\cdot}3_2$
2·5	538	...	...	...	3·6979	$9{\cdot}7_5$
2	500	...	$3{\cdot}6837_5$	...	...	$8{\cdot}2_0$
4	499	...	...	3·6830	...	$8{\cdot}1_2$
4	499	...	...	...	3·6825	$8{\cdot}0_6$
5	469	$3{\cdot}6705_5$	...	...	...	$6{\cdot}7_7$
5	469	...	3·6713	3·6713	...	$6{\cdot}8_5$

The results are shown in Fig. 2. The α-phase boundary shows a sharp discontinuity at 575° C. corresponding to the eutectoid transformation β → (α + δ); maximum solution, 10·9 at.-% indium, occurs at this temperature. The temperature of the reaction agrees closely with the value 574° C. found by Weibke and Eggers.

Since most of the experimental values of the α-phase lattice parameters are within 0·0002 kX of the parameter/composition curve, there is justification in assuming the lattice parameter of the α-phase alloys read from the parameter/composition curve to be correct to within ±0·0002 kX.

At temperatures above 575° C. it is considered that the solubility values are correct to within ±0·05 at.-% after allowing ±0·0002 kX for errors in the two-phase parameter values. Below 575° C. errors are introduced not so much through errors in parameter, as through errors in temperature measurement, since an error of 1° C. is equivalent to 0·04 at.-% in composition.

The accuracy of the final figures given for the position of the boundary at temperatures below 575° C., can be placed at about ±0·2 at.-%, which includes errors in pure and in mixed lattice-parameter values.

It will be necessary to conduct a further investigation with even purer metals to check this boundary, especially since at temperatures below 575° C. the solubility is found to exceed that determined by metallurgical methods, which is contrary to the usual findings. It would be desirable also to extend the boundary to still lower temperatures. Such an investigation is now nearing completion, so that the results will shortly be available.

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PRIMING PAINTS FOR LIGHT ALLOYS*

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SYNOPSIS

The results obtained from the exposure of light alloy specimens for 6 months in rural, industrial, and marine atmospheres, with various primers under aluminium top coats, have been published previously, and the present paper gives the further information obtained on continuation of the tests for 3½ years.

The results confirmed earlier indications. Zinc chromate and zinc tetroxychromate primers were superior in protective value to iron oxide primer, although the latter provided satisfactory protection, especially to aluminium in the less severe environments. Red lead primer was definitely harmful on light alloys, and especially on magnesium in corrosive environments. In these tests both types of chromate primer were found to be slightly preferable to red lead on mild steel. Zinc chromate and zinc tetroxychromate pigments are preferred for primers for composite structures of steel and light alloys in severe exposure conditions, while iron oxide is adequate for milder conditions.

I.—INTRODUCTION

This paper records the final results obtained, after 3½ years, from tests comparing the protective behaviour of various paint primers on aluminium and magnesium alloys and mild steel in rural, industrial, and marine atmospheres. The tests were designed primarily to determine the relative protection given by certain priming pigments when applied to light alloys, and the comparison was simplified by the use of single pigments. The four pigments selected were zinc chromate, zinc tetroxychromate, iron oxide, and red lead. Priming coats based on these four pigments, in a phenolic–linseed stand oil–oiticica oil varnish medium, were tested both alone and under two coats of aluminium in (i) medium-length linseed-oil-modified alkyd resin (D.T.D. 260A) and (ii) nitrocellulose non-drying alkyd resin media (D.T.D. 63A). Full details of the design of the tests and the results obtained after exposure for about 6 months have been given in a previous paper.[1] The chemical analyses of the metals tested are reproduced in Table I.

No attempt to clean or maintain the test panels has been made during the tests. Representative singleton withdrawals of all the paint–metal systems have been made periodically for detailed laboratory examination, and all specimens have been examined *in situ* at intervals of approximately 6 months. The specimens have been examined by the methods previously used.[1]

II.—RESULTS OF EXPOSURE TESTS

The results obtained at the completion of the tests, after exposure for up to 3½ years, are summarized below and in Tables II–VI. The results for the whole investigation are summarized in Table VI by quoting the approximate times required for the various panels to suffer slight corrosion. In assessing these values, all types of corrosion, including scratch-line effects, have been considered. Typical photographs of specimens withdrawn from the marine site are given in Figs. 1–3 (Plates LXXV–LXXVII).

The severity of the exposure conditions decreased in the order: marine (Dungeness, Kent), accelerated

Table I.—*Analysis of Metals Included in Tests.*

Material	C, %	Si, %	Fe, %	Cu, %	Mn, %	Zn, %	Mg, %	Al, %	S, %	P, %
Aluminium	...	0·24	0·31	0·008	0·008	0·010	...	*	...	...
B.S.S. 5L3	...	0·54	0·40	3·95	0·59	0·018	0·60	*	...	...
D.T.D. 120A, high-purity base .	...	0·006	0·0005	0·006	0·42	0·93	*	6·10	...	...
D.T.D. 118, normal-purity base .	...	0·005	0·021	0·004	1·54	0·005	*	...	...	...
D.T.D. 59A, normal-purity base .	...	0·010	0·022	0·002	0·28	0·59	*	7·70	...	...
D.T.D. 59A, high-purity base .	...	0·010	0·0010	0·002	0·18	0·58	*	7·83	...	...
Steel	0·16	<0·01	*	...	0·26	...	...	...	0·040	0·027

* Remainder.

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TABLE II.—*Summary of Results for Dungeness (Marine Tests) Specimens after 3½ Years' Exposure.*

Paint System	Main Paint Failures and Corrosion						
	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 120A, high purity	D.T.D. 118, normal purity	D.T.D. 59A, high purity	D.T.D. 59A, normal purity	Steel
Zinc chromate primer (*c*).	Trace of flaking.	Trace of flaking.	Very slight flaking, very slight local corrosion, and trace of corrosion at scratch.	*Considerable local corrosion, considerable pitting, and very slight corrosion at scratch.	Slight flaking, very slight local corrosion, very slight pitting, and trace of corrosion at scratch.	*Moderate flaking, moderate local corrosion, moderate pitting, and moderate corrosion at scratch.	*Moderate local corrosion and very slight corrosion at scratch.
Zinc chromate primer + D.T.D. 63A top coat (*g*).	Slight flaking.	Moderate flaking, slight local corrosion, very slight pitting, and trace of corrosion at scratch.	Considerable flaking, very slight local corrosion, very slight pitting, and trace of corrosion at scratch.	Severe flaking, considerable local corrosion, considerable pitting, and very slight corrosion at scratch.	Considerable flaking, very slight local corrosion, and trace of pitting, and trace of corrosion at scratch.	Moderate flaking, moderate local corrosion, moderate pitting, and considerable corrosion at scratch.	*Moderate local corrosion and slight corrosion at scratch.
Zinc chromate primer + D.T.D. 260A top coat (*l*).	Very slight flaking.	Moderate flaking.	Severe flaking, trace of pitting, and trace of corrosion at scratch.	Very severe flaking, slight local corrosion, moderate pitting, and trace of corrosion at scratch.	Trace of flaking, trace of pitting, and trace of corrosion at scratch.	Moderate flaking, moderate local corrosion, considerable pitting, and considerable corrosion at scratch.	*Slight local corrosion and slight corrosion at scratch.
Zinc tetroxychromate primer (*d*).	Moderate erosion.	Slight erosion and trace of pitting.	Very slight erosion and flaking, very slight local corrosion, trace of pitting, very slight corrosion at scratch.	Slight erosion and flaking, considerable local corrosion, moderate pitting, and slight corrosion at scratch.	Slight erosion and flaking, slight local corrosion, very slight pitting, and slight corrosion at scratch.	*Severe erosion and flaking, considerable local corrosion, and severe corrosion at scratch.	*Moderate local corrosion and moderate corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*).	Slight flaking.	Slight flaking.	Moderate flaking, very slight local corrosion, trace of pitting, and very slight corrosion at scratch.	Considerable flaking, moderate local corrosion, moderate pitting, and very slight corrosion at scratch.	Moderate flaking, very slight local corrosion, slight pitting, and very slight corrosion at scratch.	Very slight flaking, considerable local corrosion, considerable pitting, and severe corrosion at scratch.	*Moderate local corrosion and very severe corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*).	Very slight flaking.	Considerable flaking.	Very severe flaking, very slight local corrosion, trace of pitting, and trace of corrosion at scratch.	Severe flaking, moderate local corrosion, moderate pitting, and very slight corrosion at scratch.	Slight flaking, very slight local corrosion, trace of pitting, and trace of corrosion at scratch.	Very slight flaking, considerable local corrosion, considerable pitting, and considerable corrosion at scratch.	*Slight local corrosion and moderate corrosion at scratch.
Red lead primer (*e*).	Considerable chalking, moderate local corrosion, moderate pitting, and moderate corrosion at scratch.	Considerable chalking, considerable local corrosion, moderate pitting, and severe corrosion at scratch.	All specimens withdrawn after 4 months (Table VI).	All specimens withdrawn after 4 months (Table VI).	All specimens withdrawn after 4 months (Table VI).	All specimens withdrawn after 4 months (Table VI).	*Moderate local corrosion and slight corrosion at scratch.
Red lead primer + D.T.D. 63A top coat (*j*).	Trace of crumbling, moderate local corrosion, moderate pitting, and considerable corrosion at scratch.	Trace of crumbling, considerable local corrosion, moderate pitting, and very severe corrosion at scratch.	All specimens withdrawn after 1½ years (Table VI).	All specimens withdrawn after 1½ years (Table VI).	Very severe crumbling, considerable local corrosion, considerable pitting, and very severe corrosion at scratch.	*Very severe crumbling, moderate local corrosion, considerable pitting, and very severe corrosion at scratch.	*Moderate local corrosion and severe corrosion at scratch.
Red lead primer + D.T.D. 260A top coat (*n*).	Trace of crumbling, slight local corrosion, moderate pitting, and considerable corrosion at scratch.	Trace of crumbling, considerable local corrosion, moderate pitting, and very severe corrosion at scratch.	All specimens withdrawn after 1½ years (Table VI).	All specimens withdrawn after 1½ years (Table VI).	Very severe crumbling, considerable local corrosion, considerable pitting, and very severe corrosion at scratch.	*Very severe crumbling, severe local corrosion, considerable pitting, and very severe corrosion at scratch.	*Moderate local corrosion and severe corrosion at scratch.
Iron oxide primer (*f*).	Slight chalking, very slight local corrosion, and very slight corrosion at scratch.	Moderate chalking, slight local corrosion, trace of pitting, and very slight corrosion at scratch.	Very slight chalking, very slight local corrosion, moderate pitting, and trace of corrosion at scratch.	*Slight chalking, moderate local corrosion, and very slight corrosion at scratch.	Slight chalking, slight local corrosion, very slight pitting, and trace of corrosion at scratch.	*Very severe flaking, considerable local corrosion, moderate pitting, and very severe corrosion at scratch.	*Moderate local corrosion and very severe corrosion at scratch.
Iron oxide primer + D.T.D. 63A top coat (*k*).	Slight flaking, very slight local corrosion, and very slight corrosion at scratch.	Considerable flaking, slight local corrosion, trace of pitting, and very slight corrosion at scratch.	Slight flaking, very slight local corrosion, moderate pitting, and slight corrosion at scratch.	Very severe flaking, severe local corrosion, severe pitting, and slight corrosion at scratch.	Severe flaking, slight local corrosion, slight pitting, and trace of corrosion at scratch.	Severe flaking, severe local corrosion, considerable pitting, and and severe corrosion at scratch.	*Moderate local corrosion and very severe corrosion at scratch.
Iron oxide primer + D.T.D. 260A top coat (*o*).	Slight flaking, very slight local corrosion, and trace of corrosion at scratch.	Considerable flaking, very slight local corrosion, trace of pitting, and very slight corrosion at scratch.	Severe flaking, very slight local corrosion, slight pitting, and very slight corrosion at scratch.	Very severe flaking, severe local corrosion, considerable pitting, and slight corrosion at scratch.	Considerable flaking, slight local corrosion, very slight pitting, and trace of corrosion at scratch.	Severe flaking, severe local corrosion, considerable pitting, and severe corrosion at scratch.	*Slight local corrosion and very severe corrosion at scratch.
Plain metal (untreated) (*a*).	Slight to moderate general corrosion, and trace of pitting.	Moderate general corrosion and trace of pitting.	Slight general corrosion and moderare pitting.	*Severe general corrosion and moderate pitting.	Slight general corrosion and moderate pitting.	Considerable general corrosion and moderate pitting.	All specimens withdrawn after 1½ years (Table VI).
Plain metal (chromated or pickled) (*b*).	Slight to moderate general corrosion and trace of pitting.	Moderate general corrosion and trace of pitting.	Slight general corrosion and slight pitting.	*Severe general corrosion and moderate pitting.	Slight general corrosion and moderate pitting.	Considerable general corrosion and moderate pitting.	*Very severe general corrosion.

* All material withdrawn after 2½ years (results given after 2½ years).

TABLE III.—*Summary of Results for Widnes (Industrial) Specimens after 3½ Years' Exposure.*

Paint System	Main Paint Failures and Corrosion				
	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 118, normal purity	D.T.D. 59A, normal purity	Steel
Zinc chromate primer (*c*).	Slight pitting on back only and trace of corrosion at scratch.	Slight local corrosion on back only, very slight pitting on front only, and very slight corrosion at scratch.	*Trace of flaking and slight to moderate pitting on back only.	*Trace of flaking, slight local corrosion, and slight corrosion at scratch.	*Moderate local corrosion on front only, moderate general corrosion on back only, and severe corrosion at scratch.
Zinc chromate primer + D.T.D. 63A top coat (*g*).	Considerable flaking at scratch and very slight corrosion at scratch.	Very slight corrosion at scratch.	Slight flaking on front, very severe flaking on back, moderate local corrosion on back only, slight pitting on front only, and trace of corrosion at scratch.	*Very slight flaking.	Considerable flaking at scratch only, moderate local corrosion, and considerable corrosion at scratch.
Zinc chromate primer + D.T.D. 260A top coat (*l*).	Trace of corrosion at scratch.	Trace of corrosion at scratch.	Severe flaking, moderate local corrosion, and trace of corrosion at scratch.	*None.	Severe flaking at scratch only, very slight local corrosion on front only, and considerable corrosion at scratch.
Zinc tetroxychromate primer (*d*).	Considerable erosion, slight local corrosion on front only, slight pitting on back only, and trace of corrosion at scratch.	Severe erosion, moderate local corrosion on front only, slight pitting on back only, and very slight corrosion at scratch.	Severe erosion, moderate local corrosion, and very slight corrosion at scratch.	*Trace of flaking, slight local corrosion, and slight corrosion at scratch.	*Moderate local corrosion and severe corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*).	Trace of corrosion at scratch.	Very slight corrosion at scratch.	Considerable flaking at scratch only and trace of corrosion at scratch.	*Trace of flaking.	Severe flaking at scratch only, slight to moderate local corrosion, and very severe corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*).	Trace of corrosion at scratch.	Trace of corrosion at scratch.	Slight flaking at scratch only and trace of corrosion at scratch.	*None.	Severe flaking at scratch only and very severe corrosion at scratch.
Red lead primer (*e*).	Moderate chalking, severe general corrosion on back only, moderate pitting on front only, and slight corrosion at scratch.	Moderate chalking, considerable local corrosion on front only, severe general corrosion on back only, and slight corrosion at scratch.	All specimens withdrawn after 8 months (Table VI).	All specimens withdrawn after 8 months (Table VI).	All specimens withdrawn after 1½ years (Table VI).
Red lead primer + D.T.D. 63A top coat (*j*).	Moderate local corrosion, slight pitting, and very slight corrosion at scratch.	Considerable local corrosion, moderate pitting, and slight corrosion at scratch.	Very severe flaking, considerable general corrosion, and very slight corrosion at scratch.	*Severe flaking at scratch and moderate corrosion at scratch.	Severe corrosion at scratch.
Red lead primer + D.T.D. 260A top coat (*n*).	Slight corrosion at scratch.	Slight pitting on back only and very slight corrosion at scratch.	Moderate flaking on back only, considerable local corrosion on front only, and considerable corrosion at scratch.	*Severe flaking at scratch, moderate local corrosion, and slight corrosion at scratch.	Very slight local corrosion and severe corrosion at scratch.
Iron oxide primer (*f*).	Severe chalking, slight local corrosion, slight pitting, and very slight corrosion at scratch.	Severe chalking, considerable local corrosion, moderate pitting, and slight corrosion at scratch.	Very severe flaking and considerable general corrosion.	*Moderate chalking and slight flaking, considerable flaking at scratch, moderate local corrosion, and very slight corrosion at scratch.	All specimens withdrawn after 1½ years (Table VI).
Iron oxide primer + D.T.D. 63A top coat (*k*).	Slight flaking on back only, slight local corrosion, and slight corrosion at scratch.	Slight corrosion at scratch.	Very severe flaking, moderate local corrosion, and very slight corrosion at scratch.	*Considerable flaking and trace of pitting.	Considerable flaking at scratch only, moderate to considerable local corrosion, slight pitting, and very severe corrosion at scratch.
Iron oxide primer + D.T.D. 260A top coat (*o*).	Trace of pitting, and very slight corrosion at scratch.	Trace of corrosion at scratch.	Very severe flaking and considerable local corrosion.	*Slight flaking and slight local corrosion.	Slight flaking at scratch only, slight local corrosion on back only, and very severe corrosion at scratch.
Plain metal (untreated) (*a*).	Slight to moderate general corrosion.	Moderate general corrosion.	Considerable general corrosion and slight pitting.	*Moderate general corrosion and slight to moderate pitting.	Very severe general corrosion.
Plain metal (chromated or pickled) (*b*).	Slight to moderate general corrosion.	Moderate general corrosion.	Moderate general corrosion and moderate pitting.	*Moderate general corrosion and slight to moderate pitting.	Very severe general corrosion.

* All material withdrawn after 2½ years (results given after 2½ years).

TABLE IV.—*Summary of Results for Frodsham (Rural) Specimens after 3½ Years' Exposure.*

Paint System	Main Paint Failures and Corrosion				
	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 118, normal purity	D.T.D. 59A, normal purity	Steel
Zinc chromate primer (*c*).	None.	None.	Moderate pitting on back only and very slight corrosion at scratch.	•Very slight flaking and very slight pitting.	Slight to moderate local corrosion on front only, severe general corrosion on back only, and moderate corrosion at scratch.
Zinc chromate primer + D.T.D. 63A top coat (*g*).	Moderate flaking at scratch.	Moderate flaking at scratch.	Very slight flaking, slight pitting on back only, and slight corrosion at scratch.	Moderate flaking, and very slight local corrosion.	Severe flaking at scratch, slight pitting on front only, and slight corrosion at scratch.
Zinc chromate primer + D.T.D. 260A top coat (*l*).	None.	None.	Slight flaking at scratch and very slight corrosion at scratch.	Slight flaking, very slight pitting, and trace of corrosion at scratch.	Very slight corrosion at scratch.
Zinc tetroxychromate primer (*d*).	Severe erosion.	Severe erosion.	Severe erosion, moderate local corrosion on front only, and slight pitting on back only.	•Trace of flaking, slight local corrosion, very slight pitting, and very slight corrosion at scratch.	Moderate erosion, moderate to considerable local corrosion, and considerable corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*).	Moderate flaking at scratch.	None.	Considerable flaking at scratch and trace of corrosion at scratch.	Very slight flaking and trace of corrosion at scratch.	Considerable flaking at scratch and considerable corrosion at scratch.
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*).	None.	None.	Severe flaking at scratch and trace of corrosion at scratch.	Trace of flaking and trace of corrosion at scratch.	Very slight corrosion at scratch.
Red lead primer (*e*).	Moderate chalking and crumbling, considerable erosion, and considerable local corrosion on back only.	Moderate chalking and crumbling, considerable erosion, and moderate to considerable local corrosion.	All specimens withdrawn after 8 months (Table VI).	All specimens withdrawn after 1½ years (Table VI).	Moderate chalking and erosion, severe local corrosion, and considerable corrosion at scratch.
Red lead primer + D.T.D. 63A top coat (*j*).	Trace of corrosion at scratch.	Slight pitting on back only and trace of corrosion at scratch.	Trace of flaking at scratch and considerable corrosion at scratch.	Considerable flaking at scratch and slight corrosion at scratch.	Very slight pitting on back only and considerable corrosion at scratch.
Red lead primer + D.T.D. 260A top coat (*n*).	Slight corrosion at scratch.	None.	Slight pitting on front only and considerable corrosion at scratch.	Considerable flaking and slight corrosion at scratch.	Slight corrosion at scratch.
Iron oxide primer (*f*).	Severe chalking.	Severe chalking.	Considerable chalking, trace of flaking, considerable local corrosion, and moderate corrosion at scratch.	•Moderate chalking and very severe flaking.	Very severe flaking and severe general corrosion.
Iron oxide primer + D.T.D. 63A top coat (*k*).	None.	None.	Very severe flaking, slight pitting, and very slight corrosion at scratch.	Very severe flaking at scratch, slight pitting, and trace of corrosion at scratch.	Considerable flaking at scratch, moderate local corrosion on front only, very slight pitting on back only, and severe corrosion at scratch.
Iron oxide primer + D.T.D. 260A top coat (*o*).	None.	None.	Considerable erosion on front only, severe flaking, and slight pitting.	Moderate flaking and very slight corrosion at scratch.	Moderate flaking, very slight pitting on back only, and moderate corrosion at scratch.
Plain metal (untreated) (*a*).	Slight general corrosion	Moderate general corrosion and slight pitting.	Considerable general corrosion and moderate pitting.	Slight general corrosion and slight pitting.	Severe general corrosion.
Plain metal (chromated or pickled) (*b*).	Slight general corrosion.	Moderate general corrosion and slight pitting.	Considerable general corrosion and moderate pitting.	Slight general corrosion and slight pitting.	Severe general corrosion.

• All material withdrawn after 2½ years (results given after 2½ years).

test (sea-water spray), industrial (Widnes, Lancashire), and rural (Frodsham, Cheshire).

Apart from much greater exposure to wind at the marine station, which encouraged flaking of the top coats, more especially from the scratch line, the specimens at this station behaved in a similar manner to those in the accelerated test, where specimens exposed outdoors were sprayed twice daily with sea water, and the relative merits and disadvantages of the various materials were the same for the two sites. Further comments on the accelerated test are therefore omitted.

At each site the primers are considered in the order red lead, iron oxide, and the two chromates. Results for the panels which were primed only are given first in each case, followed by results for the full paint system with respect to each of the primer types.

1. Marine Station

The red lead primer suffered considerably by chalking and erosion in the absence of a top coat; this was apparent on both light alloys and steel. With top coats there was considerable failure owing to crumbling of the whole paint film. The red lead primer accelerated the corrosion of the light alloys and especially the magnesium alloys, which were severely corroded after about 6 weeks' exposure even with the full paint system. The rapid deterioration caused by red lead primers on magnesium alloys and, to a lesser extent, on aluminium alloys is illustrated by the values given in Tables V and VI and the specimens shown in Figs. 1 and 2 (Plates LXXV and LXXVI). The adverse effect of the red lead primer, noted with the primer alone,[1] was thus delayed but not prevented by the completion of the full paint system.

The iron oxide primer maintained its colour in the absence of a top coat, although it suffered by rather marked chalking. This was prevented by the application of the top coat. The iron oxide primer, with or without top coats, still provided appreciable protection to aluminium alloys, but protection of magnesium alloys had practically ceased after 3½ years' exposure (Figs. 1 and 2). There was only one case suggesting that the iron oxide primer might have a slightly adverse effect on the magnesium alloys. This was the normal-purity magnesium alloy, D.T.D. 59A, primed with iron oxide and without top coat, which showed more severe corrosion at the scratch than the unpainted control specimen that had been similarly chromate treated (Table II). This may possibly have been due to scratching through the chromate film on the former specimen, whereas the control specimen was not scratched.

The zinc chromate and zinc tetroxychromate primers afforded more protection than iron oxide to the light metals. This was clearly shown by those specimens where the iron oxide primer had failed, i.e. mainly on the magnesium alloys (Figs. 1 and 3, Plates LXXV and LXXVII). There was little evidence to distinguish between the two chromate primers in the full paint systems, although where no top coat had been applied the zinc tetroxychromate had suffered a moderate amount of erosion. The zinc chromate primer appeared to be slightly preferable on the less corrosion-resistant metal (B.S.S. 5L3, D.T.D. 118, and D.T.D. 59A of normal purity), whereas the zinc tetroxychromate primer appeared to be slightly preferable on the more resistant metal (99·5% purity aluminium, D.T.D. 120A, and D.T.D. 59A of high purity).

Some serious flaking of the top coats occurred on all panels exposed with chromate or iron oxide primers, although the oil-medium (D.T.D. 260A) top coat had a tendency to give better adhesion than the cellulose-medium (D.T.D. 63A) top coat. The cellulose top coat suffered considerably by erosion (Fig. 1). The serious flaking and erosion at the marine station were attributed partly to the very exposed conditions. The specimens were exposed to the full forces of prevailing cross-winds and in stormy weather to actual sea-spray. Flaking occurred almost entirely between the various paint coats rather than between paint and metal.

Chromate treatment of magnesium alloys without painting, and especially those of high-purity base, improved the corrosion-resistance markedly. After

Table V.—*Losses in Weight for D.T.D.* 118 (*Normal Purity*) *and Steel Specimens after* 2½ *Years' Exposure at the Marine Site and in the Accelerated Tests.*

Paint System	Weight Loss, g./m.^{2}			
	Marine Site (Dungeness)		Accelerated Test (Chalfont Park)	
	D.T.D. 118, normal purity	Steel	D.T.D. 118 normal purity	Steel
Zinc chromate primer (*c*)	33	251	9	241
Zinc chromate primer + D.T.D. 63A top coat (*g*)	9	414	3	1560
Zinc chromate primer + D.T.D. 260A top coat (*l*)	4	17	1	48
Zinc tetroxychromate primer (*d*)	32	469	21	1530
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*)	3	173	4	1050
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*)	0	46	3	508
Red lead primer (*e*)	177 *	448	214 †	2330
Red lead primer + D.T.D. 63A top coat (*j*)	371 ‡	173	345 ‡	1920
Red lead primer + D.T.D. 260A top coat (*n*)	263 ‡	197	398 ‡	1480
Iron oxide primer (*f*)	37	711	66	3150
Iron oxide primer + D.T.D. 63A top coat (*k*)	16	756	23	2180
Iron oxide primer + D.T.D. 260A top coat (*o*)	17	336	38	1020
Plain metal (untreated) (*a*)	90	905 ‡	342	4390 ‡
Plain metal (chromated or pickled) (*b*)	77	1560	139	4860

All material withdrawn after * 4 months (results given after 4 months), † 9 months (results given after 9 months), and ‡ 1½ years (results given after 1½ years).

TABLE VI.—*Summary Table : the Approximate Time in Years Required for the Panels to Suffer Slight Corrosion.*

Paint System	Marine Site (Dungeness)							Accelerated Tests (Chalfont Park)						
	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 120A, high purity	D.T.D. 118, normal purity	D.T.D. 59A, high purity	D.T.D. 59A, normal purity	Steel	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 120A, high purity	D.T.D. 118, normal purity	D.T.D. 59A, high purity	D.T.D. 59A, normal purity	Steel
Zinc chromate primer (*c*)	>3½	>3½	>3½	1	>3½	1	½	>3½	>3½	>3½	1½	>3½	1	½
Zinc chromate primer + D.T.D. 63A top coat (*g*)	>3½	>3½	>3½	1½	>3½	1	1½	>3½	>3½	>3½	1½	>3½	1½	1½
Zinc chromate primer + D.T.D. 260A top coat (*l*)	>3½	>3½	>3½	1½	>3½	1	2	>3½	>3½	>3½	2½	>3½	1½	1½
Zinc tetroxychromate primer (*d*)	>3½	>3½	>3½	1	3½	1	½	>3½	>3½	>3½	1½	>3½	1	½
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*)	>3½	>3½	>3½	1½	3½	1	1½	>3½	>3½	>3½	1½	>3½	1	1
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*)	>3½	>3½	>3½	1½	>3½	1	1½	>3½	>3½	>3½	1½	>3½	1½	1½
Red lead primer (*e*)	¼	¼	< ¼	< ¼	< ¼	< ¼	½	¼	¼	< ¼	< ¼	< ¼	< ¼	½
Red lead primer + D.T.D. 63A top coat (*j*)	1	1	< ¼	< ¼	< ¼	< ¼	1½	1	½	< ¼	< ¼	< ¼	< ¼	1
Red lead primer + D.T.D. 260A top coat (*n*)	1	1	< ¼	< ¼	< ¼	< ¼	1½	1	½	< ¼	< ¼	< ¼	< ¼	1½
Iron oxide primer (*f*)	>3½	3½	1½	½	3½	1	¼	3½	2½	1	1	>3½	½	½
Iron oxide primer + D.T.D. 63A top coat (*k*)	>3½	3½	2½	1	3½	1	1	3½	2½	>3½	1½	3½	1	1
Iron oxide primer + D.T.D. 260A top coat (*o*)	>3½	>3½	3½	1	3½	1	1	>3½	2½	1½	1	3½	1	1
Plain metal (untreated) (*a*)	1½	½	1½	½	1½	¼	¼	1½	1	...	¼	1½	½	¼
Plain metal (chromated or pickled) (*b*)	1½	½	1½	½	2	¼	¼	1½	1	1½	½	1½	½	¼

Paint System	Industrial Site (Widnes)					Rural Site (Frodsham)				
	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 118, normal purity	D.T.D. 59A, normal purity	Steel	Aluminium (99·5%)	B.S.S. 5L3	D.T.D. 118, normal purity	D.T.D. 59A, normal purity	Steel
Zinc chromate primer (*c*)	3½	3½	2½	1	½	>3½	>3½	1½	>2½ *	½
Zinc chromate primer + D.T.D. 63A top coat (*g*)	>3½	>3½	1½	>2½ *	1½	>3½	>3½	1½	>3½	2
Zinc chromate primer + D.T.D. 260A top coat (*l*)	>3½	>3½	3	>2½ *	1½	>3½	>3½	>3½	>3½	>3½
Zinc tetroxychromate primer (*d*)	3½	3	1½	1½	1	>3½	>3½	3	2½	1½
Zinc tetroxychromate primer + D.T.D. 63A top coat (*h*)	>3½	>3½	>3½	>2½ *	1½	>3½	>3½	>3½	>3½	2
Zinc tetroxychromate primer + D.T.D. 260A top coat (*m*)	>3½	>3½	>3½	>2½ *	1½	>3½	>3½	>3½	>3½	>3½
Red lead primer (*e*)	2	½	< ¼	< ¼	¼	1	½	< ¼	<1	1½
Red lead primer + D.T.D. 63A top coat (*j*)	2½	2½	2	1½	1	>3½	>3½	1	3½	1½
Red lead primer + D.T.D. 260A top coat (*n*)	3½	3½	1½	1½	1	>3½	>3½	1	3½	3½
Iron oxide primer (*f*)	3	1½	1	½	¼	>3½	>3½	1	<1	½
Iron oxide primer + D.T.D. 63A top coat (*k*)	3½	3½	2	2½	1	>3½	>3½	1½	3½	1½
Iron oxide primer + D.T.D. 260A top coat (*o*)	>3½	>3½	2	>2½ *	1	>3½	>3½	1½	>3½	2½
Plain metal (untreated) (*a*)	1	1	½	½	< ¼	1½	½	½	½	< ¼
Plain metal (chromated or pickled) (*b*)	1	1	½	½	< ¼	1½	½	½	½	< ¼

* Specimens withdrawn after 2½ years before showing slight corrosion.

prolonged exposure, however, pitting of the chromated metal was evident, and this pitting was deeper than on the untreated and unprotected aluminium and B.S.S. 5L3.

The results showed that except for the systems using red lead primer, all the light alloys tested were markedly superior to mild steel. For example, the worst of the light alloys is compared with steel in Table V (p. 485). Paint failure on the mild-steel specimens was due partly to flaking of the paint film and partly to rusting of the metal causing blisters which subsequently dislodged the paint films.

2. Industrial Station

In the case of magnesium alloys (D.T.D. 59A) records ceased after 2½ years, since many of these specimens were either lost or heavily damaged during a severe storm. No high-purity-base magnesium alloys were exposed at the industrial station.

Red lead primer unprotected by top coats failed by chalking, leading to crumbling of the paint film. Definite acceleration of corrosion occurred on the light metals before failure of the paint film as a whole, especially on normal-purity magnesium alloys, and severe corrosion developed on these latter alloys within 6 weeks in the absence of top coats. On the aluminium alloys some acceleration of corrosion was noted in the absence of top coats, particularly on the backs of the specimens, where there was a higher moisture retention. Red lead primer without a top coat ceased to give adequate protection to mild steel after 12 months. A very marked improvement in the behaviour of red lead primer resulted from the application of top coats, and, although corrosion occurred, no acceleration was observed on any metal. The complete paint systems based on red lead primers were on the whole inferior to the corresponding ones based on iron oxide primers for the protection of light alloys (Table III, p. 483). It seemed probable from the behaviour of specimens primed but without top coats, that acceleration of the corrosion of light alloys would occur in time with lead primers if the top coats were not adequately maintained. In general, the oil top coat (D.T.D. 260A) proved somewhat superior to the cellulose top coat (D.T.D. 63A).

Iron oxide primers also failed by chalking when unprotected by top coats, but failure developed more slowly than with red lead. Colour and gloss were maintained for about 12 months. As a rule, iron oxide in the absence of top coats gave quite good protection to all metals while the paint film was intact, but as paint failure developed protection ceased to become effective. This was attributed to lack of the chemical corrosion-inhibitive properties possessed by the chromates on all metals and by red lead on steel. With mild steel, protection by the iron oxide primer alone was poor, and rusting became severe after about 12 months. On normal-purity magnesium alloy D.T.D. 118, severe flaking of the primer with considerable corrosion occurred after about 3 years. The addition of top coats gave considerable improvement to the performance of iron oxide primers, that with D.T.D. 260A again proving the more effective, particularly on magnesium alloys and mild steel. A tendency was noted for the two primer coats on the top-coated systems to flake apart, especially on the magnesium alloy D.T.D. 118, thereby reducing the effective life of the systems.

The behaviour of the two zinc chromate primers was definitely better than that of either iron oxide or red lead. In the absence of top coats, both pigments gave good protection to the light metals for 3 years (except perhaps with the normal-purity magnesium) and on mild steel the protection was considerably better than with red lead. After 3 years, the single coat of zinc tetroxychromate failed by progressive erosion, the primer continuing to inhibit corrosion until, after 3½ years, the paint film became too thin to prevent corrosion on the least resistant metals. Zinc tetroxychromate gave rather better results than the normal chromate on mild steel.

With the top-coated systems there was little difference in performance between the two chromate pigments. The tetroxychromate appeared rather better on normal-purity magnesium alloy D.T.D. 118 and on mild steel, but the zinc chromate was more effective in preventing scratch-line corrosion on mild steel. Apart from flaking, described below, both top-coated systems gave very satisfactory results with both chromates.

There was a serious, though erratic, tendency for flaking to occur on top-coated systems based on the two zinc chromate and iron oxide primers, particularly at the scratch lines. On most metals this took place between the two primer coats, the second primer coat and the top coats flaking away together from the first primer coat. On the normal-purity magnesium alloy D.T.D. 59A, the flaking tended to occur down to the bare metal. The relative freedom from flaking with the red lead primer was attributed to the more matt surface obtained with this formulation. Undoubtedly the oil medium (D.T.D. 260A) gave better adhesion than the cellulose medium (D.T.D. 63A) over the primers tested.

The cellulose top coats (D.T.D. 63A) failed after about 3 years by erosion. The industrial site was in a very exposed situation, and this tended to accentuate the degree of erosion, particularly on the panel fronts which faced partly in the direction of the prevailing winds.

Chromate treatment of the magnesium alloys gave a definite improvement in corrosion-resistance in the case of D.T.D. 118 and, initially at least, with D.T.D. 59A.

The light alloy specimens suffered very much less corrosion than the corresponding mild-steel specimens, whether painted or not. Even the magnesium alloys (which were the most susceptible of the light alloys to corrosion) when primed with red lead (the worst exposure condition) have not shown the same depth of corrosion encountered on the best mild-steel specimens. The latter suffered particularly at the scratch lines.

3. Rural Station

No high-purity magnesium alloys were exposed at the rural station.

The red lead primer, when not protected by top coats, failed by chalking, which led to crumbling of the paint film, and considerable erosion developed on the specimens which were exposed for more than 2½ years. In the absence of top coats, definite acceleration of corrosion by this primer was evident on light metals before complete failure of the paint film; on the aluminium alloys this was restricted to the panel backs; the performance on mild steel, as at the industrial site, was not good, and marked corrosion developed. Red lead primer protected by top coats gave comparatively good results even on magnesium alloys, corrosion being mainly restricted to scratch lines (Table IV, p. 484).

The iron oxide primer unprotected by top coats suffered progressive chalking comparable with that at the industrial site. No acceleration of corrosion was noted, and useful protection was afforded to aluminium alloys while the paint film remained intact. With the magnesium alloy D.T.D. 118, moderate local corrosion developed after about 18 months and on the magnesium alloy D.T.D. 59A severe flaking occurred. Generally, however, the areas attacked were less than at the industrial and marine sites. On mild steel, corrosion was severe after 12 months.

The top-coated systems using iron oxide primer gave excellent protection to aluminium alloys for the full period of exposure, but protection of the magnesium alloys was handicapped by severe flaking. There was little to choose between the two top coats. On mild steel the D.T.D. 260A top coat gave quite good protection even after 3½ years, but on this metal the D.T.D. 63A top coat was not quite as satisfactory.

As might be expected from the milder corrosive conditions prevailing at Frodsham, the specimens primed with single coats of zinc chromate and tetroxychromate withstood corrosion even better than at the other sites. There was little to choose between the two primers until erosion seriously reduced the thickness of the tetroxychromate film after about 3 years' exposure. In spite of this erosion, no corrosion occurred on aluminium alloys after 3½ years. The tetroxychromate gave rather better protection to mild steel than the ordinary chromate. Very good protection was maintained with full paint systems incorporating chromate primers for all metals after 3½ years.

As at the other sites, erratic flaking was noted with the complete paint systems based on iron oxide, zinc chromate, and zinc tetroxychromate primers. Here again adhesion tended to fail between the two primer coats. In spite of this tendency towards flaking, no appreciable corrosion occurred with the chromate primers. The flaking was generally rather worse with the systems using the cellulose top-coat (D.T.D. 63A) than with the oil (D.T.D. 260A) and usually commenced at the scratch lines. With the magnesium alloy D.T.D. 59A flaking occurred down to the metal surface. Erosion of the cellulose top coats was perhaps even more severe at Frodsham than at the other sites, despite the comparatively sheltered situation.

A slight improvement in the performance of both the unpainted magnesium alloys resulted from chromate treatment.

As at the other sites, corrosion of unpainted light alloys was generally considerably less than with corresponding mild-steel panels. Under these comparatively mild conditions the corrosion of light alloys was prevented, and the corrosion of mild steel was only very slight with the best of the complete paint systems.

III.—DISCUSSION OF RESULTS

The first paper describing this investigation[1] summarized the provisional results obtained after about 6 months' exposure at the four stations. After exposure for 3½ years, breakdown has reached the stage where a reasonably complete assessment can be made for all the systems tested.

The exposure to sea-water spray was intended to give acceleration of corrosion as compared with the marine site at Dungeness, and although this acceleration was not achieved, the term is retained for convenience. The marine site was in a very exposed situation, with prevailing winds blowing across the fronts of the specimens. This gave a much greater degree of flaking on the fronts of the specimens, especially at the scratch lines and edges, at the marine site than in the accelerated tests, which were carried out in a fairly sheltered location. The lack of acceleration on specimens subject to sea-water spray tests is attributed partly to this factor and partly to under-estimation of the severity of conditions at Dungeness. The accelerated tests served a useful purpose, however, in providing severe marine conditions on specimens under continuous observation.

The light alloys in general suffered rather less attack at the industrial site than at the marine or accelerated test sites. The industrial and marine sites were comparable in exposure to prevailing winds, and the consequent flaking on the front surfaces of the specimens.

At the rural site the light metals were comparatively mildly corroded, but there was marked attack on mild steel. The site had relatively long periods of sunshine unfiltered by smoke layers in the atmosphere, and moisture deposition was of a high order owing to frequent night mists. Chalking was consequently heavier than at the other sites. Winds and dust storms from nearby arable land were sufficiently frequent to produce erosion and flaking.

Corrosion was frequently more severe on the panel backs than on the front surfaces, probably owing to the sheltering from direct solar radiation and wind causing slower evaporation of condensed moisture and to a much smaller removal of corrosive deposits by rain.

At all the sites definite evidence of acceleration of corrosion of light metals was observed with single coats of red lead primer, and this was particularly marked on magnesium alloys. At the marine and accelerated exposure sites red lead primer accelerated corrosion of all the light metals even when complete paint systems were applied. No such acceleration was observed at the industrial and rural sites with the complete paint systems exposed for 3½ years, but these did not give as efficient protection to light alloys as the systems free from lead pigments, and in time might give acceleration even in these atmospheres, especially if maintenance were poor. Red lead primer gave fairly effective protection to mild steel at the marine, accelerated, and rural sites, but it was less satisfactory at the industrial site.

Iron oxide primer generally gave a good measure of protection to light metals at all of the sites, the longest life being obtained at the less corrosive sites (rural and industrial) or with the more corrosion-resistant metals (aluminium alloys and high-purity magnesium alloys). Iron oxide on mild steel was in general inferior to red lead, zinc chromate, and zinc tetroxychromate. It was apparent from the type of failure that iron oxide conferred protection by physical rather than chemical means, and once film breakdown occurred corrosion proceeded owing to the absence of inhibiting properties in the pigment. These results confirm the suitability of iron oxide as a diluent for inhibiting pigments and the greater importance of maintenance when iron oxide is used without additions of inhibiting pigments.

The zinc chromate and zinc tetroxychromate primers were outstandingly successful at all the sites, and for practical purposes these two primers are considered to have comparable protective properties on light metals and mild steel. The good performance of zinc tetroxychromate was achieved with a much lower pigmentation of the non-volatile portion of the paint than with any of the other priming pigments. The pigment volume concentrations (on the total non-volatiles) for zinc tetroxychromate, zinc chromate, red lead, and turkey red oxide were 24, 36, 50, and 38% respectively. One of the authors (J. G. R.) has found from other recent work that the pigment volume concentrations of zinc tetroxychromate can with advantage be increased without loss of good brushing properties. This has a tendency to harden the primer film, which consequently suffers less from erosion in the absence of top coats, and to decrease the gloss, thus giving better adhesion between paint coats.

Adhesion between the various paint coats in the complete paint systems was generally rather unsatisfactory, except in the case of the red lead primer (which had a matt surface). The possibility of this type of failure (due to the unusual formulation of the primer) was foreseen when the tests were designed, but it was considered that the primers should be formulated from single pigments, rather than from mixed pigments and extenders, in order to simplify direct comparison of the individual priming pigments. In designing formulations for commercial use, these considerations do not apply, and mixed pigments can be employed.

In general, the cellulose top coat (D.T.D. 63A) tended to suffer erosion and to give rather more flaking between the priming coats and top coats of the full systems than the corresponding oil top coat (D.T.D. 260A). Possibly this difference was partly due to the use of a single primer medium for both systems and the better leafing characteristics of aluminium paste in oil varnish media than in cellulose media.

IV.—CONCLUSIONS

The general conclusions with respect to materials are:

(1) Red lead primers should be avoided on light metals, especially in marine atmospheres. This effect is much more marked on magnesium and its alloys than on aluminium and its alloys.

(2) Iron oxide primers with appropriate top coats are satisfactory for light and ferrous metals where corrosive conditions are not too severe.

(3) Zinc chromate or zinc tetroxychromate primers with appropriate top coats should be used on light metals for the most severe corrosive conditions. They are very suitable for ferrous metals, and are particularly appropriate for composite structures of steel and light alloys.

(4) Alkyd-linseed stand oil varnish (D.T.D. 260A) is preferred to cellulose (D.T.D. 63A) as a medium for aluminium finishing paints.

ACKNOWLEDGEMENTS

The authors wish to express their appreciation of the advice and assistance received from Messrs. Cellon's, Ltd., who made up the paints. They also thank the Magnesium Metal Corporation for the preparation of the magnesium alloy specimens.

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BIBLIOTEKA POLITECHNIKI ŚLĄS

Discussion

The Effect of Grain-Size on the Structural Changes Produced in Aluminium by Slow Deformation

By W. A. Rachinger

(*Journal*, this vol., p. 415.)

Dr. I. S. Servi,[*] S.M. (Member): The excellent experimental evidence of structural changes produced in aluminium by slow deformation at elevated temperatures is a substantial contribution to the understanding of the phenomenon of subgrain formation, because it helps to unify, at least partially, the rather contrasting theories that have been suggested on this subject. The most important experimental observations, including the effect of grain-size on structural changes at elevated temperatures, are in good agreement with the observations made, quite independently, by Servi and Grant.[†]

A comparison of as-deformed and repolished and etched structures (Figs. 3, 4, 14, 15, and 20, Plates LXI and LXII) reveals that the grain boundaries undergo marked migration during deformation. The sub-boundaries of the as-deformed structures do not coincide exactly with the sub-boundaries of the repolished and etched structures. The discontinuous dotted lines along old-grain (Fig. 15) and dendrite (Fig. 20) boundaries may be due to segregated impurities (probably oxides) that do not diffuse away at the temperature of testing. A statement concerning the etching reagent used by the author would be appreciated. It would also be interesting to know whether he has ever observed a sub-structure in cast "very large-grained aggregates" before deformation.

The results of Mr. Rachinger's investigation show that there is no physical difference between grain boundaries and sub-boundaries. Rather, there is indication that the only difference lies in the mechanism of formation, since the sub-boundary formation does not involve nucleation and growth.

The report on experimental observations of sub-grain formation by Servi, Norton, and Grant [‡] may give further explanation of the "banded appearance" of the sub-structure of Fig. 11 (Plate LXII).

The Author (*in reply*): As Dr. Servi intimates, a unification of the diverse views on the mechanism of sub-grain formation may be near at hand. The work of Cahn,[§] Honeycombe,[||] and McLean [**] indicates that the banded sub-structures arise from "polygonization under stress" of deformation bands formed by inhomogeneous straining of the grains. The mode of formation of the "polygonal" type of sub-structure at higher temperatures is more difficult to ascertain, both because of the high mobility of the sub-boundaries [††] and of the relatively small number of cells contained within each grain. One promising suggestion is that the sub-grains are, in all cases, formed from deformation bands, the width of the bands, and hence the size of the sub-grains, increasing with an increase in temperature or decrease in strain rate.

A coincidence of the deformation markings and the etched sub-boundaries is not to be expected if the sub-boundaries are mobile. The surface markings may have formed at any stage of the deformation, whereas the etched structures show only the final boundary positions. The pits visible in Fig. 14 or 15 are probably formed by gas bubbles adhering to the surface during electropolishing. It is agreed that the etching of the dendrite boundaries (Fig. 20) is probably due to segregated impurities.

The only structure observed metallographically in the cast crystals was the dendritic structure. The X-ray diffraction patterns from these crystals indicate a structure of slightly disoriented crystalline blocks. Whether this is in fact identical with the dendritic structure is not known. However, since the size of the sub-grains varies systematically with the rate and temperature of deformation, it seems unlikely that these sub-grains can be identified with any sub-structure present in the undeformed metal.

The appearance of sub-boundaries etched in a mixture of 25% HNO_3, 2% HF, and 73% water has invariably been similar to that of grain boundaries. However, the work of Wyon and Crussard [‡‡] suggests that when the angle between contiguous sub-crystals is sufficiently small the boundary will etch differently from a normal grain boundary.

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THE CONTINUITY OF SLIP LINES ACROSS A GRAIN BOUNDARY*

1469

By G. J. OGILVIE,† Ph.D., Junior Member

SYNOPSIS

Slip lines on the surfaces of polished and subsequently strained specimens of aluminium and 70 : 30 brass are sometimes continuous across a grain boundary. Investigation of the interior of the brass specimens shows that this is not merely a surface effect, but that the intersection of the slip planes does in fact lie in the grain boundary. It is shown that this intersection lies at an angle of less than 2° from one of the $\langle 110\rangle$, $\langle 112\rangle$, or $\langle 123\rangle$ directions in both the slip planes concerned. When the duration of annealing, before the final deformation, is increased, there is, within a given surface area, an increase in the number of grain-boundary sections where slip lines cross. It is therefore suggested that the boundary configuration favourable to slip-line crossing has a low interfacial free energy.

It is shown that approximately half the adjacent crystal pairs in a random aggregate are favourably oriented for slip-line crossing to occur.

I.—INTRODUCTION

In the course of a metallographic study of deformation in metals, occasional cases were seen where sets of slip lines crossed a boundary between two crystals with, in general, a change in direction. Lacombe and Beaujard [1] had also noted a similar phenomenon. The present study was undertaken to determine whether the coincidence of slip planes at the surface of the specimen was accidental or whether the effect had some crystallographic significance.

II.—EXPERIMENTAL PROCEDURE

Tensile specimens were prepared from cold-rolled $\frac{3}{32}$-in. aluminium sheet (>99·98% pure) and from commercial $\frac{3}{16}$-in. brass sheet (zinc 31·72, iron 0·04, lead 0·02%, trace of tin, balance copper). The aluminium specimens were annealed at 600° C. for 2 hr. and then electropolished, using a perchloric acid–acetic anhydride bath.[2] The brass specimens, after preliminary annealing at 600° C. *in vacuo* for 8 hr., were given a tensile strain of 3%. These specimens were then re-annealed at 600° C. *in vacuo* for periods of 1, 5, 25, 48, and 100 hr. and electropolished in orthophosphoric acid (900 g./l.). After electropolishing, the specimens were elongated 10% to produce slip lines. In all cases numerous examples were found of slip lines crossing grain boundaries (Figs. 1 and 2, Plate LXXVIII).

As slip lines in brass can be revealed by etching,[3, 4] the brass specimens were used to determine whether the phenomenon was restricted to the surface. Selected areas of the specimen, which had been annealed for 48 hr., were re-examined at levels 10, 40, and 110 microns below the original surface, the latter being removed by electropolishing.

The other brass specimens were used to study the effect of annealing time on the frequency of occurrence of boundary sections with crossing slip lines by counting the number of cases within a standard area.

III.—RESULTS

The following observations were made during the metallographic survey:

(1) The *same* slip planes meet at the boundary at different levels. This is shown by Figs. 3–5 (Plate LXXVIII), where some of the etched lines can be identified with the original slip lines. Identification of individual slip zones was effected by comparing slip-line spacings. Cut photographs were superposed to allow a direct comparison of the spacings.

(2) The portion of a grain boundary across which slip lines are continuous is always nearly straight. A boundary may have more than one such straight section, but in this case different sets of slip lines are, in general, continuous across the different sections. Fig. 2 illustrates this point: one boundary between two crystals is divided into three sections, and different sets of slip lines cross the different straight sections.

(3) Slip lines usually change direction as they cross the boundary. The frequency histogram (Fig. 6) shows that the maximum observed change in direction is 65°, while the peak of the distribution occurs at zero direction change.

It follows from (1) and (2) that the boundary is plane where slip lines are continuous across it and the intersecting slip planes meet it in parallel straight lines.

For each pair of crystals the orientation of this direction of intersection relative to both crystals was determined from measurements on the slip lines in the crystals.[5] It was found that this direction

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was ⟨110⟩, ⟨112⟩, or ⟨123⟩ in both crystals (but not necessarily the same direction in each crystal) with a maximum error of 2°. However, the boundary plane was not found to have a simple orientation in either crystal. Although slip occasionally occurred on planes other than {111}, it was found that those slip lines which crossed a boundary were always the traces of {111} planes.

It can be seen from the figures given in Table I that the number of boundary sections where slip lines cross increases with annealing time. This is so, in spite of the increase in grain-size and the corresponding decrease in the number of grain-boundary sections in the standard area. If the increase in grain-size is taken into account, increasing the annealing time from 1 to 100 hr. increases the proportion of boundary sections with crossing slip lines by a factor of 14 instead of the factor 5 indicated in Table I.

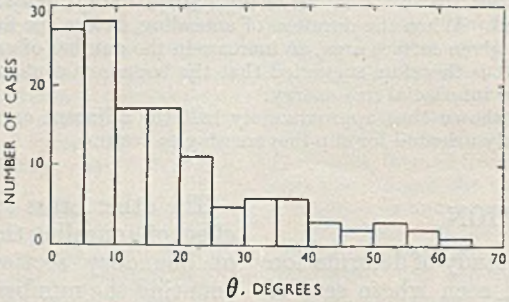

Fig. 6.—Histogram of the Deviation of Slip Lines After Crossing a Grain Boundary.

Table I.—*The Variation with Annealing Time of the Number of Boundaries at Which Slip-Line Crossing Occurs.*

Annealing Time, hr.	Average Linear Grain-Size, mm.	Number of Cases of Sets of Slip Lines Crossing a Boundary in an Area 18·57 × 4·82 mm.	Number of Grains in Standard Area, × 10^3
1	0·15	27	4
5	0·20	63	2·2
25	0·25	97	1·4
100	0·25	125	1·4

IV.—DISCUSSION

Since grain boundaries will tend to assume their most stable position during annealing and since the proportion of boundary sections with crossing slip lines increases strikingly with annealing time, it follows that plane boundaries containing a ⟨110⟩, ⟨112⟩, or ⟨123⟩ direction in each crystal are more stable than other configurations. Probably the interfacial free energy of this type of boundary is relatively small. For slip lines to cross a given boundary the following conditions must be satisfied. The relative orientation of the two crystals and of the boundary with respect to those crystals must be suitable, and the deformation must be such that the critical shear stress is exceeded on the relevant slip system in each crystal. The favourable boundary configuration may not be possible for certain ranges of relative orientations of the crystal pair. Therefore, it becomes necessary to enquire what proportion of the total number of crystal pairs in a random aggregate will have the necessary relative orientations, so that this boundary condition may be established.

For this problem it is convenient to consider a method of obtaining an arbitrary orientation from a reference orientation in a cubic crystal. Two perpendicular rotation axes are chosen, one lying in the [111] direction and the other in the (111) plane at an angle α to (say) the [$\bar{1}$01] direction. If the lattice is then rotated in turn about the fixed axis through an angle β, and about the movable axis through an angle γ, the desired orientation may be attained. One orientation may then be expressed relative to another in terms of the three parameters α, β, and γ. If α and β are restricted to a range of 60° and γ varies from 0° to 360°, then all possible orientations are described once only. A three-dimensional diagram may therefore be constructed with α, β, and γ plotted on orthogonal axes, and all orientations which could have a boundary containing, within a given maximum divergence, a ⟨110⟩, ⟨112⟩, or ⟨123⟩ direction in each crystal bounded by contour surfaces. Since the maximum divergence observed was 2°, the corresponding contours will be studied. Consideration of the ⟨110⟩, ⟨112⟩, and ⟨123⟩ directions lying only in the (111) plane of each lattice gives contours which have the form of right prisms based on parallelograms. Fig. 7 shows a section perpendicular to the γ axis of the α/β/γ diagram giving the form and location of these prisms. The total volume enclosed by them is 7·11% of the total. All other coincidences between ⟨110⟩, ⟨112⟩, and ⟨123⟩ directions will give tubular 2° contour surfaces inclined at some angle to the base of the diagram and having a variable cross-section. The average cross-section of these contours is approximately 8 deg.^{2} and the length is at least 60°. There are 1116 such contours passing through the α/β/γ diagram, and therefore the volume contained in the 2° limit (approximately 5×10^5 deg.^{2}) is approximately half the total volume ($1{\cdot}296 \times 10^6$ deg.^{2}).

To check this qualitative conclusion, direct measurements have been made along two lines through A and B (Fig. 7) perpendicular to the base plane. For fixed γ and each of the 16 intersections between the $\{111\}$ planes in the two lattices, the two angles between an intersection and the nearest $\langle 110\rangle$, $\langle 112\rangle$, or $\langle 123\rangle$ direction were determined by using a large (50-cm.) Wulff net. The larger of these two angles was taken and the minimum, δ, of the 16 angles selected in this way is plotted as a function of γ in Fig. 8 A and B. Along the line through A, 40·6% of the total orientation range was within a 2° contour and along the one through B, 78·5% was included. Details of this calculation are given in an Appendix.

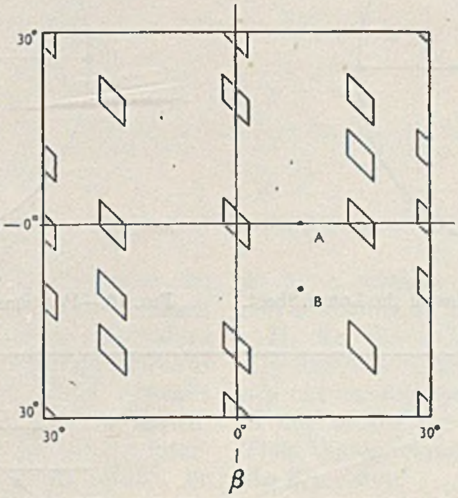

Fig. 7.—Section of the $\alpha/\beta/\gamma$ Diagram Perpendicular to the γ Axis.

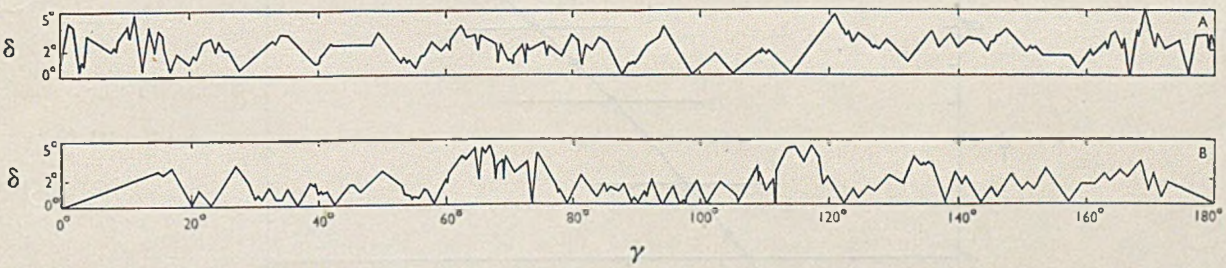

Fig. 8 A and B.—Variation of δ along the Two Lines Marked A and B in Fig. 7, Parallel to the γ Axis. The diagram is symmetrical about $\gamma = 180°$.

It seems probable, therefore, that more than half of the crystal pairs in a random aggregate are oriented with respect to each other so that at least one range of positions of their common boundary will be favourable to slip-line crossing. These positions are defined by the set of planes which have a common direction, which is the direction of intersection of the slip planes crossing the grain boundary. It is observed that the proportion of grain boundaries with crossing slip lines is small compared with the proportion of crystal pairs having a favourable relative orientation. In these experiments the effect of the deformation condition is constant, as the extension is the same in all specimens and there is no appreciable preferred orientation. Thus, the fact that the fraction of boundaries where crossing takes place increases with the time of annealing must be attributed to changes in the orientation of boundaries relative to the crystals.

Attention should be drawn to a possible implication of these conclusions. Since slip lines may cross a boundary under certain conditions, it seems likely that the effect of this boundary configuration on the plastic properties will not be as great as that of a random boundary. That is, the stiffening effect of the grain boundary will be less since dislocations can apparently pass through the boundary. Thus, for example, the elastic limit of a polycrystalline aggregate, which has boundaries only of this special type, should be less than the elastic limit of an aggregate of the same grain-size with boundaries in general positions. This point has not been studied, but it is felt that it may be worth investigation.

REFERENCES

1. P. Lacombe and L. Beaujard, *J. Inst. Metals*, 1948, **74**, 1.
2. P. A. Jacquet, *Métaux, Corrosion-Usure*, 1943, **18**, 198.
3. D. McLean, *J. Inst. Metals*, 1948, **74**, 95.
4. P. A. Jacquet, *Rev. Mét.*, 1950, **47**, 355.
5. C. S. Barrett, " Structure of Metals ", p. 40. New York: 1943 (McGraw-Hill).

APPENDIX

The Graphical Determination of δ

For the determination of δ it is necessary to have a large Wulff net on which two transparent sheets are placed. On each of these are marked the poles of low indices of one of the crystals. On the lower sheet (lattice *L*) the poles are placed so that the $[111]_L$ pole is at 90° to the centre of the projection and the $[\bar{1}01]_L$ pole is placed at $(90° - \alpha)$ from the centre, as shown in Fig. 9.

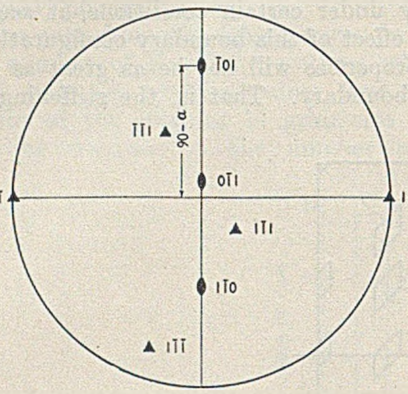

FIG. 9.—Positions of Some Low-Index Poles on the Lower Sheet.

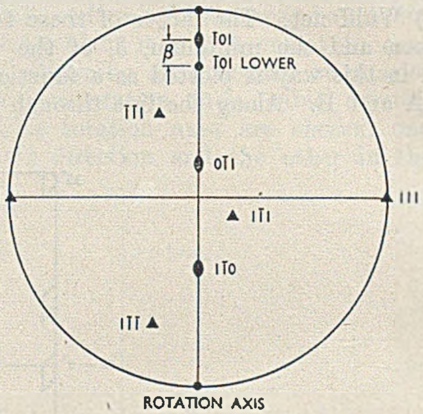

FIG. 10.—Positions of the Same Poles on the Upper Sheet.

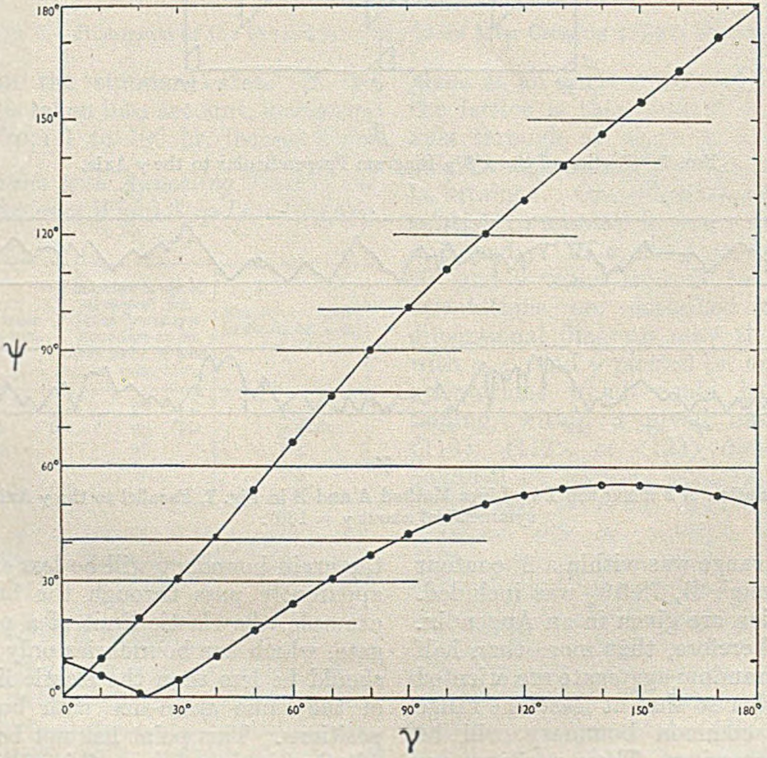

FIG. 11.—ψ/γ Curves for the Intersection Between the $(\bar{1}11)_L$ and the $(\bar{1}\bar{1}1)_U$ Planes for the Orientation Range Represented by the Line Through B in the $\alpha/\beta/\gamma$ Diagram (Fig. 7).

On the upper sheet (lattice *U*) the $[111]_U$ pole is placed at 90° to the centre of the projection, as on the lower sheet. The movable rotation axis is placed at 90° to the centre of the net and, as implied in the second paragraph of Section IV, it must lie at 90° to the $[111]_U$ pole. The two [111] poles (one on each sheet) are now superposed, and the $[\bar{1}01]_U$ pole is placed at the required angle β from the $[\bar{1}01]_L$ pole to give the array of poles shown in Fig. 10.

For each value of γ (10° intervals were chosen) the array of poles on the upper sheet must be rotated 10° about the movable rotation axis. In the calculation it is necessary to follow only the position of the $\langle 110\rangle_U$ and $\langle 111\rangle_U$ poles for reasons that will emerge subsequently. The rotation axis for this manipulation is placed coincident with the poles of the Wulff net, and the sequence of positions of each pole may be rapidly plotted by marking each 10° interval along the latitude line on which that pole is found.

If a pair of {111} planes, say the $(1\bar{1}1)_L$ and the $(11\bar{1})_U$, are chosen and their intersection plotted for each position of the $(11\bar{1})_U$ plane, the nearest ⟨110⟩, ⟨112⟩, or ⟨123⟩ direction to that intersection in lattice L and lattice U must always lie in the $(1\bar{1}1)_L$ plane and the $(11\bar{1})_U$ plane, respectively. It is necessary to measure only the angle ψ that the intersection makes with a given direction in each plane to specify its position and to obtain the values of δ in a convenient manner. A typical ψ/γ curve is shown in Fig. 11. The positions of the relevant directions can be inserted into the diagram as shown. Thus, for a given value of γ, the angle that the intersection makes with the nearest ⟨110⟩, ⟨112⟩, or ⟨123⟩ direction in each lattice may be rapidly read from the two curves. The larger of these two angles, μ, is used to determine the value of δ, and the values of γ for which μ has a maximum or minimum can be rapidly determined by sliding a vertical straight-edge across the diagram and noting the positions where each curve is at the same distance from a horizontal line. Between these points the curve is very nearly straight and is, except in exceptional circumstances, as when both curves are very nearly horizontal, sufficiently closely approximated to by a straight line. Fig. 12 gives the μ/γ plot corresponding to Fig. 11.

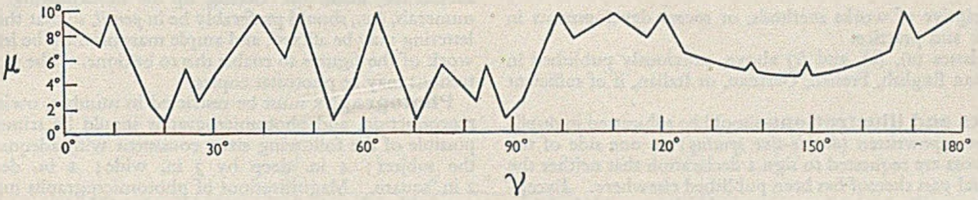

FIG. 12.—μ/γ Curve Derived from the ψ/γ Curves of Fig. 11.

If, for the other 15 possible intersections, the μ/γ curve is obtained in the same way and all the μ/γ curves superposed, then the smallest value of μ of any of the curves for given γ is the value of δ. Thus the envelope given by the minimum values is the δ/γ curve.

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A. L. Dighton and H. A. Miley, *Trans. Electrochem. Soc.*, 1942, 81, 321 (i.e. year, volume, page).

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C. Zener, "Elasticity and Anelasticity of Metals". Chicago: 1948 (University of Chicago Press).

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It is sometimes valuable to indicate the treatment of the subject by such words as: brief, exhaustive, theoretical, &c.

The synopsis should indicate newly observed facts, conclusions of an experiment or argument and, if possible, the essential parts of any new theory, treatment, apparatus, technique, &c.

It should contain the names of any new compound, mineral species, &c., and any new numerical data, such as physical constants; if this is not possible, it should draw attention to them. It is important to refer to new items and observations, even though some are incidental to the main purpose of the paper; such information may otherwise be hidden, though it is often very useful.

When giving experimental results the synopsis should indicate the methods used; for new methods the basic principle, range of operation, and degree of accuracy should be given.

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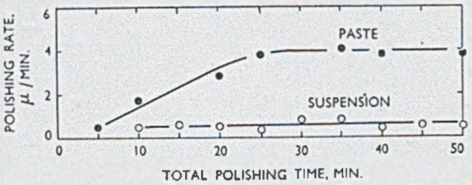

FIG. 1.—Comparative Polishing Rates for Paste-Charged and Suspension-Charged Pads. Specimen: annealed copper; diamond abrasive: (8–20) μ grade.

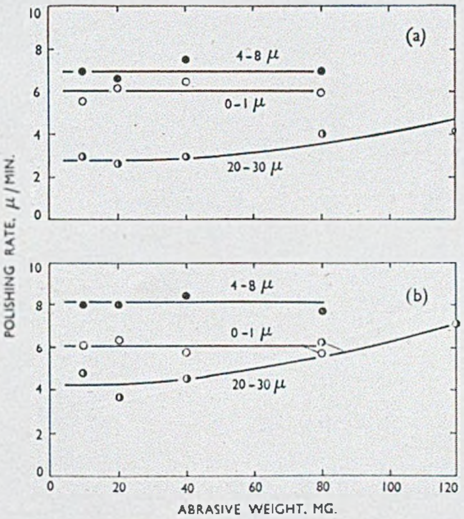

FIG. 2.—Variation in Polishing Rate with Weight of Added Abrasive. (*a*) Annealed copper; (*b*) heat-treated aluminium alloy. Abrasive added as impregnated paste.

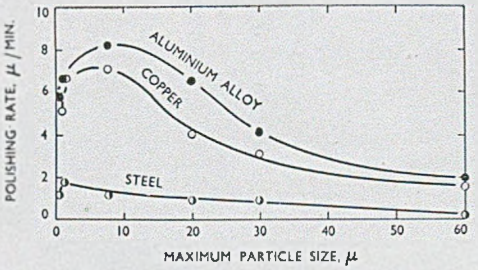

FIG. 3.—Variation in Polishing Rate with Abrasive Particle Size. Abrasive added as impregnated paste; abrasive weight 40 mg.

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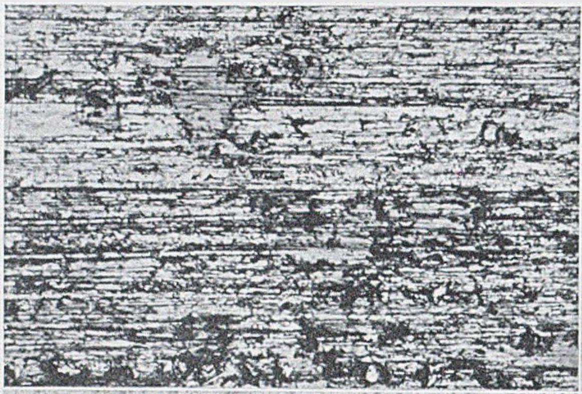

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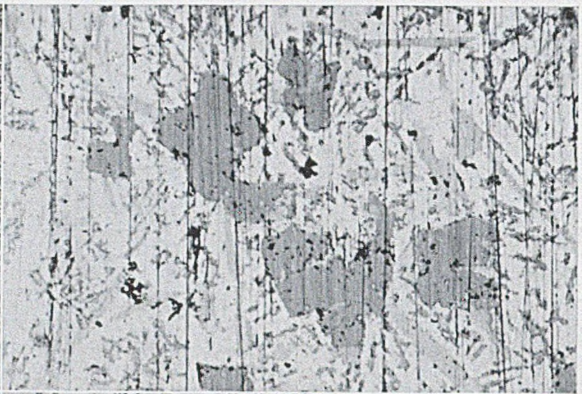

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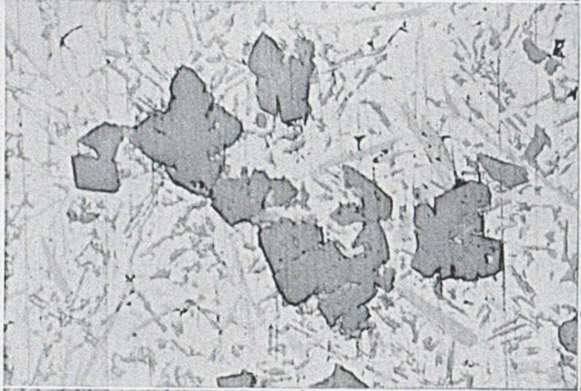

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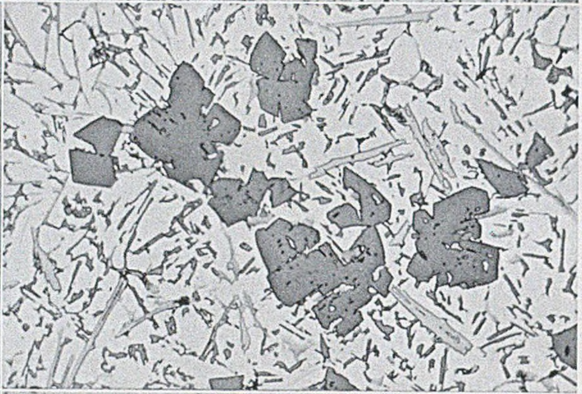

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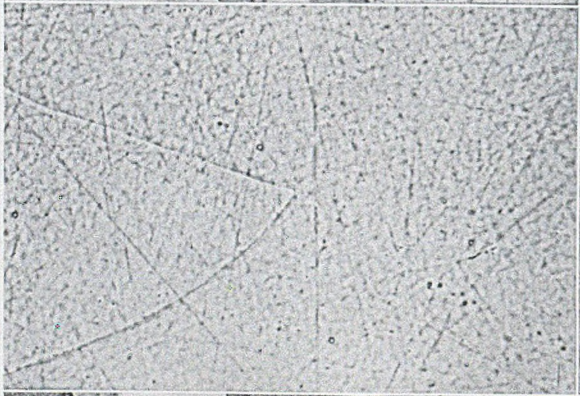

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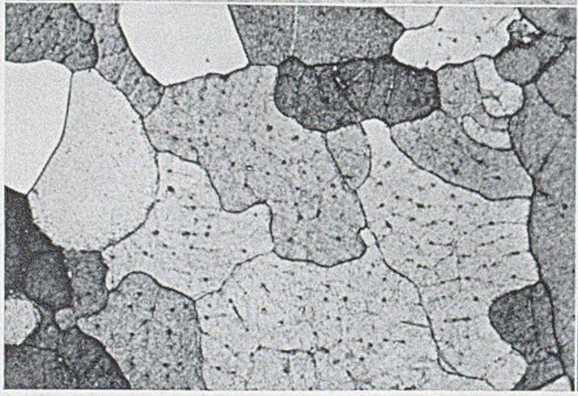

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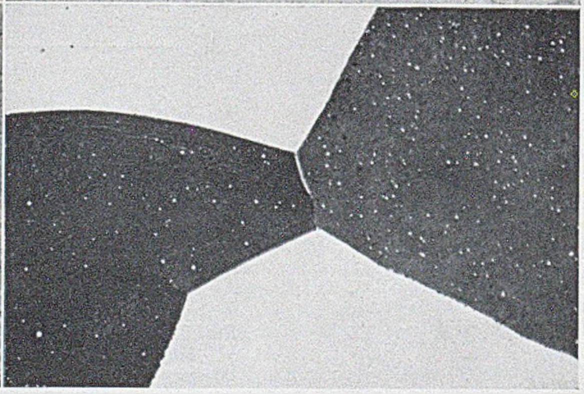


FIGS. 4–7.—Aluminium–13% Silicon Alloy at Major Stages of Polishing Process. × 250.

FIG. 4.—400-Grade silicon carbide paper.
FIG. 5.—Alumina wax lap.
FIG. 6.—(0–1) μ-grade diamond pad.
FIG. 7.—Magnesium oxide pad, etched 0·5% HF solution.

FIG. 8.—Super-Purity Aluminium. Normal magnesium-oxide pad. Phase-contrast illumination. × 100.
FIG. 9.—As for Fig. 8. Finish-polished by skidding technique.
FIG. 10.—Tin, Etched in 5% Ammonium Persulphate. × 250.
FIG. 11.—Annealed Zinc, under Polarized Light. × 100.

Mark	Primer	Top Coat	Before Cleaning	After Cleaning
g	Zinc chromate	D.T.D. 63A		
l	Zinc chromate	D.T.D. 260A		
h	Zinc tetroxy-chromate	D.T.D. 63A		
m	Zinc tetroxy-chromate	D.T.D. 260A		
j	Red lead	D.T.D. 63A		
n	Red lead	D.T.D. 260A		
k	Iron oxide	D.T.D. 63A		
o	Iron oxide	D.T.D. 260A		

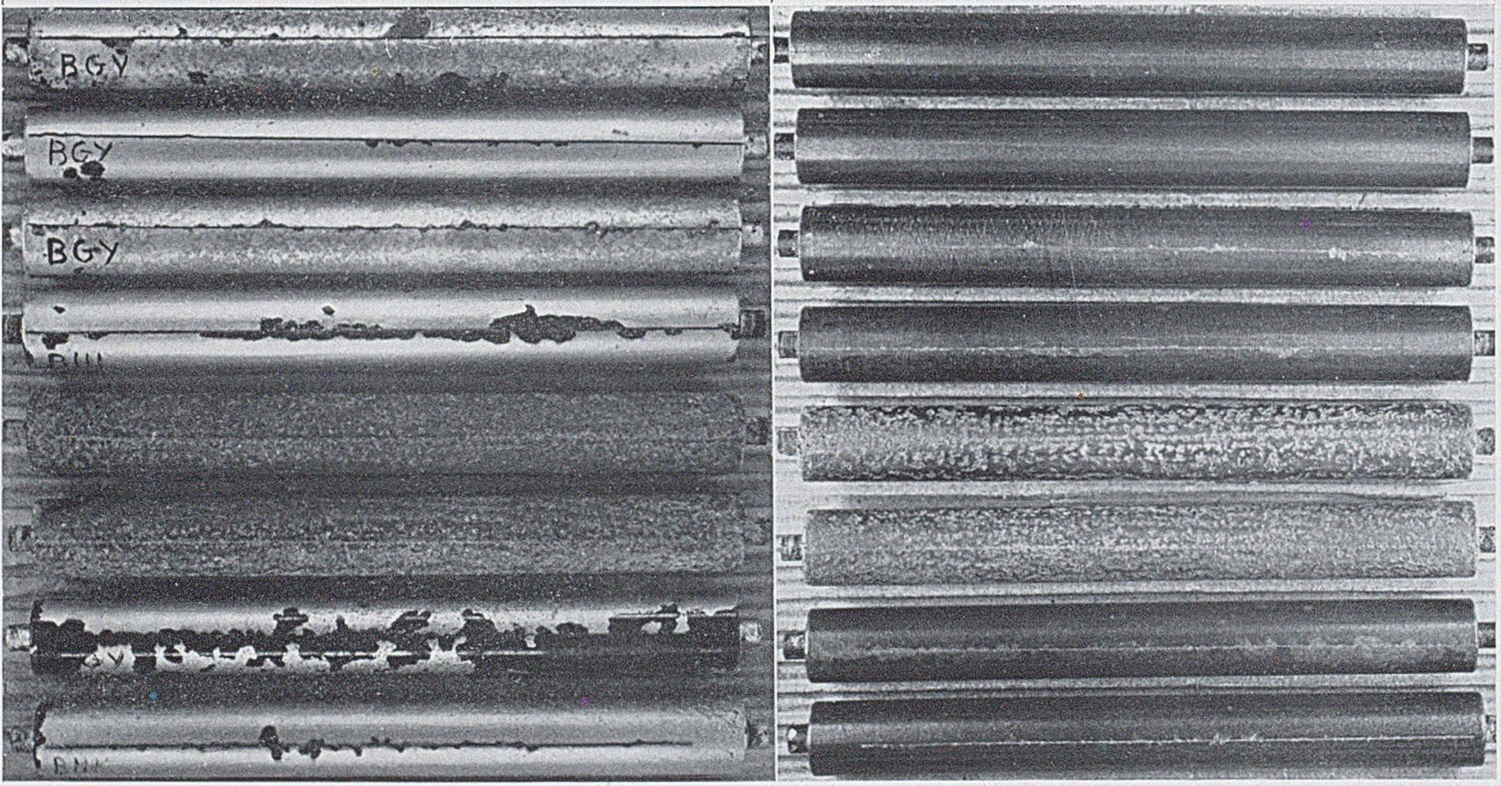
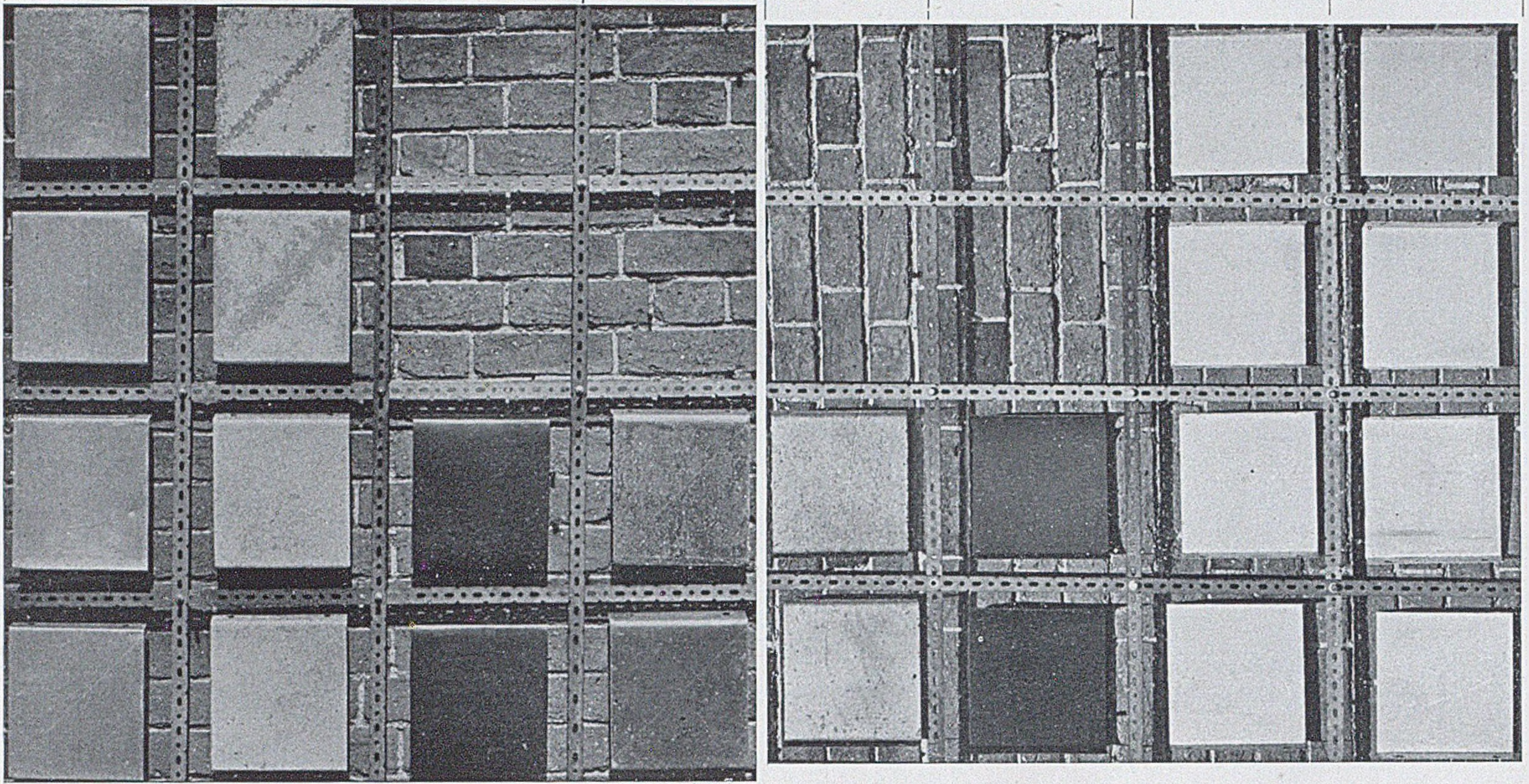


Fig. 1.—Cast Bars of High-Purity Magnesium Alloy (D.T.D. 59A) after 3½ Years' Exposure at the Marine Site.

Mark	Primer	Top Coat	Aluminium (Alloys)		Magnesium Alloys		Magnesium Alloys		Aluminium (Alloys)	
			Pure Aluminium	Alloy B.S.S. 5L3	High Purity D.T.D. 120A	Normal Purity D.T.D. 118	Normal Purity D.T.D. 118	High Purity D.T.D. 120A	Alloy B.S.S. 5L3	Pure Aluminium
j	Red lead	D.T.D. 63A								
n	Red lead	D.T.D. 260A								
k	Iron oxide	D.T.D. 63A								
o	Iron oxide	D.T.D. 260A								

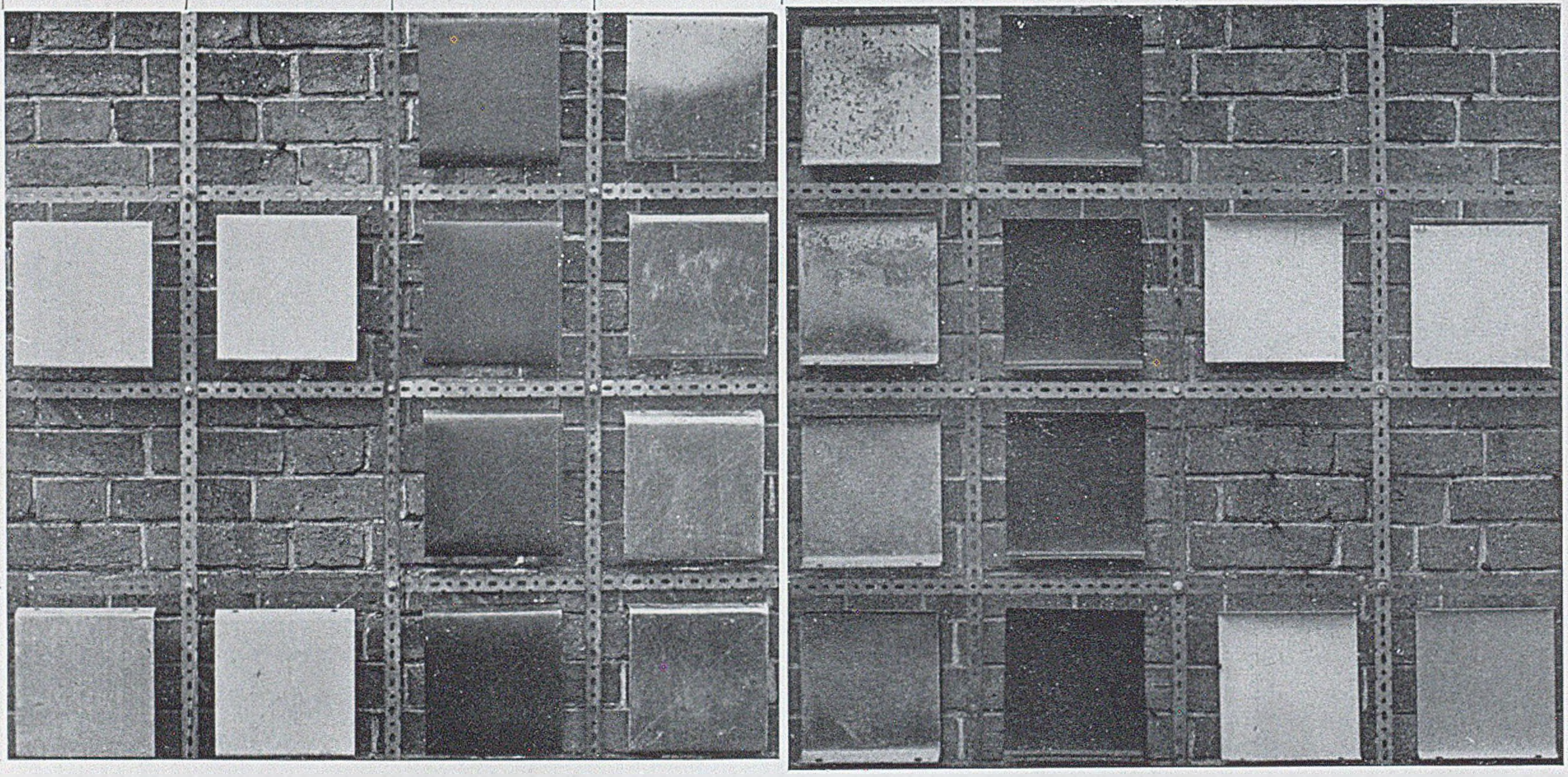


Fronts

Backs

Fig. 2.—Cleaned Panels after 3½ Years' Exposure at the Marine Site.

Mark	Primer	Top Coat	Aluminium (Alloys)		Magnesium Alloys		Magnesium Alloys		Aluminium (Alloys)	
			Pure Aluminium	Alloy B.S.S. 5L3	High Purity D.T.D. 120A	Normal Purity D.T.D. 118	Normal Purity D.T.D. 118	High Purity D.T.D. 120A	Alloy B.S.S. 5L3	Pure Aluminium
g	Zinc chromate	D.T.D. 63A								
l	Zinc chromate	D.T.D. 260A								
h	Zinc tetroxy-chromate	D.T.D. 63A								
m	Zinc tetroxy-chromate	D.T.D. 260A								

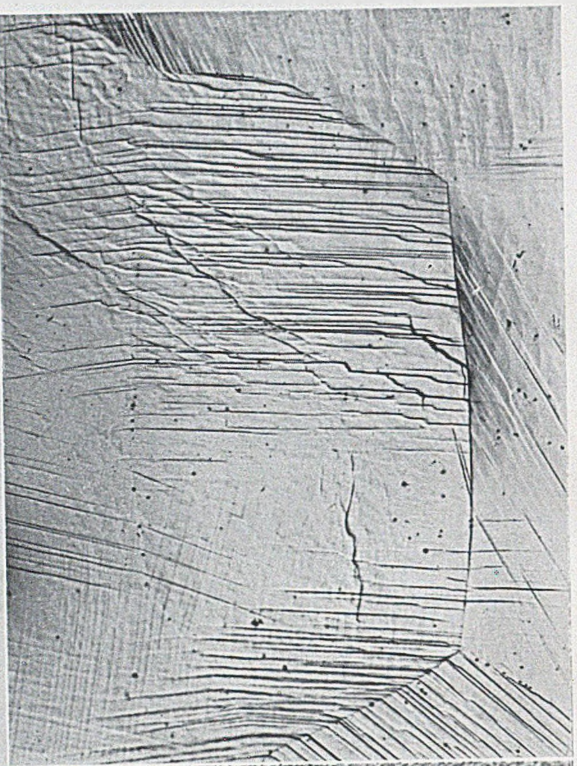


Fronts

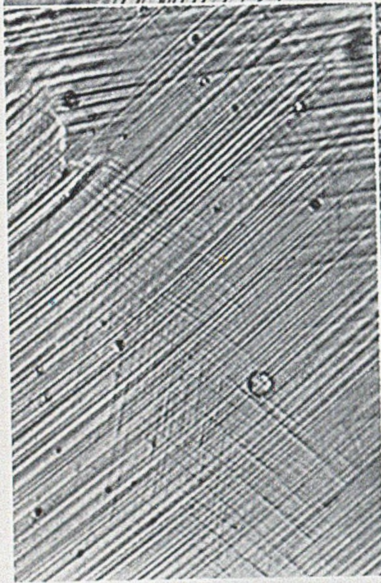
Backs

Fig. 3.—Cleaned Panels after 3½ Years' Exposure at the Marine Site.

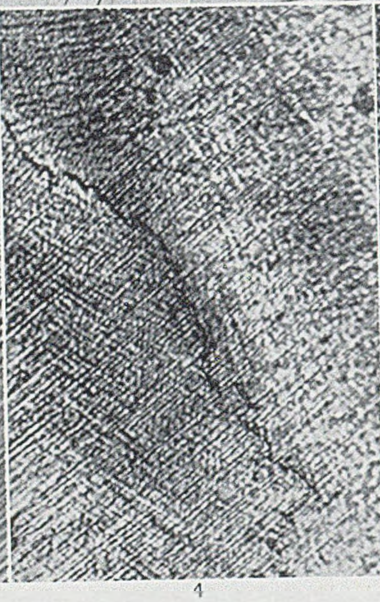
1



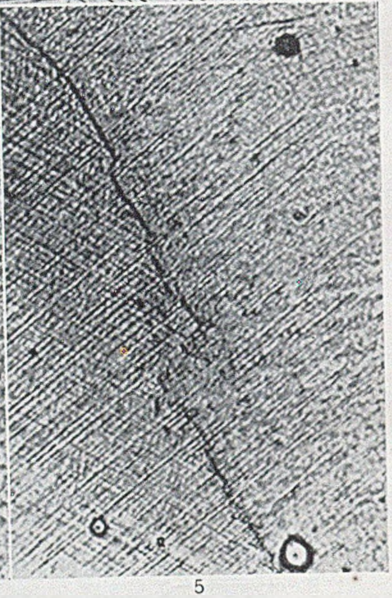
2



3



4



5

FIG. 1.—Slip Lines Crossing Grain Boundaries in Pure Aluminium. × 200.
FIG. 2.—Grain of Pure Aluminium with Its Boundary Divided into a Number of Nearly Straight Sections. × 150.
FIGS. 3–5.—Grain-Boundary Region in α-Brass. × 600.

FIG. 3.—Original surface.
FIG. 4.—Same field after electropolishing 10 microns from surface and then anodic etching in dilute $Na_2S_2O_3$.
FIG. 5.—Same field after polishing 40 microns from the surface.

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 20 JUNE 1953 PART 10

1 — PROPERTIES OF METALS

***The Influence of the Electrical Field on the Changes in Thin Films [of Aluminium and Silver].** Marcel Perrot and Jean-Pierre David (*Compt. rend.*, 1952, **234**, (18), 1753–1755).—Cf. *ibid.*, (11), 1133; *M.A.*, **20**, 4. Curves are presented showing the change with time of the elect. resistance of thin films of Al and Ag at a pressure of 10^{-6} mm. Hg when various potentials are applied. The resistance of Ag films (both granular and continuous) soon reaches a const. value, but the resistance of a granular Al film appears to increase continuously. For granular Ag films the resistance/time curve at high potentials always lies above the curve at low potentials; for continuous films, the reverse is the case.
—N. B. V.

***Combustion of Aluminium in Air.** René de Salins (*Compt. rend.*, 1952, **234**, (25), 2437–2439).—de S. has examined the spectrum emitted by powdered Al, burned in a gas flame. It consists of a continuum with superimposed lines due to Na, K, Al, Cu, Fe, and Mn, and bands due to Al oxides. Recording the radiation from the flame by means of a photoelect. cell showed that combustion of Al powder is not continuous but takes place in a series of short explosions lasting approx. 0·01 sec.—N. B. V.

***Solubility of Nitrogen in Liquid Chromium and Chromium–Silicon [Alloys].** V. S. Mozgovoi and A. M. Samarin (*Doklady Akad. Nauk S.S.S.R.*, 1950, **74**, (4), 729–732).—[In Russian]. The solubility of N was investigated by induction heating of 50-g. melts of commercial Cr under 1 atm. N. Temp. were measured and controlled by opt. pyrometry to ±15° C. The solubility falls with increasing temp., and it is concluded that the reaction $2Cr + \frac{1}{2}N_2 \rightleftharpoons Cr_2N$ takes place, for which $\Delta H = -7600$ cal. X-ray examination showed the existence of hexagonal Cr_2N ($a = 2{\cdot}74$, $c = 4{\cdot}45$ Å.). The presence of Si results in a decrease of N solubility at 1600° C. up to 10% Si, when an abrupt fall occurs, followed by a further gradual decrease; the heat of reaction for Si contents of 1·5–20% is substantially const. at $\Delta H = -10{,}980$ cal. A more detailed account appeared in *Izvest. Akad. Nauk S.S.S.R.*, **1950**, [Tekhn.], (10), 1529; *M.A.*, **20**, 385.—D. M. P.

***The Plastic Deformation of Copper Single Crystals. I.—Two Stages in the Progress of Deformation. II.—Recovery Process and Mean Length of Dislocations.** Hiroshi Kanzaki (*J. Phys. Soc. Japan*, 1951, **6**, (6), 454–455, 456–459).—[In English]. [I.—] In continuation of previous work (*ibid.*, (2), 90; *M.A.*, **19**, 760) the stress/strain curves for Cu single crystals, of purity 99·98%, have been determined at 20° C. When the resolved shear strain is ~20%, the stress increases rapidly with increasing strain, so that deformation progresses in two stages. The corresponding change of work-hardening is attributed to the two types of strains occurring in work-hardened metals discussed in the previous paper (*loc. cit.*). If this view is correct, then the number of stopped dislocations will show a rapid increase in the second stage of dislocation. Taylor (*Proc. Roy. Soc.*, 1934, [A], **145**, 388; *M.A.*, **1**, 379) has shown that in Al crystals the stress/strain diagrams correspond to the parabolic curve $s = at^2$, where s is the shear strain, t is the resolved shear stress, and a is a const., given by $a = (1/k^2G^2)(L/\lambda)$, k being a const., G the rigidity modulus, λ the slip distance produced by passage of one dislocation, and L the "mean free path" of a dislocation line. Using T.'s value $k = 0{\cdot}2$, it is found that $L = 3{\cdot}84 \times 10^{-4}$ cm. for the first stage of dislocation, and $L = 0{\cdot}68 \times 10^{-4}$ cm. for the second stage. [II.—] The anomalous sp. heat of Cu, due to recovery from strain-hardening has been measured at 0°–200° C. for single crystals of Cu deformed at −180°, −80° C., and room temp. The max. value of energy stored during deformation at room temp. is 0·1 cal./g. The energy released in the second valley of the sp.-heat anomaly, shown by the sp.-heat/temp. diagram, and associated with stopped dislocations, shows a rapid increase when the resolved strain >25%. The recovery temp. attains a max. value when the resolved strain is ~25%. Crystals deformed at −180° and −80° C. sometimes recover below room temp. The calculated sepn. distance of dislocations is 200 Å., which agrees extremely well with the value found by means of the electron microscope.
—J. S. G. T.

***The Plastic Deformation of a Copper Cylinder Under Powder Pressure.** J. Nishiwaki (*J. Phys. Soc. Japan*, 1951, **6**, (6), 538–539).—[In English]. The max. pressure, P_m, measured by a quartz piezo-indicator is found to be 1·13 times the pressure, P_c, indicated by the plastic deformation of a Cu cylinder, during the firing of a machine gun, for values of P_c in the region 30 kg./mm.² This agrees with the value $P_m/P_c = 1{\cdot}14$, generally adopted in internal ballistics. For low values of P_c, $P_m < P_c$, as here the Cu cylinder is more compressed under the firing pressure, which increases rather slowly in the gun, than under static pressure, though the max. values of both are equal.—J. S. G. T.

***Cold-Work Studies on Copper at Low Temperatures.** R. R. Eggleston (*J. Appl. Physics*, 1952, **23**, (12), 1400–1401).—A letter. Cu wire was intensely twisted at 4·2° K. and annealed at successively higher temp. for 2 min., and the resistance measured at 4·2° K. between anneals. The slope of the resistance/annealing-temp. curve showed two max. This is taken to correspond to two types of recovery, at temp. < and > −70° C. Isothermal curves obtained by a "pulse annealing" technique were taken, with measurements at 4·2° K. Activation energies in the ranges 8·7–11·6 and 13·6–17·6 kg.cal./mole were obtained for the lower and upper temp. ranges of annealing, resp. It is believed that the spread in values is due to experimental uncertainties.
—R. W. C.

***Measurement of the Heat of Recrystallization of Thin Copper Wires.** Gerhard Bockstiegel and Kurt Lücke (*Z. Metallkunde*, 1951, **42**, (8), 225–230).—Three new differential techniques—employing direct heating, indirect heating, and adiabatic methods—which, together with the apparatus used,

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

are described in detail, have been used to determine the heat of recrystn. of commercially produced Cu wires. The results are : (1) wire drawn 99·7% to 0·2 mm. dia., 2·7 cal./g., and (2) wire drawn 99·65% to 0·71 mm. dia., 2·5 cal./g., a value which, when this wire is twisted to fracture, becomes 1·0 cal./g. 5 ref.—E. N.

***Study of the Internal Stress in Electrodeposited Copper.** (Nishihara and Tsuda). See col. 712.

***The Magneto-Effect in Germanium Point Contacts.** Masao Tomura and Yukiaki Abiko (*J. Phys. Soc. Japan*, 1952, **7**, (1), 115–117).—[In English]. Attention is directed to a new amplifying phenomenon which occurs when an *n*-type Ge detector point contact is suitably oriented in a magnetic field, H. Values of dI/I vary approx. linearly with H and decrease abruptly when the forward bias voltage becomes $<\sim 0{\cdot}5$ V. The phenomenon is distinct from the magneto-resistance effect in Ge, for in this case $\Delta I/I \propto H^2$, and moreover the max. value of $\Delta I/I$ in the present case is $\sim$10 times $>$ the magneto-resistance effect (see Estermann and Foner, *Phys. Rev.*, 1950, [ii], **79**, 365 ; *M.A.*, **18**, 419).—J. S. G. T.

***On the Thermal Changes of Thin Gold Films and the Transformation Points of A. Féry.** Antoine Colombani and Gaston Rane (*Compt. rend.*, 1952, **234**, (18), 1757–1759).—C. and R. have determined the elect. resistance of thin (2000–10,000 Å.) Au films at various temp. in order to see whether a series of transformation points are obtained, similar to those observed by Féry on Pt black (*Ann. Physique*, 1933, **19**, 305 ; *J. Phys. Radium*, 1933, **4**, 301 ; *M.A.*, **1**, 66). With films >5000 Å. thick a decrease in resistance takes place as the film is slowly heated up to 140° C., corresponding to a crystn. process ; above 140° C. the resistance increases uniformly. For films 2000–5000 Å. thick the behaviour is similar to that of Pt as reported by Féry : the resistance undergoes a series of sudden isothermal decreases at 53°, 70°, 85°, 120°, and 270° C., and only above 270° C. do the properties become the same as those of ordinary Au. The sudden changes of resistance correspond to successive crystn., although it is impossible to speak of transformation points in Au as one can in Pt ; the much larger changes of resistance in the latter are attributed to degassing.
—N. B. V.

***The Temperature-Dependence of the Magneto-Resistance Effect in an Iron Single Crystal.** Yasuo Gondo and Zenya Funatogawa (*J. Phys. Soc. Japan*, 1952, **7**, (1), 41–43).—[In English]. Changes in the elect. resistance, R, of rods of Fe single crystals at temp. up to 800° C. were measured in magnetic fields of strength up to 1800 Oe. For one of the rods $\Delta R/R$ was const. from room temp. up to 300° C. and then decreased gradually to 0 at the Curie point with increasing temp. For another rod, $\Delta R/R$ decreased steadily to 0 at the Curie point. The large decrease of resistance at the Curie point was not anisotropic.—J. S. G. T.

***On the Flow-Dynamics of Molten Metals. I.—On the Discharge Coefficient of Mercury Through Small Circular Orifices.** Kiyoshi Yokata (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1950, (1), 20–25).—[In English]. The coeff. was found by measuring the change in elect. resistance of a wire immersed in Hg.—G. J. S.

Molybdenum: Our Most Promising Refractory Metal. I. I. Harwood (*Product Eng.*, 1952, **23**, (1), 121–132).—Methods of prodn. and fabrication, welding characteristics, and mech., phys., and chem. properties of Mo are described. Alloys of Mo showing promise are also discussed. 14 ref.
—M. A. H.

***Solubility and Diffusion Coefficient of Carbon in Nickel: Reaction Rates of Nickel–Carbon Alloys with Barium Oxide.** J. J. Lander, H. E. Kern, and A. L. Beach (*J. Appl. Physics*, 1952, **23**, (12), 1305–1309).—The solubility of C in Ni was determined by carburizing in a H_2/CH_4 mixture and analysing the Ni for C by combustion. The solubility (wt.-%) is given by : $\ln S = 2{\cdot}480 - 4880/T$. Diffusion coeff. D were measured by coating opposite sides of a Ni disc with C and NiO and heating it in a continuously evacuated space. The pressure of the CO and CO_2 evolved by reaction between NiO and C which had diffused through to it was measured, and D derived from it. Result : $\ln D = 0{\cdot}909 - 20{,}200/T$, with a mean deviation of $\sim$5%. The values of D are somewhat higher than older values. Reaction rates between dissolved C and BaO were found to be diffusion-controlled at most temp.—R. W. C.

***The Effect of Electrolytic Treatment of a Nickel [Diode] Sleeve on Electron Emission.** Junkichi Nakai and Shōgo Nakamura (*J. Phys. Soc. Japan*, 1951, **6**, (6), 532–533).—[In English]. Electrolytic polishing of a Ni diode sleeve is found to increase the cathode electron emission of the diode from 1·5 to 5 times.—J. S. G. T.

***Effect of Impurities on the Self-Diffusion of Silver.** R. E. Hoffman and D. Turnbull (*J. Appl. Physics*, 1952, **23**, (12), 1409–1410).—A letter. Addn. of small amounts of Pb to Ag (up to 1·30 at.-%) had no certainly significant effect on the activation energy for self-diffusion. At const. temp., however, a mole fraction x of Pb affects the diffusion coeff., D, according to the equation : $D = D_{Ag} + xD'$. $D' = 7{\cdot}6 \exp(-40{,}500/RT)$ cm.^{2} sec.$^{-1}$. It is shown that this influence cannot be due to reduction of the cohesion of the Ag. It is attributed to a " lead-catalysed " transport of Ag, perhaps due to a direct interchange of Pb and Ag atoms.
—R. W. C.

***The Transformation of Tin at Low Temperatures.** Haruji Ishikawa (*J. Phys. Soc. Japan*, 1951, **6**, (6), 531–532).—[In English]. White Sn, if not extremely pure, is easily transformed into grey Sn when pressed into sheet form at a sufficiently low temp., e.g. −20° C., and exposed to a low temp., e.g. −10° to ~ -20° C., for a number of days. The transformation occurs more readily, the greater the deformation under pressure. The slower the compression, the faster the transformation. Organic solvents do not considerably affect the transformability, but the presence of Sn^{++} accelerates the growth of transformation nuclei. The most effective temp. for the transformation to occur is ~ -70° C.
—J. S. G. T.

***Measurements on the Formation of Thin Tin Films.** A. van Itterbeek, L. de Greve, and F. Heremans (*Appl. Sci. Research*, 1952, [B], **2**, (5), 352–360).—[In English.] The sp. resistance of sputtered Sn films was determined as a function of thickness, and the change of sp. resistance was measured during the formation of the films. When the films were sputtered in H, the resistance was infinite owing to the agglomeration of Sn into separate grains. However, a finite resistance was obtained in He, Ne, and A, providing that the film thickness exceeded a certain min. value. The sp. resistance decreased slowly at first, then sharply with increasing thickness obtained by successive sputterings. The increase in the rate of decrease is explained by the filling up of the voids between the small Sn grains after a certain number of sputtering operations. The change of resistance of the Sn films was also observed when they were brought into contact with O. The greatest increase in the resistance occurred during the first few sec., and the rate of increase depended on the pressure. The homogeneity of the films was examined by means of electron microscopy.—S. Ma.

Advances in Titanium Metallurgy in the United States of America. [—I.] H. O. Nicolaus (*Metall*, 1952, **6**, (1/2), 1–5).—A review of developments, particularly during the last five years. The compn., phys. properties, origin, and principal deposits of Ti ores are tabulated. The prodn. of Ti is described, with particular ref. to the Kroll process. Methods for the prodn. of ingot metal from powder metal, by induction melting, and by arc melting are described and illustrated, as is the continuous process for the prodn. of ductile Ti developed by the Battelle Memorial Inst. (cf. Maddex and Eastwood, *Trans. Amer. Inst. Min. Met. Eng.*, 1950, **188**, 634 ; *M.A.*, **18**, 246). 10 ref.—E. N.

Isotopic Composition and Atomic Weight of Natural Uranium. Joseph J. Katz and Eugene Rabinowitch (*The Chemistry of Uranium. I.—The Element, Its Binary and Related Compounds* (*National Nuclear Energy Series*), 1951, **VIII-5**, 3–11).—The 3 natural isotopes of U—U^{238} (UI), U^{234}(UII), and U^{235}(AcU)—are present in the proportions 99·28%, 0·005%, and 0·71%, resp. The ratio $U^{238} : U^{235} =$

138·0 ± 0·3 is remarkably const. and shows no great variation ($\not> 0·03\%$) between ores from different localities. The most recent detn. of the at. wt. of U as 238·076 (Hönigschmid and Wittner, *Z. anorg. Chem.*, 1936, **226**, 289; *M.A.*, **3**, 143) is probably too high, since this gives a larger packing fraction to U^{238} than expected for an element of this at. wt. 12 ref.
—P. E. M.

†**Physical Properties of Uranium Metal.** Joseph J. Katz and Eugene Rabinowitch (*The Chemistry of Uranium. I.—The Element, Its Binary and Related Compounds* (*National Nuclear Energy Series*), 1951, **VIII-5**, 133–163).—U exists in 3 allotropic forms: (1) α, stable below $\sim$665° C., is orthorhombic with $a = 2·8482$, $b = 5·8565$, and $c = 4·9476$ Å. and calculated $d = 19·050$ g./c.c.; (2) β is stable between $\sim$665° and 770° C. [Tucker (*Acta Cryst.*, 1951, **4**, 425; *M.A.*, **19**, 597) has reported its structure as tetragonal with $a = 10·52$ and $c = 5·57$ Å.]; (3) γ, stable above $\sim$770° C., is b.c.c. with $a = 3·43$ Å. The transformations tend to occur over a range of a few degrees, with considerable hysteresis on cooling. The heats and entropies of the $\alpha \rightleftharpoons \beta$ and $\beta \rightleftharpoons \gamma$ transformations are 665 and 1170 cal./g.-atom and 0·71 and 0·98 e.u., resp. The m.p. of U is 1133° ± 2° C., with a heat of fusion 2·5–3·0 kg.cal./g.-atom and entropy of fusion of 2 e.u. The sp. heats, energies, and entropies of the 3 phases and the energy of vaporization are discussed. The average coeff. of linear thermal expansion ∥ to [100], [010], and [001] were determined by X-rays to be 23 ± 3, $-3·5 \pm 2$, $17 \pm 2 \times 10^{-6}$ cm./cm./°C. between 25° and 300° C., and 28 ± 2, $-1·4 \pm 1$, and $22 \pm 1 \times 10^{-6}$ cm./cm./° C. between 25° and 650° C. Vol. expansion coeff. of 37 and 44×10^{-6} c.c./c.c./°C. were determined by X-rays and dilatometer, resp. Cast U has 200–220 B.H.N. or 80–100 Rockwell A. Cold rolling or swaging may increase the hardness to 115 Rockwell A with most of the hardening taking place in the first 20% reduction in thickness. Softening begins on annealing to 150° C., but complete softening is not obtained on annealing below 600° C. Quenching U heated for 5 hr. at 900° C. produces a hardness of 115 Rockwell A, which has been attributed to the presence of carbide. The hardness changes little with temp. below $\sim$200° C. and then falls to $\sim$13 B.H.N. at 650° C. The hardness increases to 30–40 B.H.N. in the β phase, but becomes too low to measure in the γ field. Values of Young's modulus between 5 and 25×10^6 lb./in.^{2} are obtained from stress/strain curves and 29·8 from a dynamic method. Young's modulus decreases rapidly with temp., but increases in the β range. Shear moduli between 6·6 and $12·0 \times 10^6$ lb./in.^{2}, bulk moduli from 9 (cold-worked) to 15 (annealed) $\times 10^6$ lb./in.^{2} and Poisson's ratios between 0·20 (cold worked) and 0·49 (β annealed) have been reported. The 0·2% P.S. varies from 8 to 76×10^3 lb./in.^{2}; the highest values being obtained in worked or γ-quenched samples. The yield stress decreases rapidly with temp., e.g. from 43×10^3 at room temp. to $8·9 \times 10^3$ lb./in.^{2} at 600° C. Definite Y.P. (reported as 18×10^3 lb./in.^{2} at 700° C. and 13×10^3 lb./in.^{2} at 750° C.) occur in β uranium. U.T.S. values from 170,000 to 200,000 lb./in.^{2} are obtained in cold-worked U. Annealing at 625° C. reduces this to 135,000 lb./in.^{2} and at 830° C. to 65,000 lb./in.^{2}. Cast, γ-annealed, or extruded specimens have U.T.S. values from 65,000 to 90,000 lb./in.^{2}, but quenching from the 600°–1000° C. range increases this to 90–130,000 lb./in.^{2}. The U.T.S. decreases rapidly with temp., e.g. from 53×10^3 lb./in.^{2} at room temp. to 12×10^3 lb./in.^{2} at 600° C. Elongations of 0–3% and 5–10% and reductions in area of 0–8% and 3–15% are obtained for cold-worked and γ-extruded U resp. Greater ductility (up to 20%) is obtained in α-worked and annealed specimens or in specimens quenched from high α temp. Quenching from high β or γ temp. reduces ductility. Creep, fatigue, impact, and rupture tests on U are briefly described. Elect. conductivities between 2 and 4×10^1 (ohm-cm.)$^{-1}$ and temp. coeff. of resistivity from 2·0 to $2·78 \times 10^{-3}$ ohm-cm./°C. have been reported. U is a supraconductor below 0·75°–1·3° C. Thermal conductivities of $60–65 \times 10^{-3}$ cal./cm./sec. and temp. coeff. of thermal conductivity of $1·5 \times 10^{-3}$/°C. from 100° to 225° C. and $0·4 \times 10^{-3}$/°C. from 225° to 450° C. are reported. The thermo-elect. potential, work-function, magnetic susceptibility, and opt. emissivity of U are briefly discussed. 120 ref.
—P. E. M.

†**Chemical Properties of Uranium.** Joseph J. Katz and Eugene Rabinowitch (*The Chemistry of Uranium. I.—The Element, Its Binary and Related Compounds* (*National Nuclear Energy Series*), 1951, **VIII-5**, 164–173).—The reactions of U with the non-metallic elements, common gases, acids and alkalies, metal and salt soln., and miscellaneous compounds are summarized. U is highly reactive, a strong reducing agent, and probably close to Be in the electromotive series. U combines with all the non-metallic elements except the noble gases. Boiling water attacks it slowly, but the corrosion is speeded if H is present. Steam attacks U much more vigorously than does O. Reactions occur with the H halide gases, chlorinated hydrocarbon vapours, CO, CO_2, NH_3, NO, and CH_4 at elevated temp. Most common acids dissolve U, except HF where insoluble UF_4 tends to stop the reaction. Simple organic acids do not attack U unless HCl is present, in which case reaction is rapid. U displaces Hg^{++}, Ag^+, Cu^{++}, Sn^{++}, Pt^{++++}, and Au^{+++} from soln. of their salts. Alkali metal hydroxides do not attack U, but in the presence of H_2O_2 soluble peruranates are formed. U reacts with soln. of $K_2S_2O_8$ and $(NH_4)_2S_2O_8$. U reduces UF_4 and UO_2 at elevated temp. U attacks glass and silica between 700° and 800° C., but it is inert to fusion with boric acid. 39 ref.—P. E. M.

***[The Effect of] Nitrogen in Metals.** David O. Caldwell (*Rev. Sci. Instruments*, 1952, **23**, (9), 501–502).—By using a synchrocyclotron, activities were found in Mn, Fe, Nb, Rh, and Ta which were due to the presence of N in these metals. The results are discussed.—E. J.

The Influence of High Speeds of Deformation on the Mechanical Properties of Metals. Yves Pironneau (*Tech. Sci. aéronaut.*, 1952, (4), 224–241).—A review, with 21 ref.
—H. A. H.

***Delayed Fracture of Metals under Static Load.** N. J. Petch and P. Stables (*Nature*, 1952, **169**, (4307), 842–843).—It is suggested that metals should show an effect similar to the delayed fracture obtained with glass, which has been explained by Orowan (*ibid.*, 1944, **154**, 341), and that the H embrittlement of Fe may be an example. If so, there should be a time-delay for the development of the embrittlement while the H in soln. diffused to the Griffith crack. Such a delay occurs, as no embrittlement is detected at rapid rates of loading.—R. S. B.

***Mechanical Anisotropy in Some Ductile Metals.** W. A. Backofen and B. B. Hundy (*J. Inst. Metals*, 1952–53, **81**, (9), 433–438).—Fracturing test-specimens in tension after prestraining in torsion has shown that a fibrous crack-like structure, causing a considerable degree of mech. anisotropy, exists in 70 : 30 brass, Ni, Monel metal, and Armco Fe. The same programme of testing has also revealed the presence of such a structure in high-purity Al, but, for reasons that are not clear, the tensile behaviour of torsionally prestrained commercially pure (2S) Al gives no indication of its presence.—AUTHORS.

Fundamental Experiments on the Rupture of Solids. A. Guinier (*Métaux, Corrosion-Ind.*, 1952, **27**, (320), 150–155).—G. describes the methods of calculating and measuring the cohesive strength of solids and the various attempts to account for the discrepancy between calculations and experimental results. He proposes experiments to detect by X-rays the presence of sub-microscopic internal voids in metals. 15 ref.—M. A. H.

Creep of Metals and Methods of Measuring It. E. Morlet (*Métaux, Corrosion-Ind.*, 1952, **27**, (319), 107–118).—M. reviews the various types of mech. deformation, their relation to time and temp., the various fundamental ways of studying these relationships (creep, stress relaxation, &c.), and methods of representing creep results. The effects of various factors (mech. and thermal treatments, structural factors) on resistance to creep are briefly discussed. 17 ref.—M. A. H.

Reflections on Creep. A. Fotiadi (*Métaux, Corrosion-Ind.*, 1952, **27**, (318), 69–74).—F. decries the appn. of classical criteria such as P.S., total elongation, and impact bend value to materials and components of gas turbines, as inadequate and hindering development. He shows how one may learn from past experience, with metals used at normal temp., how to tackle the development of materials and components for high-temp. service. Various types of creep and fatigue tests are suggested, which are aimed at simulating basic types of service conditions.—M. A. H.

A Note on the Mathematical Analysis of Creep Curves. L. M. T. Hopkin (*J. Inst. Metals*, 1952–53, **81**, (9), 443–448).—The Andrade equation $L_t = L_0(1 + \beta t^{1/3})\exp Kt$ and an equation contg. a logarithmic term have been found to give good, but not perfect, fits of creep curves, obtained from const.-stress tests on a high-purity Pb and a Pb–1% Sn alloy, for times up to 300 hr., although the Andrade equation was the better for longer times up to 5000 hr. An attempt to correlate the values of the const. in the Andrade equation with the effects of grain-size and stress on creep behaviour was not possible for the high-purity Pb owing to relatively large errors involved in the calculation of the const. For the Pb–1% Sn alloy, however, it was found that the $\sqrt{K}$ was linearly related to stress and relative grain-boundary areas, although no simple relationship existed between these variables and the β const.—AUTHOR.

†**The Mechanism of Fatigue of Metals.** A. K. Head (*J. Mechanics Physics Solids*, 1953, **1**, (2), 134–141).—A review. After a brief discussion of the form of fatigue curves, H. discusses the stages of deformation which arise during a fatigue test. In the first stage, general work-hardening takes place, in the second the crack nucleus develops, and in the third the crack propagates. The second stage is considered the most important, but a review of published theories shows that none accounts satisfactorily for phenomena observed during this stage of the test. 48 ref.—E. O. H.

***Torsional Fatigue Failures.—II.** J. O. Almen (*Product Eng.*, 1952, **23**, (3), 168–174).—Cf. *ibid.*, 1951, **22**, (9), 167: *M.A.*, **20**, 390. The stresses in torsion bar springs subjected to shot peening and prestressing treatments are analysed. Failure is shown to result from the tensile component of the applied stress. The effects of processing variations have been studied experimentally.—M. A. H.

***Magnetic Properties, Internal Strains, and the Bauschinger Effect in Metals.** D. V. Wilson (*Nature*, 1952, **170**, (4314), 30–31).—Experiments on heat-treated 0·88% C steel show that the change in indentation hardness, which results from a moderate degree of deformation is dependent on the microstructure, and it is suggested that the "work-softening" effects discussed by Polakowski (*ibid.*, 1951, **168**, 838; *M.A.*, **19**, 698) may depend on the intensity, scale, and directional characteristics of the initial textural stress patterns.

—R. S. B.

Amplitude Pulsations in the Vibrational Strain Pattern of Metal Single Crystals. J. W. Marx (*J. Appl. Physics*, 1952, **23**, (12), 1406–1407).—A letter. Takahashi (*ibid.*, (8), 866; *M.A.*, **20**, 389) has shown that for large amplitudes, a Zn crystal subjected to forced oscillations near a resonant frequency exhibits a slow periodic alteration of peak amplitude, called "gasping". T. attributed this behaviour to discontinuous changes in the elastic const. with change of peak amplitude, owing to slip-line formation. M. shows that in principle a continuous amplitude dependence of the elastic moduli can lead to the same result.—R. W. C.

The Atomic Theory of Elasticity. G. Leibfried (*Z. Physik*, 1951, **129**, (3), 307–316).—Math. The elastic equations are deduced from the theory of slow atomic displacements, both for a primitive lattice and for the general case.—J. W. C.

Liberation of Heat During Plastic Deformation. N. S. Fastov (*Doklady Akad. Nauk S.S.S.R.*, 1952, **83**, (6), 851–854).—[In Russian]. Math. F. obtains equations for the external work done on a specimen, and the amount of heat liberated; from these an expression for the max. energy stored is obtained. This gives values of the order of 1 cal./g., in agreement with the experimental data.—G. V. E. T.

***On Plastic Deformation Process of Compression of Metals under High Speed.** Hiroshi Yamanouchi and Yuji Matsuura (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1951, (2), 7–10).—Y. and M. describe the use of a high-speed camera (1/38,000–1/76,000 sec.) to measure the deformation of a Pb test-piece under drop hammers of various weights. The results are subjected to a math. treatment.—G. J. S.

War on Wear. H. Blok (*Engineering*, 1952, **173**, (4502), 594; (4503), 625–626).—Paper read at the International Symposium on Abrasion and Wear, Delft, Nov. 1951. Of general interest, covering the different types of wear encountered between mating metal surfaces.—D. K. W.

***On the Change in the Electrical Conductivity of Metals at the Melting Point.—I.** (Mlle) Geneviève Sutra (*Compt. rend.*, 1952, **234**, (27), 2589–2591).—S. demonstrates, using published data, that for a number of metals there is a close correlation between the change of elect. resistance and the change of vol. at the m.p. Hg seems to be an exception.

—N. B. V.

***On the Change in the Electrical Conductivity of Metals at the Melting Point.—II.** (Mlle) Geneviève Sutra (*Compt. rend.*, 1952, **235**, (14), 707–709).—Cf. preceding abstract. The discontinuity of the elect. conductivity of metals at the m.p. previously noted is explained by supposing that the conductivity electrons of the metals have not the total vol. but a lesser vol. at their disposition.—J. H. W.

***Normal Resistivities [of Alkali and Alkaline-Earth Metals] at Low Temperatures.** K. Mendelssohn ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 253–256).—The elect. resistivities, r, of Na, K, Li, Cs, Rb, Be, Mg, and some of the alkaline-earth metals (some impurities present in these last) have been determined at 4°–20° K. According to Grüneisen's formula and Bloch's one-electron theory, $r \propto T^5$, where T denotes the abs. temp. The only metal of the two groups that shows complete agreement with Bloch's theory is Na; the power index of T actually observed for the different specimens is 4·85, but the T^5-law represents a limiting case for abs. zero, and at finite temp. the effect of lattice vibrations must decrease the index. For Li the index has the value 4·55. There is wide disagreement in the values of θ, the Debye characteristic temp., found by different methods for the same metal. Rb and Cs do not show agreement with the theory nor do the alkaline-earth metals. In all these cases the concept of quasi-free electrons does not seem to be applicable as in Na. All four samples of Mg used showed a min. of r similar to that found in Au. In some specimens of K, Cs, and Ba the resistance/temp. curves showed small kinks, probably due to impurities.—J. S. G. T.

***Current Sensitivity and Other Characteristics of Metal Films at Low Temperatures.** A. van Itterbeek ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 243–247).—At low temp. (~5°–50° K.) thin Ni films possess a negative temp. coeff. of resistance; moreover, the resistance decreases with increase of current. The same is true of thin Fe films and for very thin films of Pt, W, and Nb. Thin films of Ni and Fe, examined by the electron microscope, are found to have an amorphous structure. In thin films of Au, Ag, and Sn the grains are larger and probably separated by a layer a few mol. thick; the electron diffraction rings of these films are very sharp compared with those of Fe films. The thinner the Nb film the lower the temp. of transition from a negative temp. coeff. of resistance to a positive coeff.—J. S. G. T.

***New Experiments on the Supraconductivity of Metal-Layers Condensed at Low Temperatures.** Rudolf Hilsch ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 45).—Transition temp. of Sn, In, Tl, Al, Pb, and Hg condensed at He temp. from the vapour state are ~1° K. different from the values obtained for the resp. metals in bulk. The deviations appear to have some connection with Debye characteristic temp. The normal transition temp. reappears when the condensed films are annealed to room temp. It is considered that lattice distortions are responsible for the abnormal transition temp. A new method for producing alloys by simultaneous condensation of different metals has been worked out. The Sn–Cu system, produced in this atomic-disperse state has been studied. A transition temp. of 7° K. has been attained. The normal transition temp. of 3·7° K. for Sn is reduced to 1·5° K. if only 1 part Cr in 10^4 is added by simultaneous con-

densation at low temp. It has been possible to realize a content of 10% *Farbzentren* in the KCl + K system.
—J. S. G. T.

***The High-Frequency Resistance of Metals in the Normal and Supraconducting State.** C. J. Grebenkemper and John P. Hagen (*Phys. Rev.*, 1952, [ii], **86**, (5), 673–679).—The H.F. resistance of Pb, In, and Sn has been measured at a frequency of ~9000 Mc./s., using resonant-cavity methods. Preliminary results with Sn at 24,000 Mc./s. are also reported. Surface finish had a significant effect on the surface conductivity of both supraconducting and normal-conducting material. The H.F. resistance of Sn in the supraconducting state seems to vary with frequency as $\omega^{3/2}$ rather than as ω^2 forecast by London (" Superfluids ", Vol. 1, p. 90. **1950** : New York).
—P. C. L. P.

***Eddy Currents and Supercurrents in Rotating Metal Spheres at Liquid-Helium Temperatures.** P. B. Alers, J. W. McWhirter, and C. F. Squire ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 85–88).—Love, Blunt, and A. (*Phys. Rev.*, 1949, [ii], **76**, 305; *M.A.*, **17**, 151) found that a solid supraconducting sphere, rotated rapidly about an axis in the normal conducting state, cooled through the supraconducting transition temp. while still rotating, and then stopped rotating, undergoes a perfect Meissner effect. This phenomenon is studied and discussed for spheres of both Sn and Ta, the effects studied and discussed including the ratio of eddy current : supercurrent as a function of the speed of rotation. Eddy currents have the same distribution as the supraconducting Meissner currents. Supraconducting currents maintain their intensity whether the sphere is rotating or not. A complete Meissner effect has been produced in a Ta sphere by cooling the rotating sphere *very slowly* into the supraconducting state, to avoid the prodn. of frozen-in flux.
—J. S. G. T.

†Electron-Vibration Interactions and Supraconductivity. J. Bardeen (*Rev. Modern Physics*, 1951, **23**, (3), 261–270).—Approx. methods of calculation of the interaction energy in theories of supraconductivity are discussed. Sections are devoted to Fröhlich's method, to an extension of the adiabatic method used by B., to an appn. of Tomonaga's one-dimensional theory of plasma oscillations, and to the use of a canonical transformation of the type used by Bloch and Nordsieck to eliminate the interaction term in radiation theory. No discussion is given of how typical supraconducting properties follow from the model.—D. P. M.

***Supraconductivity and Lattice Vibrations.** J. Bardeen ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 5–10).—The isotope effect, discovered by Maxwell (*Phys. Rev.*, 1950, [ii], **78**, 477; *M.A.*, **18**, 245) and by Reynolds, Serin, Wright, and Nesbitt (*ibid.*, p. 487; *M.A.*, **18**, 245) indicates that supraconductivity arises from interactions between electrons and lattice vibrations. B. has developed a theory of the supraconducting state based on the wave-functions and energies of individual electrons as affected by the lattice vibrations (*Rev. Modern Physics*, 1951, **23**, 261; preceding abstract). The nature of the wave-functions and the energies of electrons in supraconductors are now discussed, and it is shown how the theory leads to the London phenomenological equations and explains the electromagnetic properties of the supraconducting phase. Exact conditions under which supraconductivity occurs are still uncertain.—J. S. G. T.

***Nature of the Supraconducting Transition.** L. Tisza ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 11–20).—A partly phenomenological theory, affording a comprehensive qual. framework for the quantum mechanics of solids, is developed. A " fundamental assumption " of crystal physics, viz. that the lowest soln. of the many-centre problem of an extended system of nuclei and electrons is periodical, is introduced, and definite conditions for two kinds of two-phase transitions are deduced thereby; these conditions seem to be applicable to practically all transitions. The assumption that the conditions are relevant to the supraconducting state leads to a virtually unique electronic mechanism for supraconductivity. The picture incorporates the basic idea of the Fröhlich–Bardeen theory ascribing supraconductivity to electron–lattice interaction, the supraconducting state being associated with resonating superstructures, stabilized by closely coupled deformations of the lattice.—J. S. G. T.

Theory of Supraconductivity. O. Klein (*Nature*, 1952, **169**, (4301), 578–579).—K. calculates the probability for the emission of phonons by an electron coupled with sound waves and moving with a velocity greater than that of the waves, as is the case with electrons in metals with energies greater than the Fermi surface. Bloch's model will not give rise to supraconductivity, but if it is assumed that a small part of the electron *d* remains stationary during the vibrations, then it is shown that $T_c \propto M^{-\frac{1}{2}}$ and that the small excess positive charge carried by the moving atoms must be $\sim(0{\cdot}001\text{–}0{\cdot}01)e$.—R. S. B.

***Theory of the Supraconducting State.** H. Fröhlich ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 1–3).—The main problem in explaining the supraconducting state in metals is to find an interaction of the right magnitude. Bloch's theory of the interaction of electrons in crystals with lattice vibrations must lead to an interaction between electrons. The conventional theory of metals is based on the free-electron hypothesis, which has neither been derived theoretically nor stated in exact terms. F. shows, very briefly, how energy considerations lead to an interaction, between electrons in momentum space, which tends to shift electrons from the top of the Fermi distribution to higher values, can describe phenomenologically the magnetic behaviour of supraconductors, and leads to a selection of supraconductors in fair agreement with observations.—J. S. G. T.

Field Theoretic Description of Interactions in a Supraconductor. M. Dresden ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 21).—Tomonaga (*Progress of Theoretical Physics*, 1950, **5**, 544) has described the appn. of a non-ideal Fermi gas to problems in supraconductivity; it is now suggested that all interactions in a supraconductor can be described in terms of two interacting scalar Bose fields and that therefrom properties usually associated with the supraconducting state may be deduced.—J. S. G. T.

***A Note on the Theory of Semi-Conductors.** P. T. Landsberg (*Proc. Phys. Soc.*, 1952, [A], **65**, (8), 604–608).—L. analyses current theories of semi-conductivity, and groups them in two classes, depending on whether electron spin is taken into account or not. A detailed analysis is made of the first class, and this is shown to yield the more preferable approach, in particular in connection with the detn. of activation energies. Certain modifications to special cases, e.g. thermoelect. power and Hall effect formula, when the spin of donor electrons is considered, are also discussed.—E. O. H.

***Electron Spin in Semi-Conductors.** E. A. Guggenheim (*Proc. Phys. Soc.*, 1953, [A], **66**, (1), 121–122).—A letter. G. discusses a paper, by Landsberg (*ibid.*, 1952, [A], **65**, 604; preceding abstract) which dealt with two possible models of semi-conductors, where electron spin is accounted for or not. G. shows here that the " direct " approach, which was not preferred by L., can in fact take in electron spin more concisely than the thermodynamic approach. The appn. to certain cases is discussed.—E. O. H.

***Further Results in the General Theory of Barrier-Layer Rectifiers.** P. T. Landsberg (*Proc. Phys. Soc.*, 1952, [B], **65**, (6), 397–409).—L. follows up previous papers on barrier-layer rectifiers (*Proc. Roy. Soc.*, 1951, [A], **206**, 463, 477; *M.A.*, **19**, 626, 627) by considering a number of topics in the light of the theory. These include the temp. dependence of barrier height and rectifier resistance, which includes forward characteristics, temp. dependence of the (reverse) current/voltage characteristics, leading to a general discussion of current/voltage relations. Thermal instability is dealt with, curves of const. power dissipation are calculated, and finally a theory of current creep is established.—E. O. H.

***Thermo- and Galvano-Magnetic Coefficients for Semi-Conductors.** E. H. Putley (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 991–993).—A letter. The various thermo- and galvano-magnetic coeff. obtained for free electrons are applied and compared for semi-conductors. In these materials, the lattice contribution to thermal conduction is normally many times larger than the electronic contribution, and this modification to the coeff. is discussed.—E. O. H.

The Measurement of Drift Mobility in Semi-Conductors. R. Lawrence and A. F. Gibson (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 994–995).—A letter. A method of measuring drift mobilities is described, using a D.C. emitter current and a pulsed sweep field of 140 μ sec. duration and up to 50 V./cm. Experiments on *n*-type Ge are described. The width of the carrier arrival pulse is affected by diffusion, and the results are in good agreement with theory, briefly outlined here.—E. O. H.

***The Investigation of Ideal Magnetization Curves for Soft Polycrystalline Ferromagnetic Materials.** L. A. Chernikova (*Zhur. Eksper. Teoret. Fiziki*, 1951, **21**, (4), 514–523).—[In Russian]. Ch. made measurements on specimens of Mo Permalloy (Ni 78, Mo 3·8, Fe 18% + impurities), Cu Permalloy (Ni 74, Cu 8·8, Mo 3%, rest Fe and impurities), Armco Fe, and Ni, both by ballistic methods on ring specimens and by the magnetometer method on wires. Ideal magnetization was attained by superposing an alternating field, slowly reduced from saturation to zero, on the magnetizing field. Ballistic measurements were made by the usual method and by a differential method in which a current was induced in the measuring coil by suddenly increasing the magnetizing field to a value sufficient to produce magnetic saturation; the latter method is more sensitive for the initial part of the curve. There was a close correlation of the *I* values of the curve with the initial susceptibility and coercive force on the normal magnetization curve for all specimens. Tests with Cu Permalloy showed a rise in the ideal magnetization curve after annealing in asbestos or in H (the two curves were identical) and an even greater rise after annealing in a magnetic field of 15 Oe. Mo Permalloy showed no variation of the curve with temp. in the range +100° to −183° C., but Cu Permalloy annealed in a magnetic field showed much weaker magnetization at −80° and −183° C., compared with the curves at +18° and +100° C., which were identical. Comparison with the curves calculated by Kondorsky (*J. Physics* (*U.S.S.R.*), 1942, **6**, 93; *M.A.*, **10**, 38) for uniaxial and multiaxial ferromagnetic materials shows that a Mo Permalloy with low coercive force gives an uniaxial type of curve, presumably because of internal stresses, while a specimen of high coercive force gives a curve closer to that calculated for a multiaxial specimen. Armco Fe in weak fields was shown to behave similarly, approximating to the multiaxial curve only in stronger fields. Under mech. tension, Ni was shown to give lower *I* values on the ideal magnetization curves, while Mo Permalloy gave higher values; this is consistent with their resp. magnetostrictions.—G. B. H.

Single-Domain Structure in Ferromagnetic Materials and the Magnetic Properties of Finely Dispersed Substances.—II. E. Kondorsky (*Doklady Acad. Nauk S.S.S.R.*, 1950, **74**, (2), 213–216).—[In Russian]. Cf. *ibid.*, 1950, **70**, 215; *M.A.*, **18**, 500. K. derives formulæ for the dimensions of ferromagnetic particles such that they shall retain single-domain character, and for the value of the coercive force (H_c). Subject to certain assumptions and restrictions, the following relations are evolved: (1) *Spherical particles.*—The condition for single-domain character is:

$$r \leqslant \frac{1}{I_s}\sqrt{\frac{10cA}{a(N - 2K/I_s^2 - H/I_s)}}$$

where r = radius of particle, $c = 1$ for b.c.c. or 2 for f.c.c. materials, $A = Z_0^2J$ (Z_0 is the number of valency electrons and J is an exchange integral), N = demagnetization factor of the particle, which depends on the concentration of the particles, K = const. of magnetic anisotropy, H = imposed field. At $H = -H_{c\text{max.}}$ the abs. single-domain character will be retained if:

$$r \leqslant r_0 = \frac{1}{I_s}\sqrt{\frac{10cA}{aN}}$$

If r approaches r_0 and other conditions are satisfied :

$$H_c = \frac{2K}{I_s}\left(1 - \frac{7}{4} \times \frac{1 - r_0^2/r^2}{1 + K/NI_s^2}\right)$$

whence $H_{c_{\text{max.}}} = 2K/I_s$.

(2) *Particles representing ellipsoids of rotation elongated in one direction.*—The corresponding relations are evolved (*a*) if l/r is small:

$$r \leqslant \frac{1}{I_s}\sqrt{\frac{10cA}{a(N_l - 2K/I_s^2 - H/I_s)}}$$

$$r \leqslant r_0 = \frac{1}{I_s}\sqrt{\frac{10cA}{aN_d}} = l_0$$

and $H_{c_{\text{max.}}} = 2K/I_s + I_s(N_d - N_l)$; (*b*) for larger values of l/r more complex relations are evolved. Values of r_0 and l_0 for Fe and Ni calculated from these and other equations are presented. 5 ref.—D. M. P.

Magnetic Powder Patterns. H. G. Gerlach (*Appl. Sci. Research*, 1951, [B], **2**, (4), 245–248).—[In English]. G. describes differences in magnetic powder patterns obtained after annealing and polishing rolled strip of Fe contg. 4·5% Si with colloidal suspension of Fe_2O_3. The effect of shifting the domain boundaries on applying a field ‖ the crystal surface alters the pattern in two different ways, thus confirming that two possible patterns can really exist.—S. Ma.

***Zener's Theory of Ferromagnetism.** A. Teviotdale (*Proc. Phys. Soc.*, 1952, [A], **65**, (11), 957–958).—A letter dealing in particular with a criticism of papers by Zener (*Phys. Rev.*, 1951, [ii], **81**, 440; *M.A.*, **19**, 194) and by Carr (*ibid.*, 1952, [ii], **85**, 590; *M.A.*, **20**, 405). T. shows firstly that other spin interaction equilibrium conditions can be derived from Z.'s equations, and their significance is discussed. C.'s treatment is criticized on the grounds that the const. he used are inaccurate, and that his math. treatment of the equilibrium conditions is incorrect. It is concluded that Z.'s treatment is not in agreement with experiment.
—E. O. H.

Some Post-War Developments in Magnetism. L. F. Bates (*Proc. Phys. Soc.*, 1952, [A], **65**, (8), 577–594).—Presidential address. Topics discussed include: (i) the geometry of domains, motion of domain walls, and the effect of inclusions; (ii) recent studies on the gyromagnetic ratio; (iii) ferromagnetic resonance, and the values of the Landé *g* factor; (iv) paramagnetic resonance and nuclear spin; (v) nuclear resonance, and measurements of the nuclear magnetic moment of protons and deuterons; and finally (vi) neutron diffraction studies and brief mention of antiferromagnetism. 18 ref.—E. O. H.

†Antiferromagnetism and Ferrimagnetism. Louis Néel (*Proc. Phys. Soc.*, 1952, [A], **65**, (11), 869–885).—The Seventh Holweck Lecture, 1952. After a brief historical introduction to the concept of antiferromagnetism, the magnetic properties of Cr and Mn are discussed in the light of N.'s theory. Next, other examples of antiferromagnetic materials, such as MnO and α-Fe_2O_3, are considered. The limited nature of the present theories is also mentioned. The discussion of ferrimagnetism deals in detail with the magnetic properties of the ferrites (Fe_2O_3MO), where M is a bivalent metal, and the concept of ferrimagnetism is shown to be confirmed by neutron diffraction and other studies. The lecture concludes with a summary of work on other ferrimagnetics, such as Mn_2Sb and pyrrhotite. 44 ref.—E. O. H.

***Antiferromagnetism by the Spin Wave Method. I.—The Energy Levels. II.—Magnetic Properties. III.—Applications to More Complex Systems.** J. M. Ziman (*Proc. Phys. Soc.*, 1952, [A], **65**, (7), 540–547, 548–556; 1953, [A], **66**, (1), 89–94).—[I.—] Theoretical. The Hamiltonian operator of a system of particles with ordered antiparallel spins in a cryst. lattice is set up, and the results are used to discuss the case when the spins deviate from exact antiparallelism. The anisotropy of the crystal is shown to play a large part in fixing the domain axis, and the energy levels of the array are calculated for an external field ‖ this axis. The ground state energy is shown to be in good agreement with other calculations. [II.—] The results of [I] are used to calculate the partition function of the system, and hence the susceptibility $\chi_{\|}$ ‖ the domain axis. $\chi_{\perp}$ is also calculated and shown to be const. The magnetic resonance frequencies are also determined. The remainder of the paper is concerned with correction terms, in particular the correction to $\chi_{\|}$. Finally, the expression for $\chi_{\|}$ is plotted against temp., correlated with experiment, and the results discussed. [III.—] The work is extended to b.c.c. and f.c.c. lattices, where superexchange forces between next-nearest neighbours are important. The Curie temp. is calculated as a function of r, the ratio of next-nearest- to nearest-neighbour interactions. Difficulties are met in correlating this parameter with experimental results, and the possibilities of negative interaction are discussed. —E. O. H.

***Spin Degeneracy and the Theory of Collective Electron Ferromagnetism.** A. B. Lidiard (*Proc. Phys. Soc.*, 1952, [A], **65**, (11), 885–893).—Theoretical. A new approach to the problem of exchange energy is made and a model is set up which does not depend on the presence (or absence) of spin degeneracy. The mathematics are amenable to exact treatment, and lead to results which are an extension of Stoner's theory (*Proc. Roy. Soc.*, 1938, [A], **165**, 372; *M.A.*, **5**, 427). As in this case, the results depend on the overlapping of two unspecified energy bands. These bands could be the $3d$–$4s$ bands in the Fe group, but the theory should also apply to the paramagnetic Pt group. Some possible extensions of the theory are also briefly considered. —E. O. H.

***Energy of a Bloch Wall on the Band Picture.** E. P. Wohlfarth (*Proc. Phys. Soc.*, 1952, [A], **65**, (12), 1053–1054).—A letter. W. discusses theoretical estimates of the "stiffness const." A which is used in determining the characteristics of domain boundaries, and derives another expression some five times smaller than previous estimates. However, W. considers that experimental data do not yet allow one to choose between any theoretical estimates of A made so far.—E. O. H.

Kinetic Energy of the Bloch Wall. Akira Sugiyama (*J. Phys. Soc. Japan*, 1951, **6**, (6), 524–525).—[In English]. Kittel's approx. derivation of the kinetic energy of the Bloch wall, discussed in magnetic theory (*Phys. Rev.*, 1950, [ii], **80**, 918; *M.A.*, **19**, 107) is made rigorous by taking into account the apparent increase of the exchange energy const. with motion.—J. S. G. T.

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2 — PROPERTIES OF ALLOYS

***Flow and Fracture Characteristics of the Aluminium Alloy 24S-T4 as Affected by Strain Thermal History.** S. I. Liu and E. J. Ripling (*J. Metals*, 1953, **5**, (1), 66–68).—Soln.-treated, quenched, and aged specimens of this commercial Al alloy—which has a f.c.c. structure—were prestrained (in either tension or compression) by various amounts at fixed temp. and then further strained to failure at either room or lower temp. Prestraining at room temp. and testing at −78° or −196° C. produced a lower flow stress curve than would the same total strain at the low temp. Prestretching at the low temp. produced a higher room-temp. flow curve than would an equal strain at room temp. Room-temp. tensile prestrains had little effect on the low-temp. fracture stress, while room-temp. compressive prestrains were harmful to the fracture stress only when the magnitude of the precompression was quite large. When the room-temp. prestraining and the low-temp. testing were both by tension, the dependence of retained ductility on prestrain could be described by the equation previously presented by R. for steel (cf. R. and Sachs, *Trans. Amer. Inst. Min. Met. Eng.*, 1949, **185**, 78). The results are discussed in relation to those obtained by other workers from similar experiments on Al and Cu. 7 ref.—E. N.

***Equilibrium Relations at 460° C. in Aluminium-Rich Alloys Containing 0–7% Copper, 0–7% Magnesium, and 0·6% Silicon.** H. J. Axon (*J. Inst. Metals*, 1952–53, **81**, (9), 449–450).—The equilibrium isothermal at 460° C. is given for quaternary alloys rich in Al and contg. 0–7 wt.-% Mg, 0–7 wt.-% Cu, and const. (0·6 wt.-%) Si. The phases encountered are the Al-rich solid soln., $CuAl_2$, Mg_2Si, Si, the ternary phase Al_2CuMg (S), and the quaternary phase $Al_5Cu_2Mg_8Si_6$ (Q). The general form of the isothermal contg. 0·6% Si is similar to that previously reported for alloys contg. 1·2% Si (*ibid.*, (4), 209; *M.A.*, **20**, 331), but the phase fields are all translated towards the Al-rich corner.—Author.

The Granulation of Eutectic Silicon in Silumin by Heat-Treatment. Hachie Sawamoto (*Suiyokwai-Shi*, 1951, **12**, (1), 14–16).—[In Japanese]. Experiments have shown that eutectic Si in Silumin is granulated by low-temp. annealing and spheroidized by high-temp. annealing, and that annealed Silumin can be worked by forging and rolling. Max. elongation of 23% can be obtained by annealing at 450° C. for 20 hr.—Author.

***The Influence of the Manganese, Silicon, and Iron Contents on the Tensile Properties of Aluminium-Zinc-Magnesium Alloys.** W. Mannchen (*Metall*, 1952, **6**, (1/2), 24–26).—A study has been made of the effects of variations in the Mn + Si + Fe contents—in amounts ranging from a trace to ~0·5% for each individual metal—on the tensile properties of an Al–Zn (4·5%)–Mg (3·5%) alloy, with and without the addn. of 0·35% Cu. The tests were carried out on 4-mm.-dia. rod (produced by drawing from 8-mm.-dia. extruded rod) which had been soln.-treated in an air furnace for 45 min. at 450° C., quenched in H_2O, and aged for 50 days at 20° C. The U.T.S., P.S., and shear strength were found to improve with increase in Mn content and with decrease in Si content; variations in the Fe content had little or no influence on the results obtained. The elongation of Cu-free alloys decreased with increasing content of the other alloying elements, but when Cu was present an improvement took place if the Mn content was increased simultaneously with the Fe and Si contents. 14 ref.—E. N.

No Heat-Treat[ment] Needed for These Strong Aluminium [-Zinc-Magnesium] Alloys. Donald L. Colwell (*Amer. Foundryman*, 1952, **22**, (6), 60–65).—The properties of Al-base alloys contg. Zn and Mg as major alloying addn. are given.—V. K.

The Utilization of High-Strength Light Alloys. F. Vinsonneau (*Tech. Sci. aéronaut.*, 1952, (4), 205–223).—A critical survey of the properties of high-strength Zn-contg. light alloys for use in aeroplane manufacture with particular ref. to French practice. Detailed comparisons are made of the chem. compn. and mech. properties of the following alloys:

Zn content, %	France	Gt. Britain	U.S.A.
5–6	...	RR 77 Spec. D.T.D. 683, 687, 697A	75S Spec. AN-A10 and AN-A11
7–8	A-Z8GU Zicral–Air Spec. 9050	RR 88 Spec. D.T.D. 363A	...

Zicral is compared with the more usual light alloys from the points of view of resistance to rapidly alternating stresses, fatigue, and heat, and results are included showing its value over A-U4G1 in a number of tests having a direct bearing on aeroplane construction. The difficulties associated with the use of this type of alloy (heat-treatment, joining, surface treatment) are also discussed.—H. A. H.

***On the Improvement of the Machinability of Cast Aluminium Alloys.** Fusao Hayama (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1951, (2), 20–24).—H. discusses the effect of addn. of Bi and Sn on the machinability and surface finish of Al–4·0% Cu–4·5% Si alloy (Lautal series). The results show that both these elements improve the machinability and surface finish; the presence of Fe, Mn, and Zn does not reduce the efficacy of Sn or Bi.—G. J. S.

***Solubility of Nitrogen in Liquid Chromium and Chromium–Silicon [Alloys].** (Mozgovoi and Samarin). See col. 673.

***Study of Ferromagnetic Resonance in the Alloy CrTe.** T. M. Perekalina (*Doklady Akad. Nauk S.S.S.R.*, 1952, **84**, (3), 475–476).—[In Russian]. Measurements were made of the H.F. magnetic permeability of CrTe by the wave-guide impedance method of Birks (*Proc. Phys. Soc.*, 1948, **60**, 282), using plates pressed from equal vol. of the finely powdered material and paraffin wax.—G. V. E. T.

Beryllium-Copper Wire: Its [Properties,] Processing, and Uses. John T. Richards (*Wire and Wire Products*, 1952, **27**, (2), 149–154, 192–195).—R. describes the alloy of Cu with 1·9% Be and traces of Ni or Co, considers its structure, and gives a phase diagram indicating the soln.-annealing and age-hardening range. Graphs show the work-hardening rates resulting from cold-drawing for Cu–Be, Cu–Si, Cu, and phosphor bronze. Heat-treating practices, including soln.-annealing, pptn.-hardening, and homogenizing, are reviewed. Finally the manufacture and performance of coiled springs made from Cu–Be alloy are described. 13 ref.—C. P. F.

Beryllium-Copper Wire: Property and Design Considerations. John T. Richards (*Wire and Wire Products*, 1952, **27**, (3), 257–262, 304–307).—R. illustrates by graphs and tables the effects of the Be content on the strength and conductivity of Cu wire, showing that an increase in Be causes a drop in conductivity and an increase in U.T.S. for both annealed and cold-drawn conditions. The influence of cold drawing and heat-treatment upon *d* is described. The effect of reducing the temp. to −94° F. (−70° C.) is to increase the U.T.S. of unhardened and heat-treated Cu–Be alloy by 12 and 9% of the U.T.S. at 68° F. (20° C.), resp. The corrosion-resistance is shown to compare favourably with that of Cu under difficult atmospheric conditions. 20 ref.—C. P. F.

***Study of Halman Electrical Resistance Alloy [Cu–Mn–Al]. VI.—The Effects of Aluminium on the Properties.** Kiichiro Shinji (*Suiyokwai-Shi*, 1952, **11**, (10), 546–549).—[In Japanese]. A new Ni-free alloy consisting of Cu, Mn (7·5–7·8%), and Al and suitable for use as resistance wire in precision instruments, has been manufactured by S. With increase of Al content: (i) the sp. resistance increases to a max. at 6% Al, (ii) temp. coeff. decreases to a min. at 3% Al, (iii) the thermo-e.m.f. against Cu decreases to a negative direction and begins to increase in a positive direction at 3% Al, and (iv) the U.T.S. increases.—AUTHOR.

***Study of Halman Electrical Resistance Alloy [Cu–Mn–Al]. VII.—The Best Composition for a Standard Electrical Resistance Alloy.** Kiichiro Shinji (*Suiyokwai-Shi*, 1952, **12**, (2), 1–3).—[In Japanese]. S. has investigated the relations between sp. resistance, temp. coeff. of resistance, and thermo-e.m.f. over a wide range of Al and Mn contents. Results show that the best compn. for a standard elect. resistance alloy is that at which the temp. coeff. of resistance and themo-e.m.f. are zero or nearly zero, the Al content is <5%, Fe content <1·2%, and Si content is as small as possible.—AUTHOR.

Novoston: A New [Copper-Base] Propeller Alloy. —— (*Marine Eng.*, 1953, **76**, (913), 76–77).—The new alloy has a Cu base, major alloying elements being Mn, Al, and Ni. It is claimed to have better mech. properties than ordinary propeller bronze (Ni–Al bronze). The lower sp. gr. of the alloy is a great advantage in propeller design.—S. MA.

***The Viscosity of Molten Copper–Silver and Copper–Gold Alloys.** Erich Gebhardt and Georg Wörwag (*Z. Metallkunde*, 1951, **42**, (12), 358–361).—Using methods previously described (cf. G. and Becker, *ibid.*, (4), 111; *M.A.*, **20**, 241), the viscosity of high-purity Cu–Ag (25, 50, 72, and 85% Ag) and Cu–Au (27, 58·5, and 75% Au) alloys has been determined within the temp. range liquidus–1300° C. The viscosity values, calculated from the measured values of the log decrement of damping, are portrayed graphically, as a function of temp. and compn. They show that the viscosity: (1) of pure Cu (99·99%) is 3·19 centipoises at 1200° C., compared with previously determined values of 4·63 and 2·84 centipoises for Au and Ag, resp., at the same temp.; (2) of all alloys and their component metals (*a*) increases with decrease in temp., and (*b*) has temp. coeff. which are dependent on the Cu content of the alloy; and (3) curves in the (*a*) Cu–Ag system tend to converge in the region of 2·6–2·8 centipoises at 1300° C., and (*b*) Cu–Au system tend to converge to an extrapolated value of ~6·0 centipoises at 900° C. All compn. given are in wt.-%. 7 ref.—E. N.

***The Structure of the Copper–Silver–Indium System. I.—Equilibria in the Copper–Silver Region.** Erich Gebhardt and Manfrid Dreher (*Z. Metallkunde*, 1951, **42**, (8), 230–238).—A thermal and microscopic study has been made of the system up to contents of Cu 65, Ag 100, and In 35%, the alloys being prepared from 99·9% pure metals by melting in sintered corundum crucibles under H atmospheres. Alloys contg. up to 10% In were soft and easily worked; those contg. 20–30% In were hard and difficult to work, some being extremely brittle. Equilibrium conditions were established by annealing *in vacuo* for 35 or 24 days at 450° or 550° C., resp. From the results, which are discussed at length, diagrams have been produced showing the surfaces of primary crystn., and temp./concentration equilibria at const. contents of 10 and 20% In. Photomicrographs show typical structures. The following four-phase equilibria were found: (1) at 600° C., melt $+ \alpha_1 = \alpha + \beta_1$, (2) at 575° C., melt $+ \alpha = \beta + \beta_1$, (3) at 490° C., $\beta = \alpha + \beta_1 + \delta$, and (4) at 475° C., $\beta_1 = \alpha + \alpha_1 + \delta$, where α and α_1 are Cu–In and Ag–In solid-soln., resp., contg. <0·6% Ag and <1·0% Cu at 400° C., resp., and <0·1% of either metal at room temp.; β and β_1 correspond to solid soln. based on the intermetallic compounds Cu_4In and Ag_3In, resp., in the corresponding binary systems; and δ corresponds to the similar phase in the Cu–In system. All compn. are in wt.-%. 6 ref.—E. N.

***Mechanism of the Precipitation of the α from the β Phase in 60 : 40 Brass.** Gunji Shinoda and Yoshitsugu Amano (*J. Phys. Soc. Japan*, 1951, **6**, (6), 508–512).—[In English]. Cu–Zn alloys contg. 40% Zn, which have $(\alpha + \beta)$ phases in equilibrium yield nearly uniform β structures when quenched from 850° to ~900° C.; on tempering the β phase, it transforms into α, undergoing a reversal process of the Kurdjumow and Sachs type. Various intermediate structures between β and f.c.c. α phases were revealed by X-ray analysis, e.g. b.c. and f.c. tetragonal structures having various axial ratios. It is considered that the transformation occurs gradually. No definite boundaries occur between the b.c. and f.c. tetragonal structures. The $\alpha/(\alpha + \beta)$ phase boundary is found to lie on the side somewhat poorer in Zn than that found by Konobejewski and Tarrasowa (*Physikal. Z. Sowjetunion*, 1936, **10**, 427; *M.A.*, **4**, 88).—J. S. G. T.

***Observations on Liquid [Electrical] Contact Bridges. III.[—Alloys of Copper, Nickel, and Cobalt with Gold].** Gerhard Schrag and Rudolf Toberer (*Z. Metallkunde*, 1951, **42**, (8), 243–245).—A continuation of earlier work (*ibid.*, p. 24; *M.A.*, **19**, 501). As large amounts of molten metal and oxides are formed each time the circuit is broken between contacts of electrolytic Cu, Mond Ni (99·5–99·7%), and Co (99·0–99·5%), the contacts soon cease to pass any current. Increasing alloying addn. of fine Au to the contact materials progressively reduce the amount of metal in the bridge to very small values; oxide formation (1) is non-existent in binary alloys of Cu with 2–3% Au, but (2) is still present in binary alloys of Ni contg. up to 16% Au, and Co contg. up to 8% Au. 3 ref.—E. N.

***Density and Specific Volume of Liquid and Solid Gold–Silver Alloys.** Erich Gebhardt and Stefan Dorner (*Z. Metallkunde*, 1951, **42**, (12), 353–358).—The *d* and the sp. vol. of Au–Ag alloys contg. 0, 10, 20, 40, 60, 80, and 100 wt.-% Au, have been determined in (1) the liq. state up to 1300° C. by an indirect-displacement method, using a molten NaCl bath, the necessary const. of which were first determined at 50° C. intervals between 800° and 1300° C., and (2) the solid state—at ~200° C. intervals up to the solidus temp.—from dilatometric and X-ray data. The results are tabulated, at 20° C., the solidus and the liquidus temp., and 1300° C. for: *Fine Ag* (m.p. 960·5° C.), *d* = 10·565, 9·936, 9·285, and 8·969, resp., and the sp. vol. = 0·0947, 0·1006, 0·1077, and 0·1115, resp., and *Fine Au* (m.p. 1063° C.), *d* = 19·423, 18·474, 17·361, and 16·950, resp., and sp. vol. = 0·0515, 0·0541, 0·0576, and 0·0590, resp. For all the alloys the experimentally determined values are in good agreement with those obtained from theoretical considerations. When alloy formation takes place, it is accompanied by a very small vol. contraction. Shrinkage is scarcely affected by alloy compn.; for alloys contg. 0, 10–80, and 100% Au, it is, approx., 6·5, 7·5–6·9, and 6·0%, resp., during solidification, and 6·0, 5·8–5·0, and 4·9%, resp., during cooling from the solidus to room temp. 31 ref.—E. N.

Rule for Temperature Dependence of Solubility of Metals in Iron. I. I. Kornilov (*Doklady Akad. Nauk S.S.S.R.*, 1950, **72**, (6), 1067–1070).—[In Russian]. The variation of solubility with temp. of a number of elements in Fe is explained on the basis of the difference in atomic dia. and the change in this difference with temp. The coeff. of expansion is taken as a first approx. to the change of atomic dia. with temp., and increasing solubility is expected if the coeff. operates to reduce the difference in atomic dia. and vice versa. The rule is used to explain the solubility changes with temp. observed in many Fe alloys. 5 ref.—D. M. P.

***Equilibrium Diagrams of the Systems Iron–Antimony–Sulphur and Cobalt–Antimony–Sulphur.** Werner Lange and Heinz Schlegel (*Z. Metallkunde*, 1951, **42**, (9), 257–268).—A thermal, microscopic, and X-ray study has been made of the system Fe–Sb–S, and from the results partial and complete equilibrium diagrams of the region Fe–$FeS_{1\cdot 08}$–Sb_2S_3–Sb have been produced. The results, which are discussed at length, show that: (1) there is an extensive miscibility gap in the system; (2) the reaction $Sb_2S_3 + 3Fe \rightleftharpoons 2Sb + 3FeS$ proceeds more rapidly from left to right, so that under appropriate conditions molten FeS and Sb are practically in equilibrium; (3) as previously reported, there exists a solid soln. of compn. $FeS_{1\cdot 08}$, which, however, evolves S at high temp.; (4) only one ternary compound exists, the double sulphide FeS (solid soln.). Sb_2S_3, known as Berthierite, which forms as the result of a peritectic reaction at 530° C.; and (5) there is no evidence of the previously reported compound FeSbS. Photomicrographs show typical structures. Thermodynamic considerations of analogous systems, metals of the Fe group—As, Sb, or Bi—and S, together with a few experimental observations, have led to the construction of a hypothetical equilibrium diagram of the Co–CoS–Sb_2S_3–Sb region of the system Co–Sb–S, which appears to be intermediate between those of Fe–Sb–S and Ni–Sb–S (for which a corrected diagram of the region Ni–NiS–Sb_2S_3–Sb is given). The system contains small miscibility gaps and a compound CoSbS, but the compound CoS.Sb_2S_3, analogous to Berthierite, does not appear to exist.—E. N.

***A Method for the Investigation of Quaternary Systems, Demonstrated with the System Lead–Zinc–Cadmium–Tin.** Erich Lindner (*Z. Metallkunde*, 1951, **42**, (12), 377–387).—A method for the representation of the constitution of quaternary systems is described, using as an example the system Pb–Zn–Cd–Sn, and taking cuts ‖ two intersecting sides of the regular tetrahedron, the faces of which represent the known ternary systems Pb–Sn–Cd, Sn–Cd–Zn, Cd–Zn–Pb, and Zn–Pb–Sn. Approx. 45% of the total vol. of this tetrahedron consists of a miscibility gap, the boundaries of which have been determined by thermal analysis. The quaternary eutectic point occurs at 138° C., and Pb 28·6, Zn 2·25, Cd 16·7, and Sn 52·45 wt.-%, i.e. it is close to the ternary eutectic of the Pb–Cd–Sn system. The course of the eutectic valleys from the ternary eutectic points of the three-component systems to the quaternary eutectic point have been determined from theoretical considerations rather than experimental observations. 7 ref.—E. N.

The Production and Properties of Magnesium Alloy Tubing. (Huffman and Ansel). See col. 727.

***Studies on Segregation in High-Strength Wrought Magnesium–Aluminium Alloys.** G. Siebel (*Metall*, 1952, **6**, (3/4), 64–69).—A report of a study of the segregation phenomena accompanying the casting of ingots of the high-strength commercial Mg–Al alloy AZ 855 and its modifications, contg. Al 7–8·5, Zn 0·5–1·5, and Mn 0·1%. Segregation is minimized if large billets (300–450 mm. dia.) for forging are cast by the *Tütenguss* process—in which the molten metal is poured into a thin-walled mould and held for some time in the molten state to allow non-metallic impurities (mainly P) and Fe to settle to the bottom of the mould, after which the metal is solidified by slow immersion of the mould in H_2O, the butt end of the ingot being finally cut off and discarded. This casting procedure, however, does not prevent the occurrence of unwanted segregation of Al_3Mg_4 crystals, which grow ‖ the *c*-axis and the basal plane of the massive columnar crystals of Mg–Al solid soln., and appear as stringers radiating from the longitudinal axis of the ingot. The particles of Al_3Mg_4 do not easily go into soln. during heat-treatment, and consequently there is a tendency for the material to crack during extrusion and forging. X-ray examination of Al_3Mg_4 isolated from ingots shows that it corresponds to the previously reported δ phase of the Al–Mg system, but appears to contain a little Zn in solid soln. The macro- and microstructures of typical ingots, extrusions, and forgings are illustrated. 11 ref.—E. N.

***The Formation of Intracrystalline Voids in Solution-Treated Magnesium–Aluminium Alloys.** E. Lardner (*J. Inst. Metals*, 1952–53, **81**, (9), 439–442).—It has been observed that occasionally the soln.-treatment of cast Mg–Al alloys results in the formation of small hexagonal voids in the centre of many grains. These voids have been shown to be orientated with their hexagonal axes ‖ the hexagonal axes of the crystals in which they occur. No explanation has been found for their formation, but it has been shown that in a sample of an alloy that does form cavities, the cavity formation increases to a max. and then eventually vanishes with increasing homogenization. It has also been shown that the cavities are probably produced on cooling after soln. treatment rather than during the progress of the heat-treatment.—AUTHOR.

***The Ternary System Magnesium–Calcium–Bismuth.** Werner Köster and Fritz Sautter (*Z. Erzberg. u. Metallhüttenwesen*, 1952, **5**, (8), 303–307).—A thermal and metallographic study of the system has been made, in order to elucidate the basic principles of the Kroll–Betterton process for debismuthizing Pb by addn. of Mg and Ca. The system contains five quasi-binary regions, and, at room temp., is divided into seven three-phase regions, viz. $Ca + CaMg_2 + Bi_2Ca_3$, $Mg + CaMg_2 + Bi_2Ca_3$, $Mg + Bi_2Ca_3 + T$, $Mg + T + Bi_2Mg_3$, $Bi + T + Bi_2Mg_3$, $Bi + T + Bi_3Ca$, and $Bi_2Ca_3 + T + Bi_3Ca$, the last three of which occupy only the small corner Bi_2Ca_3–Bi–Bi_2Mg_3 of the diagram. T, the only ternary compound which occurs, is formed in the quasi-binary region Bi_2Ca_3–Bi_2Mg_3; it has the formula Mg_2CaBi_2 (previously reported by Evers, *ibid.*, 1949, **2**, 129; *M.A.*, **17**, 744), the compn. Mg 9·7, Ca 8·1, and Bi 82·7% (*sic*), and melts congruently at ~1000° C. Primary crystn. in the system is controlled by the phase Bi_2Ca_3, which has a m.p. >1200° C. (i.e. higher than the 928° C. previously reported). The high m.p. of the phases Mg_2CaBi_2 and Bi_2Ca_3 is one reason for the high affinity of Bi for Ca and Mg, on which the success of the Kroll–Betterton process depends. The constitution of the system, and the ternary eutectics and four-phase transformations which occur, are described in detail and illustrated by phase diagrams; photomicrographs show typical structures. 4 ref.—E. N.

***The Gamma Phase of Manganese. I.—The Effect of [Alloying] Additions on the Room-Temperature Structure of Alloys Quenched from the Gamma Region.** Ulrich Zwicker (*Z. Metallkunde*, 1951, **42**, (8), 246–252).—A report of a study of the effects of alloying addn. on the stability of the high-temp. γ phase of Mn. 1–5-g. buttons of the binary alloys were prepared from electrolytic Mn and high-purity alloying metals by melting under H, homogenizing at <1000° C. *in vacuo*, or >1000° C. under H, followed by quenching or rapid cooling in H in an attempt to retain any γ phase. The lattice const. of the phases present were then determined from Debye–Scherrer X-ray-diffraction patterns. The results show that the f.c. tetragonal γ structure (γ_t) can be retained by quenching only if the alloying addn. brings about a marked decrease in the $\gamma \rightarrow \beta$ transformation temp. of Mn. When γ_t is retained as a homogeneous phase a point is reached when further alloying addn. lead to the formation of a heterogeneous two-phase—tetragonal + cubic $\gamma(\gamma_c)$—structure, followed by a homogeneous γ_c field; the transition zone corresponds to the range in which the β or α phase would be found on slow cooling of the alloys. Addn. of Fe 37·5, Al 18·4, As 3·7, In 7·8%, or Sb, Sn, or Si do not lower the $\gamma \rightarrow \beta$ transformation temp., and γ_t formation is non-existent; some γ_t, in association with β, is found in alloys contg. Ni 9·5 or Ag 2·6%. The structures found on addn. of Ge, Ga, Zn, Cu, and Pt—elements which are known appreciably to decrease the $\gamma \rightarrow \beta$ transformation temp. in Mn—are:

Element, %	Homogeneous γ_t	Heterogeneous $\gamma_t + \gamma_c$	Homogeneous γ_c
Ge . .	Up to 9·5	9·5–12·5	12·5–16·6
Ga . .	,, 19·5	20–22	22–28·9
Zn . .	,, 19	19–20	20–23·5
Cu . .	,, 13·5	14–17	17–27
Pt . .	,, 4·8	...	...
Ni . .	9·5 (includes some β)	14·3	19
Ag . .	2·6 (includes some β)	...	...

All compn. are in at.-%. 14 ref.—E. N.

***On a Transformation in the Mn_3Cr Phase in the System Manganese-Chromium.** Ulrich Zwicker (*Z. Metallkunde*, 1951, 42, (9), 277–278).—The work of Carlile, Christian, and Hume-Rothery (*J. Inst. Metals*, 1949–50, 76, 169 and 195; *M.A.*, 17, 915 and 916) on the equilibrium diagram of the system Cr–Mn, and the effects of N on the structures of certain alloys within the system, is discussed and related to Z.'s own work on the system (*Z. Metallkunde*, 1949, 40, 377; *M.A.*, 18, 21). Further thermal and dilatometric studies by Z., using alloys prepared by methods previously reported, and carefully homogenized under H, by techniques which ensured the absence of any contamination by atmospheric N, have led to the following observations, which are incorporated in a revised equilibrium diagram: (1) 65, 45, and 30 at.-% Mn are soluble in α-Cr at 960°, 900°, and 600° C., resp., and (2) a transformation $\alpha\text{-}Mn_3Cr \rightarrow \beta\text{-}Mn_3Cr$ occurs at $\sim$975° C., this temp. decreasing somewhat for Mn contents above and below that corresponding to the stoichiometric compn., and becoming $\sim$960° C. in the two-phase fields of Mn_3Cr with α-Cr and β-Mn. The $\alpha \rightarrow \beta$ transformation is accompanied by an increase in vol. 4 ref.—E. N.

***Study of the Manganese-Nitrogen System.** Ulrich Zwicker (*Z. Metallkunde*, 1951, 42, (9), 274–276).—A microscopic and X-ray study of the system has been made, and from the results a partial equilibrium diagram, up to 45 at.-% (15 wt.-%) N and between 400° and 1400° C., has been produced. The alloys were prepared by treatment of electrolytic and powdered Mn in NH_3 or N by procedures which are described. The m.p. of Mn rises with increasing addn. of N (to 1300° C. at 13 at.-% N); the $\alpha \rightarrow \beta$ transformation temp. falls to $\sim$600° C., at which temp. the β-Mn solid soln. decomposes into α-Mn and a N-rich γ solid soln. which below 465° C. is converted into $\alpha + Mn_4N$. When alloys contg. up to $\sim$9 at.-% N are quenched from the γ phase region they exhibit, at room temp., the ductile tetragonal form of γ-Mn; alloys contg. greater amounts of N show the brittle, striated, ferromagnetic f.c.c. Mn_4N phase which also exhibits some superlattice formation and which is lost—together with the ferromagnetism—on heating above 700° C., when the phase becomes ductile. 6 ref.—E. N.

***Solubility and Diffusion Coefficient of Carbon in Nickel: Reaction Rates of Nickel–Carbon Alloys with Barium Oxides.** (Lander, Kern, and Beach). See col. 675.

***The Constitution of Nickel-Rich Alloys of the Nickel–Chromium–Aluminium System.** A. Taylor and R. W. Floyd (*J. Inst. Metals*, 1952–53, 81, (9), 451–464).—The equilibrium relationships in Ni–Cr–Al alloys contg. >50 at.-% Ni have been studied over the temp. range 750°–1150° C. It has been shown that the phase fields of the Ni primary solid soln., γ, and of the β solid soln. based on NiAl both contract as the temp. falls, whereas that of the γ' phase based on Ni_3Al extends. The equilibrium between the γ and β phases which exists from the solidus breaks down at about 1000° C., giving rise to equilibrium between γ' and α-Cr by a four-phase reaction: $\beta + \gamma \rightleftharpoons \alpha + \gamma'$. The ordering of the γ phase is enhanced by the presence of Al atoms, and for alloys with >10 at.-% Al the ordering temp. is over 1150° C. The equilibrium between the γ and γ' phases is such that over a range of compn. the lattice parameters of the phases are almost identical.—AUTHORS.

***[Electrical] Resistance Anomalies in a High-Permeability Nickel-Iron-Molybdenum Alloy.** Fritz Assmus and Friedrich Pfeifer (*Z. Metallkunde*, 1951, 42, (10), 294–299).—" Supermalloy "—contg. Ni 79, Mo 5, Mn 0·5%, balance Fe—was prepared by the normal methods (cf. Boothby and Bozorth, *J. Appl. Physics*, 1947, 18, 173; *M.A.*, 14, 369), and measurements were made of its magnetic permeability and elect. resistance as functions of tempering temp. (between 300° and 800° C.) in comparison with those of 50 : 50 and 67 : 33 Ni–Fe Permalloys. The Supermalloy results show that, simultaneously with the increase of permeability obtained on tempering, there is: (1) an increase in elect. resistance at room temp., and (2) a dependence of the elect. resistance at elevated temp. on (*a*) the rate of cooling and (*b*) the time at which the material is held at the tempering temp. These anomalies in elect. resistance are: (1) very similar to those observed in a whole series of other alloys, e.g. the Ni–Cr alloys, and especially the 79 : 21 compn., and (2) in marked contrast to those observed with the Permalloys. The characteristics of the Permalloys are considered to be due to superlattice formation; the characteristics of the Supermalloys, therefore, must be due to a process or processes—of definite activation energy—which are either more complicated or different in nature—and may result from the onset of the " K-state " in the homogeneous solid soln., as postulated by H. Thomas (*Z. Physik*, 1951, 129, 219; *M.A.*, 20, 473). 16 ref.—E. N.

***The Distribution of Internal Stresses in Permalloy Subjected to Field Cooling.** Makoto Osawa and Ichiro Sugiura (*J. Phys. Soc. Japan*, 1951, 6, (6), 530–531).—[In English]. O. and S. measured, at room temp., the internal friction in a cylindrical rod of Permalloy, with and without field cooling, after being fully annealed at $\sim$1000° C. and being subjected to the following heat-treatment: maintained at 800° C. for 30 min., then rapidly cooled to 490° C. and kept at 490° ± 5° C. for definite times and finally cooled rapidly to room temp. Mean distances, L, between stress inhomogeneities in the alloy, of the order of 3×10^{-3} cm. for an annealing time of 5 min. in the presence of field cooling, and for an annealing time of 45 sec. in the absence of field cooling, were found. The observed values of L do not agree with those calculated in accordance with the Becker–Kersten theory (see B. and Döring, " Ferromagnetismus ", Berlin, 1939), nor with Sato's detn. (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1949, [A], 1, 71; *M.A.*, 18, 181).—J. S. G. T.

***The Spectral Dependence of Reflection in Alloys of Silver.** M. M. Noskov and G. P. Skornyakov (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], 14, (6), 757–761).—[In Russian]. Continuing previous work (*Doklady Akad. Nauk S.S.S.R.*, 1949, 64, (6), 809), N. and S. investigated the effect of alloying elements on the sharp min. in reflectivity and max. in transparency of Ag at λ 3210 Å. Alloys of Ag with 1, 2, and 5% Al, Cu, Au, Zn, Cd, Tl, Sn, Sb, Pd, Pb, and Bi were prepared and homogenized by annealing in sealed glass capsules. The variation of reflectivity with λ was measured by photometric comparison with the reflection from an Al surface on a medium-dispersion quartz spectrograph. The light source was an H discharge tube. Results show that all alloying elements shift the position of the min. in reflectivity to shorter λ and reduce its depth and sharpness, both effects increasing with concentration of the solid soln. The magnitude of the effect varies greatly according to the alloying element; the positions of the min. for 5% alloys being: Sb, 3020; Cu, 3208; Pb, 3195; Au, 3210; Zn, 3173; Cd, 3162; Al, 3210; Pd, 3147; Tl, 3130; Bi, 3200; Sn, 3055 Å. N. and S. suggest a mechanism for the effect, asserting that near an atom of alloyed element the electron energy bands of Ag are too greatly perturbed to participate in the reflectivity anomaly, but in other regions the lattice is compressed, so that the characteristic frequency of the transition involved is raised. Hence atoms of greater atomic radius than Ag should show the greatest effect. This fits most of the facts, if allowance is made for the limited solubility of Cu, Bi, and Sb, but Pd is an exception, having a smaller atomic radius than Ag and a large effect on the reflectivity. N. and S. attribute this to the " zero-valency " of Pd in alloys, pointing out that addn. of 70% Ag to Pd annihilates its paramagnetism by filling the *4d* band.—G. B. H.

***A Study of the Titanium-Rich Aluminium–Titanium Alloys.** W. Gruhl (*Metall*, 1952, 6, (5/6), 134–135).—A report of a metallographic and X-ray study of the Al–Ti system between 50 and 80% Ti. The alloys were prepared by alumino-thermic methods, being commercially produced master alloys for the introduction of Ti into Al casting alloys and consequently contained Fe 0·38–0·43 and Si 0·56–0·63% as impurities; they were slowly solidified and cooled so that equilibrium conditions were obtained. The results are compared with those of Ogden and co-workers (*Trans. Amer. Inst. Min. Met. Eng.*, 1951, **191**, 1150; *M.A.*, 19, 526) and show the following deviations: (1) an extremely narrow γ (AlTi) phase region, extending only from 59 to 61% Ti; (2) an α phase with a max. content of 23% Al between room temp. and 1000° C., and 35% Al at the peritectic temp., ~1480° C.; and (3) slightly larger *c/a* axis ratios for both the f.c. tetragonal γ phase, and the h.c.p. α solid soln.; these differences are ascribed to the small amounts of Fe and Si present as impurities. Pptn. of γ is restricted to the basal plane of the h.c.p. α lattice, and this plane then has a noticeable tendency to act as a cleavage plane. Photomicrographs show typical structures. All compn. are in wt.-%. 2 ref. —E. N.

Titanium Carbide: A Highly Refractory Substance of Low Density. M. Oswald (*Métaux, Corrosion-Ind.*, 1952, **27**, (318), 75–88).—O. describes the properties of TiC, with particular reference to its refractory properties, its properties as a solvent for other substances, often themselves highly refractory, and its *d*, which may be widely modified by addn. of suitable solutes. Methods of prodn. of the carbide and of solid soln. based thereon are described and appn. indicated. 21 ref.—M. A. H.

Intermetallic Compounds and Alloy Systems of Uranium. Joseph J. Katz and Eugene Rabinowitch (*The Chemistry of Uranium. I.—The Element, Its Binary and Related Compounds* (*National Nuclear Energy Series*), 1951, **VIII-5**, 174–179).—Brief discussions are given of the preparation of U alloys, of solid-soln. and intermetallic-compound formation in binary U systems, and of the phys. and chem. properties of U intermetallic compounds. Tables are given of the solubility of U in 18 metals and of these metals in the 3 U phases, of the crystal structure of 18 intermetallic compounds with U, and of the melting or decomposition temp. of 16 intermetallic compounds. 12 ref.—P. E. M.

The Uranium–Hydrogen System. Joseph J. Katz and Eugene Rabinowitch (*The Chemistry of Uranium. I.—The Element, Its Binary and Related Compounds* (*National Nuclear Energy Series*), 1951, **VIII-5**, 183–213).—The solubility of H in solid U is slight (<17 p.p.m. at 1133° C.); it increases to 28 p.p.m. on melting and rises rapidly with temp. The diffusion coeff. of H in U increases from 0·0040 in./hr. at 566° C. to 0·0125 in./hr. at 640° C. UH_3 is a stoichiometric compound for which the following thermodynamic data have been calculated: $\Delta H_{298} = 30{\cdot}4$ kg.cal./mole, $\Delta F^0_{298} = -17{\cdot}7$ kg.cal./mole, $\Delta S^0_{298} = -42{\cdot}6$ e.u. 225°–250° C. is the optimum temp. for conversion of U to UH_3. The rate of formation decreases below 225° C., where it is governed by the excess H pressure and the state of aggregation of the U; reaction occurs with powdered U at −80° C., but is not observed at liq.-air temp. The rate of hydrogenation is a function of excess H pressure above the decomposition pressure (p_0), and the initial rate of hydrogenation is represented by $V_0 = K(p - p_0)^{5/2}$. The rate of hydrogenation is independent of time and of the purity of the H, apart from an induction period which increases as the temp. is lowered. The induction period is much lower for purified H, but it is not altered by prior exposure of the U to N or O. UH_3 is cubic with $a = 6{\cdot}6310$ Å.; its structure has been discussed by Rundle (*J. Amer. Chem. Soc.*, 1947, **69**, 1719) and Pauling and Ewing (*ibid.*, 1948, **70**, 1660). UH_3 has a calculated *d* of 10·92 g./c.c. and measured (by He displacement) *d* of 10·95 g./c.c. UH_3 is pyrophoric, but may be rendered non-pyrophoric by exposure to air at −78° C. CO_2 and N are safe atmospheres up to 200°–225° C. UH_3 burns in air to H_2O and U_3O_8; in HCl, HBr, PH_3, and NH_3 to trivalent U compounds; and in Cl_2, Br_2, H_2O, HF, and $COCl_2$ to quadrivalent U compounds. Alkali soln. and organic acids do not attack UH_3. Dil. inorganic acids have little effect, but hot strong acids oxidize UH_3 to U^{4+} or UO_2^{++} salts. Soln. of Ag^+, Hg^{++}, Sb^{+++}, Cu^{++}, and Fe^{+++} salts oxidize UH_3 to U^{++++} or UO_2^{++} salts, but Bi and Pb salts have no effect. H_2O_2 (alone or with acids), ceric salts, dichromate, and permanganate and bromate in the presence of acids also oxidize UH_3. UH_3 forms colloidal systems with Hg or Na–K. It dissolves in Wood's metal or Sn, liberating H. Air precipitates UH_3 from the amalgam as the " quasi-hydride " $UH_{2\cdot 82}O_{0\cdot 2}$. Uranium deuteride ($UD_3$) has a similar structure to UH_3 with $a = 6{\cdot}625$ Å. and $d = 11{\cdot}16$ g./c.c. It is conveniently prepared by decomposing heavy water at 600°–700° C. with U and combining the D_2 formed with U at 250° C. The decomposition pressure of UD_3 is 185 mm. Hg at 357° C. compared with 134 mm. Hg for UH_3, and its rate of formation from U is much slower, but attempts to separate D from H by decomposition of hydrides led to only slight D enrichment of the gas phase. 76 ref.—P. E. M.

***Systems Zirconium–Molybdenum and Zirconium–Tungsten.** R. F. Domalga, D. J. McPherson, and M. Hansen (*J. Metals*, 1953, 5, (1), 73–79).—In these systems phase relationships up to 50 at.-% Mo or W have been carefully resolved, and a limited amount of work done to outline the 50–100 at.-% alloy regions, using thermal and metallographic analyses together with incipient melting and X-ray-diffraction techniques. The alloys were prepared from " low-Hf " (<0·05%) " iodide " Zr crystal bar (99·8% Zr), 99·9% Mo, and 99·9% W, by repeated melting in a non-consumable electrode (Mo or W) arc furnace and a massive Cu crucible. Annealing treatments were carried out *in vacuo* or under A in quartz bulbs, with temp. control of ±3° C., followed by quenching in H_2O; homogenization treatments were controlled within ±10° C. The melting and heat-treatment techniques are described in detail. From the results, new partial and complete phase diagrams have been prepared, the main features of which are as follows: *Zr–Mo.*—(1) the existence of only one intermediate phase, $ZrMo_2$ (ideally 67·78% Mo), which is formed by the peritectic reaction: melt (58% Mo) $+ \gamma \rightarrow ZrMo_2$, at 1880° ± 20° C.; it has a cubic *C*15 ($MgCu_2$-type) lattice, $a = 7{\cdot}59$ Å, in agreement with the results of other workers; (2) a eutectic between β solid soln. and $ZrMo_2$, at 31 ± 1% Mo and 1520° ± 15° C.; (3) a max. solubility of Mo in β-Zr of ~22% at the eutectic temp.; (4) a eutectic decompn.: $\beta \rightarrow \alpha + ZrMo_2$, at 7·5 ± 1% Mo and 780° ± 5° C., which proceeds very slowly; (5) <0·18% Mo soluble in α-Zr at all temp., and (6) <10% Zr soluble in Mo, this solid soln. being designated γ. *Zr–W.*—This system is very similar to the Zr–Mo system: (1) a single intermediate phase ZrW_2 (ideally 80·13% W) which is formed by the peritectic reaction: melt (~50% W) $+ \gamma \rightarrow ZrW_2$, at 2175° ± 25° C., and which has, according to other workers, a structure similar to that of $ZrMo_2$, but with $a = 7{\cdot}61$ Å.; (2) a eutectic between β solid soln. and ZrW_2 at 18 ± 2% W and 1650° ± 15° C.; (3) a max. solubility of W in β-Zr of ~8% at the eutectic temp.; (4) a eutectoid decompn.: $\beta \rightarrow \alpha + ZrW_2$ at ~0·5% W and 850° ± 15° C.; (5) <0·5% W soluble in α-Zr at all temp.; and (6) <10% Zr soluble in W. Photomicrographs show typical structures. All compn. (other than those designated at.-%), are in wt.-%. 6 ref.—E. N.

Some Characteristics of Soft Magnetic Materials Used in Instrument Manufacture. G. A. V. Sowter (*Proc. Inst. Elect. Eng.*, 1951, [II], **98**, 714–727; discussion, 753–759).—S. lists the appn. of magnetic materials of various types in instrument manufacture and discusses the properties obtainable from the various materials with D.C., A.C., and combined D.C./A.C. operation. The best alloys to be employed under particular conditions are indicated. The subjects discussed in detail include the variation of *B/H* characteristics of Mumetal, complex permeability, the 50 c./s. elect. losses of the special magnetic alloys, harmonic distortion in transformers and chokes having Ni–Fe cores, distortion coeff. of these alloys, their magnetostriction and variation of magnetic properties with temp. The relative costs of soft magnetic alloys are discussed. 7 ref.—D. M. L.

Selecting Hard-Facing Materials to Resist Impact, Heat, Friction, and Abrasion. Howard S. Avery (*Product Eng.*, 1952, **23**, (3), 154–159).—Materials suitable for resistance to impact, heat, friction, and abrasion are listed, and the properties governing their choice for various appn. are described. —M. A. H.

***The Effect of Atomic Volume on the Heat of Formation of Solid Solutions of Binary Metallic Systems.** Theo Heumann (*Z. Metallkunde*, 1951, **42**, (6), 182–189).—Using simple lattice models, equations have been derived for the calculation of the distortion energy necessary for the prodn. of solid soln. which, when applied to some binary systems, have given the following values for the distortion energy (in cal.) corresponding to a compn. of 0·5 g.-atom each metal: Ag–Cu 1640, Au–Cu 1560, Ag–Au —, Au–Pt 750, Au–Ni 2120, Ni–Pt 1830, Cu–Ni 368, Na–K 440, K–Cs 250, Cd–Zn 830, Al–Zn 225, and Bi–Sb 175. The energy of distortion contributes to the total heat of formation of the alloys, and is the result of the metals having different at. vol.—a factor which is often of decisive importance in determining the properties of alloys, e.g. the occurrence of m.p. min. within a solid-soln. series. If two metals have similar chem. properties their binding energies are generally alike, and make little or no contribution to the heat of formation of alloys, which is then largely determined by the magnitude of the distortion energy. Thus, in the systems (1) Ag–Cu and Cd–Zn the distortion energies are so great that continuous solid-soln. series cannot be formed, whereas (2) in binary systems of the alkali metals K, Rb, and Cs, and in the system Bi–Sb, where the partners have very different at. vol. but small distortion energies, continuous solid-soln. series are formed. 23 ref.—E. N.

The Calculation of [Chemical] Affinities. H. Schunck (*Metall*, 1952, **6**, (3/4), 74–76).—Math. The classical method of calculating chem. affinities is described, using as an example the reaction $2Al + \frac{3}{2}O_2 \rightarrow Al_2O_3$.—E. N.

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 Rolling and Recrystallization Textures of Pure [99·5%] Aluminium.** Hans Joachim Wallbaum (*Z. Metallkunde*, 1951, **42**, (12), 371–376).—X-ray-diffraction methods have been used to determine the textures produced in 99·5% Al sheet by hot- and cold-rolling and by recrystn. The data are presented as pole figures which show that: (1) *Hot Rolling*, from 150 to 14 mm. at 550° C. produces an approx. (110)[112] texture; if, however, rolling is carried out below 450° C. (e.g. at 370° C.) some mech. twinning occurs in a position intermediate between (110)[112] and (112)[111] to produce an additional (100)[001] texture. (2) *Cold Rolling*, from 14 to 0·5 mm. produces a mixed (110)[112] + (112)[111] + (100)[001] texture; further reduction to 0·01 mm. results in a (112)[111] + (100)[110] texture. (3) *Recrystn.*—in general, recrystn. textures are statistically identical with rolling textures, but sheet $<$0·5 mm. thick and deformed between 99·5 and 99·86% shows a (100)[210] recrystn. texture which does not appear in the original rolling texture, but which can be related to the (112)[111] texture by appn. of Burgers' theory of recrystn. This (100)[210] orientation can still be observed after the onset of secondary recrystn. 17 ref.—E. N.

***Experimental Demonstration of Two Types of [Work-] Hardening in the Plastic Deformation of [Aluminium Single] Crystals. II.—Recovery of Long Crystals Strained in Tension and in Shear.** Fritz Röhm and Werner Sautter (*Z. Metallkunde*, 1951, **42**, (10), 289–293).—Cf. R. and Kochendörfer, *ibid.*, 1950, **41**, 265; *M.A.*, **19**, 531. Recovery isotherms for long (~50-cm.) 5-mm.-dia. single crystals of 99·998% Al which have been strained in tension show that only part of the work-hardening produced by the deformation is susceptible to recovery. This is confirmed by: (1) Laue back-reflection X-ray patterns, for these show that asterism—a consequence of the deformation—is still present after recovery has taken place, and (2) the results of detn. of the degree of recovery which occurs in each of the twelve possible slip systems of these (f.c.c.) crystals. Thus the work-hardening of crystals strained in tension appears to consist of two approx. equal components—a "slip" component and a "tensile" component, of which only the former is susceptible to recovery; when straining takes place by shear, however, the slip component predominates—to the extent of 90%—and no asterism can be observed after recovery has taken place. Investigations are in progress to determine whether asterism is a qual. or a quant. criterion of the presence of the tensile component of work-hardening. 10 ref.—E. N.

***Grain-Boundary Migration [in Aluminium] Induced by a Temperature Gradient.** Hideji Suzuki (*J. Phys. Soc. Japan*, 1951, **6**, (6), 522–523).—[In English]. The phenomenon of grain-boundary migration in annealed high-purity Al, induced by a temp. gradient, has been observed, and is reported. The grain boundaries move in a direction opposite to that of the temp. gradient. The rate of migration induced by a temp. gradient is faster than that due to surface free energy which takes place towards the centre of curvature of the boundary. Grain-boundary migration due to the temp. gradient occurs owing to the necessity of increasing the entropy of the grain boundary. The migration was observed under a temp. gradient of 0·2° C./cm.—J. S. G. T.

[Discussion on a Paper by F. A. Crossley and L. F. Mondolfo on:] Mechanism of Grain Refinement in Aluminium Alloys. —— (*J. Metals*, 1952, **4**, (11), 1190–1195).—Cf. *ibid.*, 1951, **3**, 1143; *M.A.*, **19**, 533. *F. R. Morral*, *M. E. Eborall* and *A. Cibula*, *H. Bernstein*, *J. Morgan*, *P. G. England*, and *D. Turnbull* and *J. H. Hollomon* contributed to the discussion which was centred around the conflicting theories of Eborall and Cibula (*J. Inst. Metals*, 1949–50, **76**, 295 and 321; *M.A.*, **17**, 914), who postulate that the impurity nuclei responsible for refinement are TiC particles, and of C. and M., who consider that the primary nucleating agent is $TiAl_3$ formed by means of a peritectic reaction, with impurities and restriction of grain growth by the concentration-gradient mechanism acting as supplementary agents. Turnbull and Hollomon considered that the authors' results support Cibula's main thesis that refinement is effected by nucleation catalysts formed in the melt by Ti and other elements. 12 ref. —E. N.

***Mechanism of Growth of Whiskers on Cadmium.** Milton O. Peach (*J. Appl. Physics*, 1952, **23**, (12), 1401–1403).—A letter. An ingenious theory is developed to account for the growth of long monocrystal "whiskers" on the surface of large Cd and Sn crystals. The basic idea is that the whiskers grow at points where screw dislocations emerge at the crystal surface. Pairs of screw dislocations are joined inside the crystal by an edge dislocation segment. Because of an "image force", this segment "climbs" by a diffusion outwards of successive rows of atoms out of the extra lattice plane. These atoms diffuse most easily along the cores of the two screw dislocations and on reaching the surface cause the whisker to build up with the (prolonged) screw dislocation at its core. Since all the atoms arrive at the surface at one point, the "Frank spiral" grows fastest near this point and is soon "pinched off"; hence the small dia. of the whisker. Various experimental features are explained and predictions made.—R. W. C.

***The Crystal Structure of Cadmium Selenide in the Selenium Rectifier.** Saburo Nagata and Kosuke Agata (*J. Phys. Soc. Japan*, 1951, **6**, (6), 523–524).—[In English]. The CdSe in the barrier layer of a Se rectifier is shown to consist of two layers, viz. a cubical cryst. layer on the Cd side, and a very thin hexagonal cryst. layer near the Se. By heat-treatment the cubical modification is transformed into the more stable hexagonal form, characterized by the smaller molecular vol. —J. S. G. T.

***The Plastic Deformation of Copper Single Crystals. I.—Two Stages in the Progress of Deformation. II.—Recovery Process and Mean Length of Dislocations.** (Kanzaki). See col. 673.

***Crystal Structure of the Ternary Phase Cu_4MgSn.** E. I. Gladyshevsky, P. I. Kripyakevich, and M. Yu. Teslyuk (*Doklady Akad. Nauk S.S.S.R.*, 1952, **85**, (1), 81–84).—[In Russian]. Cu_4MgSn has a cubic structure ($T_d^2 = F\bar{4}3m$), with $a = 7{\cdot}030 \pm 0{\cdot}002$ kX, and thus resembles the $AuBe_5$ type.—G. V. E. T.

***Neutron-Diffraction Studies of Cold-Worked Brass.** R. J. Weiss, J. R. Clark, L. Corliss, and J. Hastings (*J. Appl. Physics*, 1952, **23**, (12), 1379–1382).—Measurements have been made with neutrons, analogous to measurements of the diffuse background and of integrated intensities and broadening of spectra with X-rays. The "background" was estimated by measuring a polychromatic beam of neutrons which had passed through a polycryst. specimen. The beam contained only $\lambda > 2d$, where d = max. lattice plane spacing. Since inelastic scattering is slight, the intensity transmitted is markedly affected by imperfections. The change of intensity is consistent with a concentration of one dislocation/30 atomic spacings. The correspondence, both qual. and quant., between this neutron scattering and electron scattering resulting in a change of resistivity, is shown. Integrated intensities of the (111) and (200) peaks were measured and found to be unaltered by cold work within 1%. "Broadening" was measured indirectly by finding the variation of transmitted intensity as a function of λ. This intensity has discontinuities at $\lambda = 2d_{111}$, $2d_{200}$, &c. The sharpness of the (200) discontinuity was diminished, which can be due only to variation of lattice parameter; particle size would have no effect. Correspondingly, the (200) diffraction peak was broadened much more than the (111) peak. The difference is attributed to elastic anisotropy. The present observations are consistent with recent measurements in the X-ray field and with conclusions drawn from these, except for the inconsistent observation by Hall and Williamson (*Proc. Phys. Soc.*, 1950, [B], **64**, 937; *M.A.*, **19**, 654) that the background intensity is substantially decreased by cold work.—R. W. C.

***Torsion Texture of 70 : 30 Brass and Armco Iron.** W. A. Backofen and B. B. Hundy (*J. Metals*, 1953, **5**, (1), 61–62).—Specimens of 70 : 30 brass and Armco Fe were twisted in a torsion-testing machine up to shear strains of 3·2 and 2, resp., and their pole figures were determined by the Geiger-counter X-ray spectrometer method. The results show that the torsion texture of (1) the brass can be described by a compn. of the three ideal orientations $(11\bar{1})[112]$, $(112)[11\bar{1}]$, and (110)[001], all three of which provide a [110] slip direction along the transverse direction of max. shear stress; there is, of course, much spread about these orientations; and (2) Armco Fe appears to consist of a strong alignment of [110] directions along the specimen axis, which is limited in its range of orientation around the axis, with $(110)[1\bar{1}2]$ superimposed; there is a [111] slip direction along the transverse direction of max. shear stress. The textures established by twisting are not altered if the direction of twisting is reversed. The torsion texture is not the texture found in cold-drawn wire. 3 ref.—E. N.

***The Short-Range Order in Cu_3Au.** Takehiko Oguchi (*J. Phys. Soc. Japan*, 1951, **6**, (6), 499–501).—[In English]. In continuation of previous work, (*ibid.*, (1), 31; *M.A.*, **19**, 704), the short-range order parameters for atom sites at various distances from a given atom in an alloy consisting of A and B atoms, present in resp. proportions 3 : 1 are calculated for an alloy of f.c.c. lattice, on the assumption that the system can be regarded as an Ising ferromagnetic model, and taking into account nearest-neighbour interactions at temp. above the Curie point. The results are applied to the discussion of short-range order in the alloy Cu_3Au, and give a satisfactory qual. explanation of the short-range order parameter for 3rd neighbours in that alloy. Essentially the paper is a more rigorous treatment of the approx. theory of Cowley (*Phys. Rev.*, 1950, [ii], **77**, 669; *M.A.*, **18**, 180), whose experimental results relating to Cu_3Au are inconsistent with that given by the Bragg-Williams approximation.—J. S. G. T.

***Electron-Diffraction Study of the Ordered Alloy Au_3Cu.** Shiro Ogawa and Denjiro Watanabe (*J. Phys. Soc. Japan*, 1952, **7**, (1), 36–40).—[In English]. Thin films of Au_3Cu, prepared by evaporation and annealing at ~200° C., examined by electron diffraction, showed clear superlattice rings. Successive steps of heating up to 500° C. showed collapse of the ordered lattice and the gradual transition of diffuse superlattice rings to broad and faint bands, which persisted at temp. much above the transition temp. at ~240° C., which is above that given by Cowley (*Phys. Rev.*, 1950, [ii], **77**, 669; *M.A.*, **18**, 180). The results generally agree with those of Germer *et al.* (*Phys. Rev.*, 1942, [ii], **61**, 614; *M.A.*, **9**, 239) relating to Cu_3Au, and support results obtained from sp. heat measurements of Cu_3Au, Au_3Cu, and CuAu at higher temp.—J. S. G. T.

***Complicated Domain Patterns on Iron–Silicon Single Crystals.** C. F. Ying, S. L. Levy, and R. Truell (*J. Appl. Physics*, 1952, **23**, (12), 1339–1345).—Powder methods of delineating domain patterns were applied at intervals during the polishing, mech. or electrolytic, of Fe–Si crystals which had first been roughly polished. Complex and unusual patterns subsisted until a deep surface layer, presumed to be in a strained condition, had been removed.—R. W. C.

***Preparation of Bicrystalline Lead Plates.** F. W. Brugman and T. J. Tiedema (*Appl. Sci. Research*, 1952, [A], **3**, (3), 250–252).—[In English]. A letter. Cold-rolled strip 1 mm. thick is cut into a U-shape, having a thick base, and melted in a furnace for ½ hr. on a gypsum sledge. The protective oxide film formed during melting maintains the shape of the specimen. The sledge is withdrawn from the furnace at a rate between ½ and 1 cm./min., till crystals are formed in the "arms", after which it is withdrawn at once to obtain a polycryst. structure in the remainder of the specimen. By twisting and bending the "arms" at those places where they merge into the specimen, the crystals can be brought into a position of desired orientation with respect to the polycryst. part. The latter procedure is illustrated with the aid of stereographic projections. To convert the specimen into the bicrystal, it is melted again on the sledge beyond the deformed regions and withdrawn from the furnace at the same rate as before. Both crystals thus obtained will have definite orientations. This method is claimed to give a better yield than the alternative method worked out by Chalmers (*Proc. Roy. Soc.*, 1940, [A], **175**, 100; *M.A.*, **7**, 277). 5 ref.—S. Ma.

***A Replica Method for Study of the Structure of Lead–Antimony Alloys.** Jeanne Burbank (*J. Metals*, 1953, **5**, (1), 55–56).—A technique has been developed for the microscopic study of the three-dimensional structure of the Pb–Sb alloys, by the formation of a chem. replica in which the internal structure of the metal may be seen directly. Basically, the process consists of taking the as-cast structure and converting one of the phases into a translucent material, while the second phase remains virtually unaffected, thus permitting microscopic examination by transmitted light. The molten alloy is prepared and drawn, by means of a vacuum pump, into a dressed and preheated brass mould, then air-cooled, thereby producing a sheet 0·003–0·004 in. thick. By this technique foils are produced having microstructures similar to those in massive castings, except that, in most cases, the foils are one grain thick. The replicas are then prepared by prolonged treatment with CH_3COOH–H_2O_2 etchant. The replicas are fragile and must be handled with care; it is necessary to make all examinations while the specimens are still moist, because the translucent material becomes powdery and opaque on drying. Photomicrographs show the typical cast structures in binary Sb alloys contg. 4 and 12% Pb. The technique is to be used to investigate the effects of heat-treatment, working, and ageing of these alloys. 4 ref.—E. N.

***Crystal Structure of the Intermetallic Compound $Mg_{32}(Al,Zn)_{49}$ and Related Phases.** Gunnar Bergman, John L. T. Waugh, and Linus Pauling (*Nature*, 1952, **169**, (4312), 1057–1058).—A complete structure determination of the compound having the approx. compn. $Mg_3Al_2Zn_3$ has been made by the stochastic method and verified by X-ray data obtained from single crystals. The structure differentiates clearly between the larger atoms of Mg and the smaller atoms of Al and Zn, and indicates the ideal formula $Mg_{32}(Al,Zn)_{49}$. It is suggested that the Mg atoms in Mg_4Al_6Cu occupy the same positions as in the above, and preliminary work on the α-phase in the Al–Mn–Si system indicates a structure closely related.—R. S. B.

***Polygonization [in Nickel] During Diffusion.** R. W. Balluffi (*J. Appl. Physics*, 1952, **23**, (12), 1407–1408).—A letter. Polygonization has been observed metallographically in Ni near the diffusion boundaries in Cu/Ni sandwiches, and also in heated Ni foils exposed to Cu vapour. This confirms the X-ray results of Barnes (*Proc. Phys. Soc.*, 1952, [B], **65**, 512; *M.A.*, **20**, 226). The strain causing this polygonization is attributed to the dimensional changes which are known to occur ⊥ the diffusion direction.—R. W. C.

***Crystal Structure of $NiMg_2$, $CuMg_2$, and $AuMg_3$.** Konrad Schubert and Kurt Anderko (*Z. Metallkunde*, 1951, **42**, (11), 321–325).—An X-ray study of the following phases reveals their structure to be: $NiMg_2$—a hexagonal cell with $a = 5·18$ and $c = 13·19$ kX, $c/a = 2·54$; 18 atoms/unit cell, $d = 3·6$ g./c.c., space group D_6^4-$C6_22$, with $3Ni_I$ in 3(*b*), $3Ni_{II}$ in 3(*d*), $6Mg_I$ in 6(*f*), and $6Mg_{II}$ in 6(*i*). The structure is homotypic with that of $CuAl_2$, and is closely related to that of $CuMg_2$. $CuMg_2$.—An orthorhombic, face-centred cell, with $a = 5·273$, $b = 9·05$, and $c = 18·21$ kX; 48 atoms/unit cell, space group D_{2h}^{24}–$Fddd$, with 16Cu in 16 (*g*), $16Mg_I$ in 16 (*g*), and $16Mg_{II}$ in 16 (*f*). The structure is homotypic with that of $CuAl_2$. $AuMg_3$.—A hexagonal cell with $a = 4·63$, and $c = 8·44$ kX; 8 atoms/unit cell. It is isotypic with Mg_3Hg and belongs to the Na_3As (DO_{18}) type. The nature of the Mg bonds is discussed. 15 ref.—E. N.

***The Crystal Structure of $PtPb_4$.** Ulrich Rösler and Konrad Schubert (*Z. Metallkunde*, 1951, **42**, (12), 395–400).—An X-ray study of the phase $PtPb_4$ shows that it has a tetragonal structure, with lattice const. of $a = 6·65_{3\pm10}$ and $c = 5·96_{6\pm10}$ kX, $c/a = 0·897$. The unit cell has an at. vol. of 25·8 kX^3, and contains two $PtPb_4$ mol.; the space group is D_{4h}^3–$P4/nbm$, with 2Pt in 2(*a*) and 8Pb in 8(*m*). The structure, which is related to the $CuAl_2$ (*C*16) type, is discussed at length. 19 ref. —E. N.

***An Electron-Diffraction Study of Selenium Rectifiers.** Jiro Yamaguchi, Saburo Nagata, and Yuichi Matsuo (*J. Phys. Soc. Japan*, 1951, **6**, (6), 521).—[In English]. Structures of Se rectifiers, revealed by electron diffraction, agree with those previously found by X-ray analysis (*ibid.*, 1950, **5**, 385; *M.A.*, **19**, 718). The thickness of the CdSe layer is $0·7–1·5 \times 10^{-6}$ cm., and the roughness of the surface is of the order $10^{-4}–10^{-5}$ cm.—J. S. G. T.

***Preferred Orientations in Iodide Titanium.** Carl J. McHargue and Joseph P. Hammond (*J. Metals*, 1953, **5**, (1), 57–61).—X-ray detn. have been made of the deformation textures of "iodide" Ti (which has a h.c.p. structure) contg. N 0·002, Mn 0·004, Fe 0·0065, Al 0·0065, Pb 0·0025, Cu 0·01, Sn 0·002%, and Ti remainder. The following results were obtained: *Cold-Rolled Wire*, reduced in dia. by 91%. The $[10\bar{1}0]$ direction is ∥ the wire axis; scatter about the "ideal" is ±6° and ±15° to the wire axis in the centre regions and surface layers, resp. *Recrystallized Wire*, annealed in evacuated quartz tubes at 1000°, 1300°, and 1500° F. (540°, 705°, and 815° C., resp.). These have an average orientation contg. the $[11\bar{2}0]$ direction ∥ the wire axis, but the actual orientation has the normal to the $(21\bar{3}0)$ plane ∥ the axis. *Cold-Rolled Sheet*, reduced 97% at ~10%/pass, with the rolling direction changed 180° after each pass. The orientation describing the texture is (0002) rotated 27° and $[10\bar{1}0]$ ∥ the rolling direction, most of the spread from the "ideal" orientation being in the transverse direction. *Recrystallized Cold-Rolled Sheet.* Material annealed at 1000° F. retains the cold-rolled texture, while that annealed at 1300° F. has a dual structure—cold-rolled and recrystallized. Annealing at 1500° F. produces a fully recrystallized structure with only the texture contg. $[11\bar{2}0]$ ∥ the rolling direction, and the (0002) tilted 27° to the rolling plane. *Sheet Hot Rolled at* 1050° *F.* (565° *C.*). This has a $(0002)[10\bar{1}0]$ texture with spread more pronounced in the transverse than the rolling direction. *Sheet Hot Rolled at* 1450° *F.* (785° *C.*). (0002) is aligned ∥ the rolling plane, with a spread of ±30° in the rolling direction and ±40° in the cross direction from the "ideal" orientation; there is no preferred direction ∥ the rolling direction. *Effects of the* $\alpha \rightarrow \beta$ *Transformation on Textures.* After the double transformation $\alpha \rightarrow \beta \rightarrow \alpha$, the texture is the same as the high temp. (1500° F.) recrystn. texture. 13 ref.—E. N.

***The Preparation of Uranium Crystals.** R. W. Cahn (*Atomic Energy Research Establishment Rep.*, 1951, (**M/R744**), 18 pp.).—Fine-grained α-U may be converted to a coarse-grained structure by giving it a critical extension of 0·8–1·4% and then annealing it with its temp. slowly raised from ~500° to 630° C. The grains produced are structurally perfect to within 20′ over areas of approx. 2 mm. dia. The largest grains produced are only 2–3 mm. in dia. This is probably due to easy nucleation along twin interfaces, and a method of producing single crystals by giving a critical extension above the temp. range where twinning occurs is suggested. Long grains of α-U may be produced by cooling through the $\beta \rightarrow \alpha$ transformation, using a suitable temp. gradient to reduce nucleation. A furnace producing gradients of up to 80° C./cm. at 670° C. is described. Long crystals are produced if the specimen is moved through the gradient at a speed of 0·5–4 mm./hr. The original condition of the material appears to have no effect on the transformation which apparently occurs by nucleation and growth. The crystals produced contain a cellular sub-structure, the cells are structurally perfect and generally have a slightly different orientation from their neighbours. The size of the sub-grains appears to decrease and the disorientation to increase as the speed of the transformation is increased. It is tentatively suggested that the freshly formed α grains deform by kinking under the transformation stresses. This would explain the disorientation (which can occur in several directions) and the existence of "bridges" of sub-grains linking apparently distinct grains. The interpretation of Laue photographs of U is discussed, and an appendix contg. the angles between the more important planes of α-U is given.—P. E. M.

***The Plastic Deformation of a Zinc Bicrystal.—II.** Tomoyosi Kawada (*J. Phys. Soc. Japan*, 1951, **6**, (6), 485–488).—[In English]. Cf. *ibid.*, (5), 326; *M.A.*, **19**, 719. In continuation of previous work (see e.g., Taylor, *J. Inst. Metals*, 1938, **62**, 307; *M.A.*, **5**, 425) the stress/strain diagrams and change of appearance of Zn bicrystals, i.e. crystals each consisting of two crystals separated by a longitudinal grain boundary, were studied, taking into account the mutual interaction of the neighbouring crystals and disregarding mech. properties of the grain boundary. On this basis a satisfactory explanation of the stress/strain curve, the change of shape of the cross-section, and twisting, of the bicrystals was derived, and the experimental results, combined with those of Miller (*Trans. Amer. Inst. Min. Met. Eng.*, 1934, **111**, 135; see *M.A.*, **2**, 5) enable the stress/strain diagrams for polycryst. metals to be discussed in terms of those for single crystals. —J. S. G. T.

***Bend-Plane Phenomena in the Deformation of Zinc Monocrystals.** John J. Gilman and T. A. Read (*J. Metals*, 1953, **5**, (1), 49–55).—The inhomogeneities of deformation which occur at the restrained ends of Zn crystals, at tensile kink bands, and at compressive kink bands are described in terms of bend planes which are believed to consist of walls of edge dislocations. The bend plane bisects the angle between the slip directions on either side of it and contains the normal of the plane of the two slip directions. This is used to explain the migration of the bend plane through the lattice at an end restraint, the "anomalous" rotation of the lattice within a tensile kink band, and the geometric conditions which obtain during the formation of a compressive kink band. Experiments are described in which tensile kink bands were produced artificially in Zn (99·999+%) single crystals of triangular cross-section (side $\sim$0·2 in.) and a range of orientations, by transverse scratching of one of the flat surfaces. These lead to the conclusion that tensile kink bands form in Zn if a specimen contains a region of inhibited slip. The lag of the inhibited slip planes in the reorientation process causes bending moments on either side of the inhibited planes. These lead to bend plane formation and thence to a tensile kink band. Two new bend plane phenomena were observed in the experiments: (1) some crystals cleaved part way through their cross-sections during elongation, tongue-like pieces of the crystals projected from the rest and bent sharply backward upon further deformation; the very sharp angle between the outer edge of the bent piece and the edge of the crystal indicates a bend plane contg. the sharp angle and bisecting the angle between the slip direction of the tongue and the slip direction of the rest of the crystal; and (2) if certain geometrical conditions were fulfilled a longitudinal scratch made on one side of the crystal before deformation caused a bend plane to form $\parallel$ the tension axis of the crystal, as a result of slip inhibition near the scratch. 21 ref. —E. N.

A Modern Method for the Preparation of Metallographic Specimens. —— (*Tech. Sci. aéronaut.*, 1952, (3), 200–202).—Some details are given of two Danish instruments, the Micropol and the Disa-Electropol, for the rapid electrolytic polishing of areas of 1 mm.^{2} and 100 mm.^{2}, resp., of a wide variety of ferrous and non-ferrous metals. The former is claimed to be particularly suitable for *in situ* examination, whereas the latter is designed for the treatment of more orthodox met. specimens.—H. A. H.

***Thin-Film Interferometric Techniques for High-Magnification Topographical Studies.** S. Tolansky and M. Omar (*Nature*, 1952, **170**, (4315), 81–82).—A simple interferometric technique for the examination of surface topography, which does not require the use of opt. flats, is described. The techniques involve the prodn. of a thin transparent film of Canada balsam on the surface; the upper surface of this film acts as the reference "flat". The methods have the advantage that high magnification (up to 3-mm. objective) can be used, and inaccessible regions examined. The heights of surface discontinuities can be measured by using monochromatic illumination.—R. S. B.

***Electron-Diffraction Investigation of Polished Surfaces.** T. S. Renzema (*J. Appl. Physics*, 1952, **23**, (12), 1412).—A letter. Polished surfaces of Si, Ge, B, and Be, have been tested by electron diffraction. Two diffuse halos were obtained at similar $\sin \theta/\eta$ values to those found for other metals previously investigated. Liq. metals give halos of radii depending on atomic radius; it is therefore unlikely that polished surfaces of metals such as Be and B with small atomic radii would give halos of the same size as normal metals such as Au if the surface layer were amorphous, i.e. pseudo-liquid. Various phys. properties of polished Si surfaces altered on annealing, but the diffraction pattern was unchanged. The pattern is attributed to a strongly adsorbed surface layer of hydrocarbon, present indifferently on all metals. Various oils have given diffraction patterns similar to those from polished metals.—R. W. C.

***Regarding Sigma-Phase Formation.** David S. Bloom and Nicholas J. Grant (*J. Metals*, 1953, **5**, (1), 88).—The hypotheses advanced by Sully (*J. Inst. Metals*, 1951–52, **80**, 173; *M.A.*, **19**, 372) and by Beck and co-workers (*Trans. Amer. Inst. Min. Met. Eng.*, 1951, **191**, 872; *M.A.*, **19**, 371) are discussed—hypotheses which are based on Pauling's theories of the electronic configuration of the elements of the first transition group—and it is shown that a similar criterion can be developed without making ref. to Pauling's theories. If the number of electrons considered to exist in the $3d$–$4s$ levels of the metals V, Cr, Mn, Fe, Co, and Ni are 5, 6, 7, 8, 9, and 10, resp., and if the boundaries for the incidence of the σ structure in: V–Mn, V–Fe, V–Co, and V–Ni are 24·3, 37–57 (at 700° C.), 40–54·9, and 55–65% V, resp., Cr–Mn, Cr–Fe, and Cr–Co are 19–24 (at 800° C.), 43·5–49 (at 600° C.), and 56·6–61% Cr, resp., and Mo–Fe, and Mo–Co are 47–50 (at 1400° C.) and 59–61 (at 1500° C.), % Mo, resp., then if the number of electrons in the $3d$–$4s$ levels/atom is calculated for the σ boundary values, the numerical values are all in the vicinity of 7 electrons/atom. This is true, also, for the ternary Cr–Mo–Ni σ phase which occurs in a system in which none of the binary compn. are known to develop such a phase. The position of Mn with ref. to the other elements and to the σ phase is discussed in relation to the complex crystal structure of the metal itself and in association with interalloying elements of the first transition group, critical examination of which, it is considered, might lead to further advances in the study of the σ phase. 3 ref.—E. N.

Field of Stress Originating During the Decomposition of a Solid Solution in the Vicinity of a Spherical Nucleus of the New Phase. L. N. Aleksandrov and B. Ya. Lyubov (*Doklady Akad. Nauk S.S.S.R.*, 1952, **83**, (6), 833–835).—[In Russian]. Math. A. and L. deduce equations for the stress as a function of the distance from the centre of the nucleus; these are used to calculate the boundaries of the plastic zones in the vicinity of a ferrite nucleus formed in austenite under chosen conditions.—G. V. E. T.

The Texture of Metals after Plastic Deformation. P. Bastien (*Métaux, Corrosion-Ind.*, 1952, **27**, (319), 95–106).—B. describes the formation of textures resulting from segregates and from preferred orientation (chiefly in connection with steels), and their effects on properties. 14 ref.—M. A. H.

***Measurement of Internal Boundaries in Three-Dimensional Structures by Random Sectioning.** Cyril Stanley Smith and Lester Guttman (*J. Metals*, 1953, **5**, (1), 81–87).—It is shown, from a study of geometric probabilities—with no assumptions other than randomness of sectioning (usually realizable in experiment)—that the extent of one- and two-dimensional internal boundaries of a solid (e.g. grain-edge length and grain-boundary area in a polycryst. metal) can be calculated from measurements on plane sections. The method involves counting the number of intersections of the boundary with planes of known area or lines of known length oriented and located at random. Thus, if a line randomly intercepts an array of surfaces or interfaces, then the average number of intersections/unit length is exactly equal to one-half the true ratio of surface area to volume. Formulæ are developed which enable the following properties to be rigorously determinable from random two-dimensional sections of 3-dimensional structures: (1) average cross-section of cell or particle; (2) average linear intercept; (3) ratio of area of total two-dimensional features to total volume (surface-to-volume ratio for separate particles); (4) ratio of length of total one-dimensional features to total volume (edge-to-volume ratio for separate particles); (5) ratio of total one-dimensional to total two-dimensional features (edge-to-surface ratio for separate particles); and (6) volume fraction of particles of α phase. The following properties are determinable, also, on the assumption that the particles are of one size and spherical: (*a*) average volume of grain or particle, (*b*) number of particles/unit volume of whole sample, and (*c*) average surface area/grain or particle. The results are considered to be of importance, since quant. measurements of microstructure to obtain the volumetric relations of various phases—which have been carried out by petrographers for many years—are becoming increasingly popular among metallurgists. 11 ref.—E. N.

Isomorphism and Bond Type. W. S. Fyfe (*Amer. Mineralogist*, 1951, 36, (7/8), 538–542).—A bond between different atoms cannot be described as a purely ionic or a purely covalent bond, but only as a hybrid bond, resonating between those two forms. The tendency to form ionic bonds is great whenever the ionization potential is low and the electron affinity is high. The combination of ionization potential and electron affinity is known as the " electro-negativity " of the atom. A table of the electro-negativity of the elements and a chart of the amount of ionic character of the bond as a function of the difference in electro-negativities are given. The approx. type of bond and the co-ordination number of binary compounds of Zn, Mg, Fe with F, Cl, B, I, O, S, Se, Te, and of Mg with Si, Ge, Sn are given.—I. S. S.

Direct Determination of Interplanar Spacing from X-Ray Patterns Recorded on Charts. R. F. Shurtz (*Amer. Mineralogist*, 1951, 36, (7/8), 626–629).—A calibration curve made by plotting the interplanar spacing as a function of the angle 2θ crosswise on a strip of blank chart is a time-saving device to read the interplanar spacings directly from a recorded X-ray pattern. The calibration curve is placed above the chart contg. the pattern, and the values of the spacing are read on the curve with the help of a T-square. The method is not only convenient, but also useful to correct for slippage that accumulates between the beginning and the end of the chart.—I. S. S.

***Accurate Measurement of Lattice Constants by Means of Electron Diffraction: The Lattice Constant of KCl Referred to Au.** Norio Kato (*J. Phys. Soc. Japan*, 1951, 6, (6), 502–507).—[In English]. Precautions necessary in the accurate measurement of crystal lattice const. by the method of electron diffraction, using the technique of superimposed patterns due to Finch and Wilman (*J. Chem. Soc.*, 1934, 751), are briefly discussed. Measurement of the superposed Debye–Scherrer rings given by a film of KCl deposited on Au foil or film, annealed at 300° C., gave the following results relating to the lattice const., *a*, of KCl: (1) with KCl immediately after prodn. by evaporation *in vacuo*, *a* was 0·3– ~0·15% > the value obtained by X-ray analysis; (2) on exposure of the film to air, the deviation from the X-ray value decreased to a stationary value of +0·15%; (3) film prepared by pptn. from aq. soln. gave this stationary deviation. It is not known whether the deviation is attributable to KCl or Au; the variation of the lattice const. of Au by etching is not likely to be the cause, but it cannot be excluded *a priori*. 18 ref.—J. S. G. T.

4 — DENTAL METALLURGY

***Some Observations Concerning Chrome–Cobalt Denture Bases.** John Osborne and G. A. Lammie (*Brit. Dental J.*, 1953, 94, (3), 55–67).—The properties of dental golds and Cr–Co alloys of the type Co 60, Cr 30, Mo 5% are compared. It is shown that the Cr–Co–Mo alloys can be used successfully for the manufacture of dental bases and possess some advantages over the dental golds. The changes in base design which are possible with these alloys are discussed. The investment casting, the cleaning, and the polishing of the bases are described in detail.—S. V. R.

***A Statistical Analysis of the Mechanical Properties of Cast and Wrought Gold Dental Alloys.** S. H. Bush (*Amer. Soc. Test. Mat. Bull.*, 1952, (185), 46–50).—Using data obtained from manufacturers' charts and a number of experimental results, it was found that there was a linear correlation between proportional limit and U.T.S., proportional limit and B.H.N., and U.T.S. and B.H.N., for a range of cast and wrought Au dental alloys. The correlations are given in graphical form with the 5% limits of error. The most satisfactory relationship was between proportional limit and U.T.S. The scatter occurring when B.H.N. was involved was considered to be largely due to variations in heat-treatment, methods of making mech. tests, &c. 8 ref.

—P. T. G.

5 — POWDER METALLURGY

***Magnetic Study of Diffusion During Sintering of a Nickel–Copper Alloy [Powder Compact].** Werner Köster and Josef Raffelsieper (*Z. Metallkunde*, 1951, 42, (12), 387–391).—A report of a study of the sintering of 89 : 11 Ni–Cu compacts prepared from carbonyl Ni and electrolytic Cu powders by heating in purified H for 1–65 hr. at 800°–1100° C. or up to 25 hr. at 1200° C. Since Ni–Cu solid soln. contg. up to 34% Cu are ferromagnetic at room temp., the diffusion process was followed by magnetic methods—measurements of the magnetization-saturation value of the alloy as a function of temp.—the principles of which are outlined and are based on the work of Gerlach (*Metallforschung*, 1947, 2, 275; *M.A.*, 16, 264). The results show that the kinetics of sintering are the same irrespective of whether the sintering is carried out below or above the m.p. of Cu. Completely homogeneous compacts are obtained only after 25 hr. at 1200° C.; the activation energy of the process is 28,500 cal./mole, a value which is in good agreement with that for the activation energy of diffusion of Ni in Cu. 5 ref.—E. N.

***Degree of Dispersion of a Nickel Skeleton Catalyst.** L. M. Kefeli and N. G. Sevast'yanov (*Doklady Akad. Nauk S.S.S.R.*, 1952, 83, (6), 863–864).—[In Russian]. X-ray measurements were made of the particle size of various Ni powders made by leaching Ni_2Al_3 with NaOH; low temp. and dilute soln. favoured the prodn. of fine powders.—G. V. E. T.

***Sintering of Ultrafine Ferromagnetic Powders.** N. I. Ananthanarayanan and J. F. Libsch (*J. Metals*, 1953, 5, (1), 79–80).—A study has been made of the variation of coercive force with sintering temp. of powders—predominantly 1 μ or less in size—prepared by the low-temp. redn. of finely divided jeweller's rouge (Fe_2O_3 98 and SiO_2 2%) in purified H. The results show that with increase of temp. the coercive force decreases: (1) gradually up to 1000° F. (540° C.), (2) very rapidly—from ~500 to 70 Oe.—in the relatively narrow temp. range 1000°–1600° F. (870° C.), and (3) very gradually at higher temp. Stage (1) indicates that little sintering has occurred and that the single-domain character of the particles constituting the specimen has been essentially preserved; stage (2) marks the excessive formation of domain boundaries between the individual particles by sintering; while stage (3) is to be associated with elimination of porosity, the slow progress of which is characteristic of the latter stages of sintering. Although the use of coercive force as a parameter for sintering behaviour is limited to ferromagnetic materials, the elect. resistivity of metals and alloys—another structure-sensitive property—also undergoes a large change below a critical size, and it is considered, therefore, that the measurement of elect. resistance might perhaps form a still more general approach to the study of the building up of polycryst. aggregates from ultrafine powders by powder-metallurgy techniques. 11 ref.—E. N.

***Bubble and Zone Formation in Hard-Metal Alloys.** Walther Dawihl (*Z. Metallkunde*, 1951, 42, (7), 193–197).—Experiments have been carried out on WC–Co and WC–

TiC–Co mixtures to explain the formation of bubbles and zoned structure in hard-metal alloys, detn. having been made of the occurrence of these phenomena in relation to sintering temp., Co and TiC contents, and d of the compacts. Bubble formation occurs if a stage is reached in sintering when the gases which are generated cannot continue to escape from within the compacts, the wetting capacity of the liq. phase for the hard-metal framework is relatively large, and the gas pressure exceeds the strength of this framework. Zoning, or segregation, occurs at temp. when only small amounts of liq. phases are formed and, consequently, little wetting occurs. Soln. of WC in solid TiC helps to degas the latter, increases its susceptibility to wetting, and thereby—as does also increase in sintering temp.—lessens segregation, particularly of C. 5 ref.—E. N.

The Technique of Hot Pressing Metal Carbides, and a Semi-Automatic Hot Press. J. Rietveld (*Metall*, 1952, **6**, (3/4), 81–82).—R. discusses hot-pressing methods in relation to: (1) the nature and condition of the binder, (2) the sintering temp., and its measurement by total radiation pyrometers, (3) the applied pressure, (4) types of hot presses, including (5) a semi-automatic type of hot press.—E. N.

†Routine Quality Control in Hard Metal Production. H. Franssen (*Metall*, 1952, **6**, (1/2), 12–21).—The desirable phys., chem., and tech. properties of the more important hard-metal powder-metallurgy materials—throughout the gamut of the prodn. processes from crude ore to final hard-metal product—are tabulated, and recommended methods for their evaluation are discussed. 49 ref.—E. N.

Acceleration of Sintering in a Single Phase: Effect of Minor Additions. Serge Taevorian (*Compt. rend.*, 1952, **234**, (24), 2363–2365).—In the case of two substances A and B that form a solid soln. in the region of A, if B has a small energy of surface activation, and hence a low m.p., a small addn. of B will lead to the formation on the grains of A of a surface-diffusion layer, which will enable sintering of A to be carried out at a relatively low temp., near the m.p. of B. This process is described as "sintering by surface activation". If the treatment is sufficiently prolonged, the product will be able to withstand temp. much above the sintering temp., since the surface-diffusion layer gradually passes into the interior of the grains of A, resulting in a homogeneous solid soln. with properties similar to those of A. The addn. of B should be of the order of 0·5 mol.-%. The possibility thus exists of using on an indust. scale a new method of sintering both refractories and metals, in which no liq. phase is necessary.—N. B. V.

6—CORROSION AND RELATED PHENOMENA

***On the Rate of Solution of Aluminium in Sodium Hydroxide.** G. P. Bolognesi (*Alluminio*, 1952, **21**, (1), 27–41).—Experiments were carried out on the rate of soln. of Al (99·5% and 99·99% purity) in NaOH soln. (0·1–10N), with and without Hg. The rate was measured by the evolution of H. In concentrations of NaOH of 0·1–1N for 99·5% Al and $<5N$ for 99·99% Al, the presence of Hg retarded H evolution, while it favoured this in stronger soln. 41 ref.—I. S. M.

***The Effect of Excreta on Aluminium Alloys.** R. Butler and E. W. Skerrey (*Light Metals*, 1952, **15**, (173), 261–266).—A series of tests simulating service conditions have been carried out on Al alloys S1C, S1C anodized, NS4, NS5, HS10W, HS10WP, and HC14T, with Cu and mild steel included for comparison. Specimens were placed in sewage at a factory sewage farm, and also behind and extending below the flushing pipes in a factory urinal and in the drainage gulley of a cattle byre. The condition of the specimens after 18 months in the urinal and 3½ years in the cattle byre and sewage farm is reported. The tests at the last place are continuing. The results showed good behaviour by all the Al alloys, which compared favourably with Cu and were markedly superior to mild steel. The urinal seemed to produce more corrosion than the other two test environments. The alloy NS4 (2½% Mg) gave best all-round results, and S1C and HS10W were also good. Anodizing effected an improvement where abrasion was not involved. The other materials were more subject to pitting than was NS4; with HC14T, once the cladding is perforated penetration will be expected to be more rapid.—A. W. B.

***On the Stress-Corrosion of Aluminium–Magnesium Alloys.** Hideo Nishimura and Masato Yasui (*Suiyokwai-Shi*, 1952, **12**, (2), 34–36).—[In Japanese.] Variations of hardness of Al alloys contg. 2–8% Mg which were quenched from 450° C. and tempered in the range 150°–350° C. for various times, are reported. Addn. of Mn and Cr increase the hardness and decrease the time necessary to attain the max. hardness by tempering. Cr was more effective in increasing the resistance to stress-corrosion than was Mn. These phenomena have been examined by visual- and electron-microscopy, and the results are discussed.—Authors.

***Forms and Methods of Corrosion Attack on Aluminium Alloys. I.—Cast Binary Alloys. II.—Effect of Rolling and Heat-Treatment on the Behaviour of an [Al–Cu–Mg] Alloy.** Hans Kostron (*Z. Metallkunde*, 1951, **42**, (4), 107–110; (5), 133–137).—[I.—] Binary alloys of Al 99·5% (contg. Fe 0·3 and Si 0·2%) with Mg 5, Zn 10, Cu 5, and Si 3% were prepared and chill cast into ingots having an average grain-size of 0·020–0·025 mm. Specimens in the following conditions, as-cast, machined, annealed (heterogenized) for 16 hr. at 350° C. and furnace-cooled, and soln.-treated (homogenized) for 16 hr. at 450°–500° C. and quenched, were then exposed for 26 days to a soln. contg. NaCl 3 and H_2O_2 0·1% in a DVL alternate-immersion apparatus. The types of corrosion which resulted are described and illustrated by photomicrographs. Corrosion of castings is, except in the case of the Al–Zn alloy, greater after the casting skin has been removed by machining. The attack is usually intercryst., owing to the p.d. which exists between the ground mass of α solid-soln. crystals and the residual melt solidified around them; corrosion occurs by dissoln. of the grain-boundary substance or of the α grain boundaries in contact with it, according as to which is the less noble. The corrosion is influenced by the degree of segregation within the structure of the alloy, and is considerably affected by subsequent heat-treatment. Thus, annealing sets up local-cell effects and tends to broaden the nature of the attack, which then proceeds as a more uniform intragranular penetration; soln.-treatment tends to minimize concentration differences and thereby localizes the attack to pitting corrosion. 3 ref. [II.—] The alloy studied contained Cu 3·41, Mn 0·75, Mg 1·07, Si 0·56, Fe 0·55%, balance Al. Ingots, 140 mm. thick, were continuously cast, hot rolled to 5 mm. at 460°–470° C., annealed at 380° C., and cold rolled. Heterogenization was carried out by heating at 350° C. for 16 hr. and furnace cooling, homogenization by heating for ½ or 8 hr., according to the size of the specimen, at 500° C., followed by quenching. Mech. tests were carried out after every stage in the prepn. of the material, as were corrosion tests by alternating-immersion (10 min. exposure, 50 min. drying) for 100 days in a 3% NaCl soln. The results of the mech. tests are shown graphically, while those of the corrosion tests are described and illustrated by photomicrographs. Corrosion, generally, was found to be similar to that described above. The cast material, with a dendritic structure of supersaturated homogeneous α solid-soln. crystals surrounded by $CuAl_2$ eutectic, has the highest resistance to corrosion, which occurs in this case mainly by intercryst. attack, with a little intracryst. penetration. On hot rolling, the alloy structure is broken up, with intercryst. and intracryst. corrosion becoming general. Subsequent annealing and soln.-treatments lead to preferential intercryst. corrosion. With continued cold rolling the material becomes more oriented ∥ the rolling planes, and layer corrosion occurs.—E. N.

***Oxidation of Aluminium–Magnesium Alloys During Foundry Operations and the Protective Role of Small Beryllium Additions.** J. Calvet and V. Potomkine (*Recherche aéronaut.*, 1952, (29), 21–28).—Information on the oxidation of molten Al–Mg alloys and the protective effect of Be addn. is reviewed. Experiments with melts (without Be addn.) showed that progressive and rapid oxidation, marked by the appearance of Mg-rich protuberances on the surface, occurred as the Mg content was increased. The oxidation-inhibiting properties of various materials were tested, and Be was found to be extremely effective in much smaller quantities than were hitherto considered necessary (0·02%), e.g. 0·007% Be gave complete protection to an Al–12% Mg bath maintained at 780° C. for 2 hr. Ca, Fe, and Mn were also effective, but to a very much lesser degree. Finally, the protective role of Be in preventing the oxidation of Al–Mg alloy cast in metal or green sand moulds was studied, as well as Be losses incurred by agitation of the surface of the molten metal bath and chlorination. The latter caused appreciable Be loss.—H. A. H.

***Contribution to the Accelerated Corrosion Testing of Chromium/Nickel Electrodeposits.** Hermann Hefele (*Metalloberfläche*, 1952, [B], **4**, (2), 17–24).—[I.—] The characteristic features of the three types of phenomena—porosity and blistering, cracking or splitting, and peeling—associated with the atmospheric corrosion of Cr/Ni and Cr/Ni/Cu electrodeposits on Fe, Al, Cu, Zn, and their alloys are described and illustrated in detail. [II.—] The following widely used corrosion tests are discussed: (1) 3% NaCl spray—which is very suitable for Cr-plated Fe, Al, and Zn; (2) exposure to humid atmospheres, which may or may not be enriched in CO_2, SO_2, or NH_3; (3) coating thickness measurements—by microscopic, and magnetic methods, and by drop and jet tests; and (4) porosity testing by (*a*) boiling H_2O, (*b*) the ferroxyl method—for Cr-plated steel and Cu-base alloys, (*c*) $CuSO_4$ method—for coatings on Al- and Zn-base alloys, and (*d*) organic reagent papers. Although porosity tests give no indication of susceptibility to cracking and peeling, they have the advantage that they can be carried out quickly—in 1–120 min., instead of the 12–192 hr. (and, in special cases, 600–2000 hr.) customary for salt-spray testing. [III.—] 25 modifications of the ferroxyl test—as regards compn. of soln., mode of appn., and the results obtained by various workers—are reviewed. [IV.—] A description is given of a comparative study of a specially severe ferroxyl test, the salt-spray test, and weathering tests. This shows that although there is no connection between the results of porosity and weathering tests on Cr/Ni plated brass, there is limited agreement between the results of porosity and salt-spray tests on Cr/Ni/Cu-plated steel. [V.—] Ferroxyl tests on various types of Ni deposits are discussed. They indicate that bright Ni plating is unsatisfactory for use as an interlayer in Cr-plating, since it contains smaller but more numerous pores than a matt deposit of the same thickness. [VI.—] The use and appn. of a HCl spray test are reported. In this test the article is sprayed for 5 hr. with a 1·7–2% HCl soln. (sp. gr. 1·007–1·01) at room temp., after which it is left for 10 hr. in the spray cabinet, then washed, dried, and the amount of corrosion evaluated (simple immersion in HCl cannot be substituted for spraying). With Cr-plated Cu-base articles good agreement is found between the results of this test and those from weathering tests of 1 year's duration. The test cannot be applied to articles made of Zn-base alloys, for with these the Cr deposit dissolves away. It is concluded that, at present, porosity testing is in an unsatisfactory state of development; much work remains to be done before it will be of use for the true appraisal of service life. 20 ref.
—E. N.

***The Rate of Solution of Copper in Nitric Acid.** J. L. de Hauss (*Chim. analyt.*, 1952, **34**, (9), 185–188).—Electrolytic Cu dissolves in HNO_3 at a rate which: (1) increases proportionally as the acid concentration rises from $1N$ to $3N$, and then (2) rises suddenly when the strength is raised to $4N$ and above, (3) is lower in agitated than in quiescent soln. of the same strength, and (4) accelerates when oxidizing agents such as $K_2Mn_2O_8$ or $K_2Cr_2O_7$, or reducing agents such as KNO_2 or $FeSO_4$, are added to the acid. Cu which has been directly immersed in $4N$-HNO_3 for 5–10 min. and then transferred to acid of $<3N$ concentration exhibits passivity, probably as the result of the conc. acid causing an oxide film to form on the metal surface; this passivity vanishes (*a*) slowly with increasing time of exposure to the dil. acid soln., eventually taking place at the rate normally associated with this strength of acid, or (*b*) rapidly if the metal is retransferred to acid of $4N$ or greater strength. It is considered that these observations accord with and confirm Evans' hypothesis as to the electrochem. nature of the reaction between Cu and HNO_3. 6 ref.—E. N.

***Inhibition of the Dissolution of Copper in Nitric Acid.** S. A. Balezin and G. S. Parfenov (*Doklady Akad. Nauk S.S.S.R.*, 1952, **85**, (1), 133–136).—[In Russian]. Change-in-weight tests on electrolytic Cu in HNO_3 (2-$7N$; 8°–80° C.) showed that at 20° C. attack is rapid at concentrations $>3N$, and in $3N$-HNO_3 at temp. $>30°$ C. On agitating the acid, the rate of attack was considerably reduced; this is ascribed to the rapid removal of freshly-formed HNO_2 from the vicinity of the metal. $Na_2S_2O_3$ (10 m.mole/l.) is a better inhibitor than Na_2S or Na_2SO_3. Urea (10 m.mole/l.) is a good inhibitor; anodic and cathodic polarization curves show that its action is a cathodic process; urea (and thiourea) destroys HNO_2 as it is formed. The inhibiting action of $KMnO_4$, H_2O_2, and $KClO_3$ increases in that order.
—G. V. E. T.

***The Formation of Cuprous-Oxide-Rich Layers by the Action of Air on the Exposed Surface of Horizontally Cast Copper Wire-Bars.** Paul Klare (*Z. Erzberg. u. Metallhüttenwesen,* 1952, **5**, (9), 350–353; discussion, 353–354).—Large horizontal Cu wire-bars were cast at pouring temp. and mould temp. of 1160° and 60°–70° C., resp., with the casting-spoon stationarily disposed above (1) a mould which traversed longitudinally beneath it or (2) the centre or one end of a static mould. The O content of the surface layers of the ingots was then determined by metallographic and analytical methods; in all cases it was found to be ~0·4 and 0·1% at depths of 0·5 and 1·0 mm., resp., thereafter decreasing slowly until a const. value of 0·01% was attained at a depth of 5 mm. The O-rich skin—with the O present as Cu_2O—is considered to be formed as a result of an interface reaction between the surface of the liq. metal as it rises in the mould and the air with which it is in contact, rather than from air being sucked into the pouring stream and there reacting to form Cu_2O, which then rises to the surface of the metal during the filling of the mould. The effects of covering the mould immediately after pouring and of variations in the pouring and mould temp., the rate of pouring, and the ingot wt. are discussed.—E. N.

***Sodium Benzoate and Sodium Nitrite as Corrosion-Inhibitors in Ethylene Glycol Anti-Freeze Solutions. I.—Laboratory Investigations [of Brass and Cast Iron].** F. Wormwell and A. D. Mercer (*J. Appl. Chem.*, 1953, **3**, (1), 22–27).—The corrosion of cast Fe and of soldered brass strips in ethylene glycol soln. contg. $NaC_7H_5O_2$ and/or $NaNO_2$ was studied under conditions simulating the temp. changes in engine-cooling systems, Cu specimens being included in one series of tests. $NaNO_2$ efficiently inhibited corrosion of cast Fe under the conditions stated, but in concentrations $\geqslant$0·1% greatly accelerated corrosive attack on the soldered brass, the corrosion increasing with increasing $NaNO_2$ content. Corrosion of cast Fe, soldered joints, brass, and Cu in 20% ethylene glycol contg. 0·1% $NaNO_2$ was, however, simultaneously inhibited by the addn. of 1·5% of $NaC_7H_5O_2$.—J. R.

***Relation Between the Phase Diagrams of [Magnesium] Alloys and Their Susceptibility to Corrosion.** Yu. A. Klyachko and S. A. Shapiro (*Doklady Akad. Nauk S.S.S.R.*, 1950, **72**, (4), 707–709).—[In Russian]. Mg–Sn alloys were prepared from pure metals, homogenized, and their corrosion susceptibility in 0·$4N$-HCl assessed by measurement of the rate of H evolution. After the experiment, the total vol. of H evolved, Mg in soln., and Sn deposited were compared, and it was found that the experimental error was 0–1%. The corrodibility decreases with increase of Sn in solid soln. in Mg, increases in the 2-phase region (the increase is retarded by increasing amounts of eutectic), reaches a max. at Mg_2Sn, and falls to a min. at the Sn-rich eutectic. Mg–Al alloys were tested in 0·4 and 0·8*N*-HCl; the two sets of data show good agreement and again reproduce the features of the phase diagram.—D. M. P.

***Participation of Nitrogen in the Process of High-Temperature Oxidation of Titanium in Air.** V. I. Arkharov and G. P. Luchkin (*Doklady Akad. Nauk S.S.S.R.*, 1952, **83**, (6), 837–839).—[In Russian]. In O at a pressure of 160 mm. Hg and temp. between 700° and 1200° C., the oxidation of Ti follows the law: $\Delta G = Ae^{-b/T}$, where ΔG is the increase in weight, T the temp., and A and b are const. In air at atmospheric pressure, a similar law is obeyed at 700°–1000° C., but the oxidation increases with temp. much more rapidly above 840° C. than it does below that temp., so that above 1150° C. Ti oxidizes more rapidly in air than in O. X-ray investigation shows that in all the experiments TiO_2 is produced in the form of rutile, but there is a thin layer of TiO next to the metal. An intermediate layer of Ti_2O_3 is formed in air at 1100°–1200° C. The lattice const. of oxide films formed in O differ from those of films formed in air, but when the former are heated at 1100° C. for 24 hr. in N their lattice const. approach those of the latter. The lattice const. of films formed in air were found to change with the depth in the film. These effects are due to the introduction of N into the lattice; this can only take place at temp. above ~900° C. Other experiments were made with Fe–Ti alloys ($\geqslant$8% Ti).
—G. V. E. T.

†Chemical Properties of Uranium. (Katz and Rabinowitch). See col. 677.

***The Hydration of Some Rare Earth Oxides and Its Effect on the Atmospheric Corrosion of the Corresponding Metals.** (Mme) Suzanne Barbezat and Jean Loriers (*Compt. rend.*, 1952, **234**, (20), 1978–1980).—Cf. *ibid.*, 1950, **231**, 522; *M.A.*, **18**, 608. X-ray powder results show that the oxides Ce_2O_3, La_2O_3, and Pr_2O_3 are able to fix water and form the hydroxides, $Me(OH)_3$, which crystallize in the same hexagonal system. The dioxides CeO_2, LaO_2, and PrO_2, on the other hand, do not undergo hydration. Hydration of the sesquioxides is accompanied by a considerable increase in vol., a fact which accounts for the corrosive action of moist air on metallic Ce and La: the hydroxide, because of its large vol., cannot adhere to the film of sesquioxide, and so becomes detached, at the same time damaging the protective film below.
—N. B. V.

***Distortion Effect of the Surface Layer of a Metal During Oxidation.** P. D. Dankov and P. V. Churaev (*Doklady Akad. Nauk S.S.S.R.*, 1950, **73**, (6), 1221–1224).—[In Russian]. The deformation of the surface layer of a metal during oxidation was studied by evaporating a thin film of the metal on to a mica sheet and observing the deflection of a pointer attached to the free end of the sheet when O at 10–12 mm. Hg was admitted to the evacuated chamber. For Fe (and Ni) the deflection corresponded to a compressive stress in the oxidized layer and increased rapidly for ~30 min. before becoming steady; considerable further deformation occurred when moist air was admitted at atmospheric pressure. For Mg, the deformation corresponded to a tensile stress and became steady after ~10 min.; admission of moist air caused the deformation to disappear. For both metals the deformation due to O recurred when the air was pumped out. Dry air had no effect on either metal. On the basis of a number of assumptions, the stresses in the films are calculated and shown to be less than those expected from the ratios of molecular vol. of the metals and oxides. 4 ref.—D. M. P.

The Attack of Metals by Free Radicals and Atoms. F. W. Thompson and A. R. Ubbelohde (*J. Appl. Chem.*, 1953, **3**, (1), 27–36).—Examples of scattered reactions involving attack of ferrous and various non-ferrous metal surfaces during chem. changes in contiguous gases or liquids are discussed in terms of thermodynamics and kinetics. A transient concentration of atoms or free radicals in excess of the equilibrium concentration may occur, and may favour internal reduction of carbides, nitrides, sulphides, and oxides, leading to very high disruption pressures at internal voids in the metal. Other instances of enhanced attack owing to such concentrations are discussed. 53 ref.—J. R.

Corrosion in the Food and Kindred Industries. C. van Duijn, Jr. (*Polytechn. Tijdschr.*, 1952, [A], **7**, (49/50), 839a–842a).—Corrosion factors, based on wt. loss when in contact with a variety of foodstuffs, &c., are discussed for the following materials used in these industries: Fe and steel, Zn, Sn, Ni, Monel, red Cu, bronze, brass, Al, and Ag. Special effects of individual foodstuffs in each case are also given.—I. S. M.

Korrosionstabellen metallischer Werkstoffe, geordnet nach angreifenden Stoffen. (Ritter). See col. 734.

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7 — PROTECTION

(Other than by Electrodeposition)

Chemical and Anodic Treatments [of Al]. V. F. Henley (*Light Metals*, 1952, **15**, (168), 93; (169), 114–115; (170), 147–148; (171), 184; (172), 220–221; (173), 266–267; (174), 284–285).—A review of British, U.S.A., and French patents granted since 1946, classified under the following headings: (*a*) processes used before the main treatment (mainly chemical); see No. (168)–(169); (*b*) chem. and electrobrightening and polishing; see No. (170)–(171); (*c*) anodizing; see No. (172)–(173); and (*d*) processes following anodizing; see No. (173)–(174).—A. W. B.

***Photographic Reproduction on Oxidized [Anodized] Aluminium.** N. I. Kirillov and A. S. Kheinman (*Zhur. Priklad. Khim.*, 1951, **24**, (10), 1019–1025; also (abridged translation) *Metal Ind.*, 1952, **80**, (2), 31–33).—[In Russian]. Hard-rolled Al sheet (99·4%, 0·38% Si, 0·3% Fe), 0·3–0·5 mm. thick was degreased, cleaned by immersion for 2–3 min. at 50°–60° C. in a soln. contg. (g./l.) 46 Na_2HPO_4, 8 NaOH, 26 water glass, rinsed, and immediately anodized. Anodizing was carried out for 10–60 min. at 20°–50° C. and 1–2·5 amp./dm.^{2} in 5–20% H_2SO_4, with Pb cathodes; some work was also done with CrO_3 baths. The coating thickness was determined by sectioning and by Dahl's double-focusing method (*Z. Metallkunde*, 1939, **31**, 145; *M.A.*, **6**, 366). For sensitizing with Ag halides, the most suitable films were those obtained after 45–60 min. in 10% H_2SO_4 baths at 30° C. using an anodic c.d. of 1·5 amp./dm.^{2}. Comparison of the vol. and weight of the oxide films shows that 50% of the vol. consists of pores; to fill these pores sufficiently with AgBr the plate must be alternately immersed in KBr and $AgNO_3$ soln. for a large number of times. The number of immersions necessary is reduced by using thick films. Prolonged sensitization, elevated temp., and immersion in KI soln. causes corrosion of the films. An amidol developer is recommended, and factors affecting the colour of the image (which varies from yellow to black) are discussed. Blue or violet images could be obtained by sensitizing the film with diazo compounds.—G. V. E. T.

New Processes for the Surface Protection of Light Metal [Aluminium Alloy] Castings. —— (*Metall*, 1952, **6**, (5/6), 138–139).—The importance of thoroughly cleaning and degreasing castings before surface finishing is discussed, and the various methods of forming protective coatings are noted. If paint, lacquer, or varnish coatings are to be applied, the surface must be adequately etched beforehand; recent innovations in this field include the Alprox process and the Standofix Special Primer. The latter is also suitable for appn. to most common heavy metals and decorative alloys; it not only ensures that the lacquer coating adheres well, but also confers a high degree of corrosion-resistance, particularly to cold, heat, H_2O vapour, and salt spray.—E. N.

Lacquer Coatings on Aluminium. Max Schenk (*Metall,* 1952, **6**, (5/6), 136–138).—The lacquering of Al and its alloys is discussed; it is successful only if the coating is continuous and intimately united to the metal, and this necessitates a pretreatment process. Mech. pretreatment processes have largely been superseded; chem. processes, e.g. the M.B.V. process, are seldom satisfactory; anodic oxidation—by the Ematal, CrO_3, or H_2SO_4 processes (in that order of preference)—is generally suitable. However, when only one coat of lacquer is to be applied for purely decorative purposes rather than for corrosion protection, increasing use is being made of the Alprox process—in which the article is treated by dipping or spraying at room temp., first with an alkaline soln., for 2–15 min., to produce an oxide double layer which, after rinsing, is then partially removed by an acid soln. to leave a fine oxide skeleton; this forms such an excellent key for painting that, in practice, the finished article has a high degree of corrosion-resistance under normal conditions. —E. N.

***Hydrophobization of Metal [Copper] Powders as a Means of Protecting Them from Corrosion.** A. I. Levin and A. V. Pomosov (*Doklady Akad. Nauk S.S.S.R.*, 1950, **72**, (6), 1075–1078).—[In Russian]. Corrosion of Cu powders is due to the presence of a moisture film, and proceeds solely at the boundary between liquid and metal; if this film is prevented from forming, corrosion should not take place. Hydrophobic adsorption films of various surface-active compounds were formed on the powders, and their wettability by water, acids, or alkalies and their corrosion rates were assessed. The results are tabulated, and it was found that treatment with Na soap soln. is particularly effective. The concentration of the stabilizing substance in the washing soln. must exceed a certain value for the full effect to be realized (e.g. Na soap soln. = 0·1%). 7 ref.—D. M. P.

The Hot-Dip Galvanizing of Structural Steel Sections. R. A. Painter (*Proc. Inst. Elect. Eng.*, 1952, [II], **99**, (67), 39–46).—P. outlines the extreme importance of protective measures to prevent corrosion of steel structures exposed to atmospheric conditions, and gives an account of the different forms of protection that can be employed. An account follows of: (1) the hot-dip galvanizing process, details being given of pickling and dipping, the structure of the galvanized coat, the effect of bath temp., time of immersion, and speed of withdrawal; (2) the effect on the finished appearance of variations in the process, particularly the addn. of Al to the Zn bath; and (3) the effects of variations in the mild-steel base, viz. compn. and surface quality. The serviceability of galvanized steelwork is discussed. Tests for thickness and porosity of coating are described. Information is given on the provision of test-pieces, the effect of galvanizing on mild steel, visible faults in finished work, the care of galvanized work, and painting procedures for galvanized work. 15 ref.—D. M. L.

***The Reaction of Iron with Molten Zinc [in Galvanizing] (2).** Hideo Nishimura, Akira Teramae, Kazuo Nakayama, and Eishi Kaibuchi (*Suiyokwai-Shi,* 1952, **12**, (2), 23–25).—[In Japanese.] When Fe or steel is dipped in molten Zn, Fe–Zn alloy layers form on the surface, the reaction being influenced by various factors. The present paper reports a study of the following factors: (1) C content of steel, (2) heat-treatment of the steel, and (3) Al and Cd content of Zn bath.—AUTHORS.

Progress in the Prevention of Corrosion of Metals. W. H. J. Vernon (*Metall,* 1952, **6**, (3/4), 73–74).—V. briefly discusses recent work of the Chemical Research Laboratory, Teddington, on (1) the development of vapour-phase inhibitors—based on Na benzoate, and (2) bacterial corrosion, and the use of antibiotic inhibitors incorporated in the main protective coating, together with supplementary cathodic protection. —E. N.

Phosphate Coatings for Military Products. Norman P. Gentieu (*Product Eng.*, 1952, **23**, (2), 183–190).—Types of phosphate coatings specified by the U.S. armed services for the protection of metals in various appn. are reviewed and details of the procedure are tabulated.—M. A. H.

Phosphating Principle and Practice. —— (*Product Finishing* (*Lond.*), 1953, **6**, (2), 42–50, 53–57).—A report together with the discussion on 2 papers on phosphating presented at a conference organized by I.C.I., Ltd., at Birmingham. —H. A. H.

American Specifications for Phosphate Coatings. —— (*Product Finishing* (*Lond.*), 1952, **5**, (10), 49–53).—Short summaries of 16 U.S. Government specifications covering the use of phosphate coatings on ferrous and non-ferrous metal surfaces for protective and lubrication purposes.—H. A. H.

Prevention of Machinery Corrosion [by Temporary Protectives]. —— (*Indust. Finishing* (*Lond.*), 1952, **5**, (51), 102–104, 106, 108, 110–111, 113–115).—Extracted from "Corrosion and Temporary Protectives" published by the Shell Petroleum Co., Ltd.—H. A. H.

Sodium Benzoate as a Corrosion Inhibitor. —— (*Product Finishing* (*Lond.*), 1952, **5**, (8), 90–97).—Condensed from a booklet "Sodium Benzoate as a Corrosion Inhibitor" published by Monsanto Chemicals, Ltd.—H. A. H.

Metal Spraying in Russia: New Pistol Designs Incorporate Melting by Electric Arc and by Low-Pressure Acetylene. —— (*Electroplating,* 1952, **5**, (9), 315, 317–318).—A survey of recent Russian developments.—H. A. H.

8 — ELECTRODEPOSITION

A New Process for Electrodeposition of Aluminium. —— (*Machinery* (*Lond.*), 1952, **81**, (2074), 285–286).—See *M.A.*, **20**, 181.—C. P. F.

New Aluminium Plating Process. —— (*Indust. Finishing* (*Lond.*), 1952, **5**, (52), 220–221).—See *M.A.*, **20**, 181.—H. A. H.

A Practical Guide for Chromium Platers. —— (*Galvano,* 1952, (186), 22–24; (187), 23–25).—Methods of determining the thickness and porosity of Cr electrodeposits, and some of the health hazards associated with Cr plating, are discussed. —H. A. H.

Practical Experiences with Chromium [Electrodeposits] Without Nickel Undercoats. D. Cam (*Plating,* 1953, **40**, (1), 47).—See *Plating Notes*, 1952, **4**, (3), 100; *M.A.*, **20**, 273. —S. MA.

Some Unusual Chromium-Plating Operations. Alan W. Brown (*Machinery* (*Lond.*), 1952, **81**, (2076), 362–366).—Describes some of the work of the Cro-Engineering Division of the Cro-Plate Co., Inc., of Hartford, U.S.A. Wet-blasting permits Cr to be plated directly on to base metals. When required, a porous Cr surface can be produced by etching the base metal before plating.—C. P. F.

How Briggs Chrome Plates Stainless Assemblies. Herbert E. Head (*Steel,* 1950, **127**, (1), 82–84).—An illustrated description of the Cr plating of stainless steel at Briggs Mfg. Co., Detroit. —S. R. W.

Hard-Chrome Plating on Aircraft Parts. Gilbert C. Close (*Products Finishing,* 1952, **17**, (2), 32–36, 38, 40).—G. T. C.

Hard Chrome: What the User Should Know. —— P. Morisset (*Usine nouvelle,* 1952, **8**, (36), 29–30).—A short general account of the properties and indust. appn. of Cr electrodeposits, with some notes on the fundamental principles of Cr plating.—H. A. H.

***Contribution to the Accelerated Corrosion Testing of Chromium/Nickel Electrodeposits.** (Hefele). See col. 707.

Electrodeposits of Cobalt or Nickel Containing Phosphorus. —— (*Indust. Finishing* (*Lond.*), 1952, **5**, (54), 358–359).—See *M.A.*, **17**, 582, 669, 738.—H. A. H.

***Study of the Internal Stress in Electrodeposited Copper.** Kiyokado Nishihara and Seizo Tsuda (*Suiyokwai-Shi,* 1952, **12**, (2), 25–27).—[In Japanese]. Measurements of the internal stress in electrodeposited Cu by a contractometer show that: (1) The deflection obtained from Cu electrodeposited from an

acid SO_4 soln. is smaller than that from Cu electrodeposited from an alkali cyanide soln. (2) The effect of free acid and Cu concentration is negligible. (3) The deflection decreases when the temp. of the electrolyte is raised. At higher temp. deflections indicated an expansive stress. (4) The deflection increases when the c.d. is increased. (5) The deflection increases when organic matter is added to the acid soln.
—AUTHORS.

***The Influence of Ultrasonics on Electrolytic Processes. III.—The Influence of Frequency on the Agitation Effects of Ultrasonics [in Nickel Plating].** Albert Roll and Gerhard Schrag (*Z. Metallkunde,* 1951, **42,** (7), 197–201).—Cf. *ibid.,* 1950, **41,** 413; *M.A.,* **19,** 609. A study has been made of the effects of variations of frequency (f) and intensity (I) of ultrasonic waves impressed on a Ni-plating bath, contg. $NiSO_4.7H_2O$ 40, and Na citrate 35 g./l., having a brass cathode and Ni anode, and operating at 20° C. The impressed waves were of 560, 34, and 15 kc./s., and 3, 0·3, and 0·15 W./cm.^{2}, resp.; L.F. radiations of 100 c./s., and 10^{-4} W./cm.^{2} were also employed; the methods of generation are described and illustrated. Although current efficiency max. occurred at high c.d. under the action of ultrasonic radiations, they appeared to bear no definite relationship to the phys. const. of the radiations. Cathodic deposition potentials became less noble the higher the frequency of radiation at the given intensities. The results are discussed, and it is shown that the agitation effects $\propto I$ and $1/f$, and that there is a relationship between the λ of the radiation and the dimensions of the cathode. 8 ref.—E. N.

***The Influence of Ultrasonics on Electrolytic Processes. IV.—Effect of Ultrasonics on Bright-Nickel Plating. V.—The Effect of Ultrasonics on the Properties of Electrodeposited Metals.** Albert Roll (*Z. Metallkunde,* 1951, **42,** (8), 238–243; (9), 271–273).—Cf. preceding abstract. [IV.—] A report of the variations which occur in the nature of the deposits when ultrasonic radiations of 34 kc./s. and intensities (I) of 0·02–0·30 W./cm.^{2} are superimposed $\parallel$ the cathode in Ni-plating baths of the foregoing compn. Although the fundamental nature of the deposits is unaltered, the following differences are observed: (1) the range of c.d. over which bright deposits are obtained is shifted to higher values as I increases, e.g. at 20° C., from 3 m.amp./cm.^{2} when $I = 0$ to 40 m. amp./cm.^{2} when $I = 0{\cdot}3$ W./cm.^{2}; there is a parabolic relationship between I and the product of the c.d. and current efficiency, and (2) rough but bright deposits, which are free from porosity, are obtained over a wide range of c.d. adjacent to and below that at which the normal bright deposits are produced. The results are discussed, and it is considered that they can be explained on the hypothesis that the metal ions travelling towards the cathode acquire an oscillatory motion under the action of the ultrasonic radiation. The ions, therefore, impinge on the cathode at angles other than 90°, and there is an optimum range of angles over which bright deposits are obtained. Smaller impingement angles produce cryst. deposits, while larger angles produce the bright but uneven deposits. Formulæ are derived for the calculation of the impingement angle, which, therefore, is of decisive importance for the prodn. of bright deposits. 6 ref. [V.—] The nature of the surface, the macro and microstructures, and the microhardness of the various types of bright deposits (5–10 μ thick) obtained under the aforesaid conditions have been studied. The results show that ultrasonic radiation is analogous to mech. agitation as regards its effects on these phys. characteristics. This conclusion is confirmed by examination of Ag and Cu deposits produced under similar conditions. Cavitation effects occur when the ultrasonic waves are of such intensity that they cause the metal ions to impinge almost or completely $\perp$ the surface of the cathode. 9 ref.—E. N.

Methods of Testing Nickel-Plating Baths and Nickel Coatings. J. Elze (*Metall,* 1952, **6,** (1/2), 5–11).—Recommended methods for carrying out the following tests are discussed at some length. *For Ni coatings*: (1) adhesion, (2) porosity, (3) hardness, (4) crystal structure, (5) suitability for any subsequent Cr plating, (6) corrosion protection of the basis metal, (7) internal stresses, (8) suitability for subsequent polishing, (9) brightness, (10) flatness, (11) metallographic cross-section, and (12) influence of pretreatment on the plating quality of the basis metal. *For Ni baths*: (1) the effects of surface-active agents in rapid baths, and (2) current distribution; in addn. the following routine tests are necessary: (3) pH, (4) chem. analysis, (5) sp. gr., (6) voltage, and (7) turbidity—in regard to the efficiency of filtration. 12 ref.—E. N.

Photocolorimetric Method for Manganese, and Chromium in Nickel Electrolyte. (Ginzburg and Livshits). See col. 720.

Bright Nickel and Chrome Plating. Frank Spicer (*Indust. Finishing* (*Lond.*), 1952, 5, (52), 192–194).—S. describes his recommended procedure for the satisfactory bright Ni and Cr plating of wrist-watch backs.—H. A. H.

Substitutes for Nickel Plate: A Survey of Industrial Experience. —— (*Electroplating,* 1952, 5, (11), 361–366).—Reprinted from a publication of the British Non-Ferrous Metals Research Association.—H. A. H.

***New Electrolytes for Tin Plating.** V. I. Sotnikova and M. A. Loshkarev (*Zhur. Priklad. Khim.,* 1951, **24,** (4), 361–372).—[In Russian]. The effect of various addn. agents on throwing power and porosity was studied using baths contg. $SnSO_4$ $0{\cdot}25N$, H_2SO_4 $2N$, gelatin 1 g./l., at 20°–25° C.; 10–15 g./l. crude phenol and 30 g./l. cresolsulphonic acid were used, but the other agents (α- and β-naphthol, thymol, xylenol, and diphenylamine) were added to saturation. In most cases, the throwing power/cathodic c.d. (D_k) curve rose from 60–70% at $D_k = 0{\cdot}5$ m.amp./cm.^{2} to a max. at $D_k = \sim 1$ m.amp./cm.^{2}, fell sharply and then flattened out (at 40–50%) at $D_k = \sim 5$ m.amp./cm.^{2} but with baths contg. diphenylamine it rose again to 60% at $D_k = 20$–25 m.amp./cm.^{2}. The porosity of deposits 6 μ thick was a min. at $D_k =$ 5–10 m.amp./cm.^{2}; coatings 12 μ thick from baths contg. α-naphthol or diphenylamine at $D_k = 10$ m.amp./cm.^{2} were practically pore-free. The porosity increased on raising the temp. to 40° C. On operating baths for some days it was found that the consumption of diphenylamine or α-naphthol was 30–50 times less than that of phenol. Use of one or more of these agents at $D_k = 0{\cdot}5$–2 amp./dm.^{2} is recommended: α- and β-naphthol, thymol, diphenylamine, xylenol.
—G. V. E. T.

Electrotinning the Awkward Job. —— (*Tin and Its Uses,* 1952, (26), 10–11; also *Indust. Finishing* (*Lond.*), 1952, 5, (50), 58–60).—Pieces of Al strip are attached at suitable points to the metal requiring coating and electrotinning from the standard Na stannate bath then ensures deposition of Sn in areas which would not otherwise be covered. Al dissolves in the soln., releasing an equivalent amount of Sn; if the area of Al is small compared with the area to be plated, the c.d. on the Al surface is so high that Sn does not adhere to it and the process can continue. With lower concentrations of Na stannate, the operation can be carried out without an external current.—K. S.

"Premat": A New [Electrolytic] Galvanizing Process. R. Zirilli (*Pro-Metal,* 1952, 5, (29), 66–72).—[In French and German]. Electrolytic galvanizing, using unpublished soln. known to be strongly alkaline and to contain cyanide, yields uniform, fine-grained, bright Zn coatings in a few min. Special passivating baths form a complex colloidal surface film contg. Cr chromate which has a pleasing appearance and high corrosion-resistance. This is claimed to be the only method of galvanizing which forms a stable and corrosion-resistant film of pleasant, bright appearance.—K. S.

A Modern Electro-Galvanizing Process. A. E. Williams (*Indust. Finishing* (*Lond.*), 1952, 5, (51), 116, 118, 120–121).—A brief description of the "Bryanising" plant at British Ropes, Ltd., for the continuous electro-galvanizing (99·999% Zn deposit; 4–5 oz./ft.^{2}) of steel wires.—H. A. H.

Continuous Electro-Galvanizing. —— (*Elect. Rev.,* 1952, **150,** (3867), 3–7).—A brief description is given of the continuous electro-galvanizing plant at Maclean and Co. (Metal Windows), Ltd., Hamilton, Lanarkshire.—S. R. W.

Periodic Reverse-Current Electroplating. E. E. Halls (*Product Finishing* (*Lond.*), 1952, 5, (8), 61–66, 69–73).—A general review.—H. A. H.

The Quality Control of Electrodeposits. P. Morisset (*Rev. gén. mécan.*, 1952, [N.S.], **36**, (43), 195–201).—A general review of methods for determining the thickness, adhesion, and porosity of electrodeposited Cu, Ni, Zn, and Cr coatings, &c.—H. A. H.

***The Portable *p*H Meter: An Aid in Plating-Thickness Determination.** G. Howells (*Indust. Finishing* (*Lond.*), 1952, **5**, (54), 346, 348–349).—Relatively simple methods are described for the accurate detn. of the end-point in the B.N.F. jet-test and the Francis thickness test for measuring electrodeposit thicknesses, by using a *p*H meter as a high-resistance input millivoltmeter.—H. A. H.

Drying Electroplated Articles: Which Method? Frank Spicer (*Indust. Finishing* (*Lond.*), 1952, **5**, (54), 328, 330, 332).—A discussion on the resp. merits of sawdust drying, centrifuges, the use of box-type ovens, compressed air, or water-dispelling fluids for drying electroplated components.
—H. A. H.

Research on Electrodeposition. William Blum (*Indust. Finishing* (*Lond.*), 1952, **4**, (49), 952–954; **5**, (50), 45–48).—A review of the work that has been carried out on electrodeposition by the U.S. National Bureau of Standards.
—H. A. H.

9—ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

The Present Position of the Electrometallurgy of the Non-Ferrous Metals. Georg Eger (*Z. Erzberg. u. Metallhüttenwesen*, 1952, **5**, (7), 270–276; (8), 310–314).—A comprehensive review of recent advances (since 1946) in the electrothermal and electrolytic extraction and refining of: Cu, Ag, Au, Pb—including the Cambi and Piontelli process which uses an amido-sulphonate electrolyte and produces Pb of 99·992% purity, Zn—including the Zn amalgam electrolytic process producing 99·999% metal, Sn, Mn—the prodn. of 99·97% metal by the Bureau of Mines electrolytic process, Cr, Cd, Ni, Co, Al—including three-layer electrolysis for the refining of scrap Al and its alloys, and L.F. induction furnaces for melting ingots, Mg, and Be. Mention is made of the electrogalvanizing of wire, the direct prodn. of sheet, tube, and wire by electrolytic methods, the prodn. of metal powders, and the problems of D.C. supply. 20 ref.—E. N.

Electrocrystallization of Metals. V. V. Mikhailov (*Uspekhi Khim.*, 1949, **18**, (6), 724–736).—[In Russian]. A critical review, with 53 ref.—N. B. V.

***Anodic Solution of Gold Alloys.** J. Seifert (*Chem. Listy*, 1952, **46**, 250–251; *C. Abs.*, 1953, **47**, 54).—Soln. of Au alloys in *aqua regia* can be replaced by electrolytic soln. in conc. HCl by using the alloy as the anode and pure Au plate as the cathode and separating the electrodes by a diaphragm. Electrolysis is best carried out at 3–6 V. and 20°–35° C. A sufficient concentration of HCl is essential. The current yield is 94%.

Recent Advances in the Electrolytic Extraction of Manganese, Chromium, and Cobalt. J. W. Cuthbertson (*Chem. and Ind.*, **1952**, (48), 1165–1170).—Presented at a Symposium organized jointly by sections of the Society of Chemical Industry and the Royal Institute of Chemistry. Electrolytic extraction of Mn and Cr is encouraging the development of low-grade ore deposits. Continuous extraction of Mn can be based on its deposition from a $MnSO_4/(NH_4)_2SO_4$ electrolyte and regeneration of the spent cell liquor by leaching the ore; pilot-scale and commercial processes are described and future prospects are envisaged. Two-stage extraction of Cr, involving the prodn. and subsequent electrolysis of chromic acid, is described and discussed. The various stages of a pilot-scale process developed in the U.S.A. for the electrolytic prodn. of Co are outlined. 22 ref.—J. R.

***The Electrodeposition of Plutonium.** O. A. Cook (*The Transuranium Elements: Research Papers Part I* (*National Nuclear Energy Series*), 1949, **IV-14B**, 147–161).—Experiments were carried out on the electrodeposition of U and tracer quantities of Np and Pu in the hope of effecting sepn. of these elements. Electrolysis of $UO_2(NO_3)_2$ soln. in the presence of CH_3COONH_4 and CH_3COOH showed that the rate of deposition of U (as oxides) increased with the temp. and *p*H of the soln. and optimum *p*H values for good adherence of deposit and rate of deposition are 5–7. The metal to acetate salt ratio should be $\ngtr$0·05–1. The conditions for Np deposition are similar to those for U, and the 2 metals may be quant. codeposited. Although Pu probably deposits more easily than U at room temp., the conditions for deposition are similar so that electrolytic sepn. is not feasible. Neither Pu nor U is deposited on an Hg cathode. Attempts to codeposit Pu quant. with Mn, either cathodically or anodically, were unsuccessful. Details of baths, currents, and temp. employed are given. 24 ref.—P. E. M.

***The Chemical Properties of Elements 94 and 93 [Plutonium and Neptunium].** G. T. Seaborg and A. C. Wahl (*The Transuranium Elements: Research Papers Part I* (*National Nuclear Energy Series*), 1949, **IV-14B**, 25–38).—Describes work done at Berkeley, Calif., in 1940–42. Most of the results reported by McMillan and Abelson (*Phys. Rev.*, 1940, [ii], **57**, 1185), on Np have been confirmed experimentally. An oxidized and reduced state of both elements was recognized—probably PuO_2^{++} and NpO_2^{++} and Pu^{4+} and Np^{3+} resp.—and a table of the solubility or insolubility of various compounds is given. Neither Pu nor Np sols are reduced by Zn. Np sols are oxidized or reduced with comparative ease, Pu with difficulty, although the latter is catalysed by Ce(III)/Ce(IV) couple. Standard oxidation–reduction potentials from the metal to the reduced ionic state are $>+0{\cdot}5$ V. for both metals, and the standard potential from the reduced to oxidized ionic states is $\sim-1{\cdot}35$ V. for Np and between -1 and $-1{\cdot}4$ V. for Pu. The chem. properties of Np and Pu resemble those of Ac, Th, and U in the 4th long period rather than the adjacent elements in their sub-groups (Re and Os resp.), suggesting that a rare-earth type of group is starting at Ac. 7 ref.—P. E. M.

***Isotope Exchange Between Metallic Silver and Its Ions in Solution.** (Mme) Maria do Carmo Anta and Maurice Cottin (*Compt. rend.*, 1952, **234**, (17), 1686–1688).—Ag powder, produced either by reduction of $AgNO_3$ by NaH_2PO_2 or by electrodeposition from $AgNO_3$ soln., was allowed to remain in contact with 0·2*M*-$AgNO_3$ soln. for various periods at 100°–150° C. ^{110}Ag was used as radioactive tracer, and the activity of the metal and of the soln. was measured after the experiments. The extent of isotope exchange depends markedly on the method of preparing the Ag, and increases with time and temp. After long periods of the exchange, the powder formed by chem. reduction recrystallizes and alters in appearance.—N. B. V.

***Activation of Metal Electrode Surfaces.** N. N. Balashova, Yu. S. Tsareva, and A. T. Vagramyan (*Doklady Akad. Nauk S.S.S.R.*, 1950, **71**, (1), 73–75).—[In Russian]. Cleaned Cu and Ag electrodes were immersed in 2*N*-$AgNO_3$ for 10 min. and washed to obtain a standard passive condition. Polarization curves for the electrodes in 2*N*-$AgNO_3$ were obtained for the standard condition and after washing in KCN, which resulted in activation of the cathode. The activating effect of KCN was confirmed by observing the increase in the number of crystals formed after treatment over the number formed on the standard electrodes. The activation by KCN occurred in a few sec., whereas passivation by $AgNO_3$ required ~10 min. It was found that ethyl alcohol activated Cu, but had no effect on Ag. 12 ref.—D. M. P.

A Theory of Hydrogen Overvoltage and Its Relation to the Electrodeposition of Metals. George Dubpernell and Roberta Dubpernell (*Plating*, 1953, **40**, (1), 53–58; discussion, (2), 151–154, 159).—It is suggested that H overvoltage is probably due not to H at all, but to the electrodeposition of minute traces of alkali metals such as Na. This is possible at low c.d. owing to the depolarizing effect of the metal of the cathode (tendency to form an alloy), and perhaps to the forces of adsorption. It is shown that the back e.m.f. of polarization corresponds closely to the electrode potentials of alkali metal alloys, whereas no electromotively active form of H has been identified which is capable of generating these high negative potentials. Sp. data are given for Hg–Na and Pb–Na alloys, and it is pointed out that the metals with the highest cathodic overvoltage readily form alloys with the alkali metals such as Na. Measurements are reported with a Lucite Haring cell, using H_2SO_4 and H_2O specially purified by distillation from iron stills, and maintained out of contact with glass. A somewhat lower polarization was found with Pt electrodes at low c.d., which was distinctly increased by a small addn. of Na_2SO_4 corresponding to an amount of Na to be found due to the soln. of glass. A larger addn. of Na ion had little effect. The history of the subject is reviewed, and it is shown that the new point of view is useful and agrees with observations in many fields. An explanation is offered of the banded structure frequently exhibited by bright electrodeposits. 70 ref.—S. Ma.

***The Adsorption of Cations by Metals Demonstrated by Means of Radioactive Tracers.** Julio Palacios and A. Baptista (*Compt. rend.*, 1952, **234**, (17), 1676–1677).—A number of metals were tested with a Geiger–Müller counter after having been immersed for various times in 0·27*M*-$ZnSO_4$ soln. contg. ^{65}Zn. Au, Pd, and Pt showed small activity, indicating pure adsorption, whereas Ag, Cu, and Fe showed much greater activity, indicative of cation exchange as well as adsorption. Autoradiographs reveal that exchange begins in small isolated areas.—N. B. V.

10 — REFINING

***SnO_2 Inclusions in Refined Copper.** P. Széki and Z. Hegedüs (*Acta Tech. Acad. Sci. Hungar.*, 1952, **4**, (1/4), 293–303).—[In English]. Crystals of SnO_2 occasionally appear at the surface of cast Cu ingots from the refining furnace. Their opt. properties are similar to those of natural and artificial SnO_2. The shape of SnO_2 inclusions developed in molten Cu depends on the speed of cooling. Both hot and cold working of the metal are adversely affected by the presence of the crystals, cold working being affected to the greater extent. During hot rolling the inclusions become dislocated and aligned in rows, the main axis of the ditetragonal bi-pyramids of the SnO_2 crystals becoming aligned principally in the direction of rolling. SnO_2 inclusions have a detrimental effect upon the mech. properties of Cu, more especially by reducing the elongation and contraction ⊥ the direction of rolling.—J. S. G. T.

***Solvent Extraction of Indium.** H. M. Irving, F. J. C. Rossotti, and J. G. Drysdale (*Nature*, 1952, **169**, (4302), 619–620).—Investigations using In^{114} at a concentration of 30 mg./l. have shown that the % In halide extracted by diethyl ether from the corresponding halogen acid increase in the order $Cl < Br < I$, the maximum extraction occurring at decreasing acid concentrations in the same order. Quant. extraction of InI from HI (0·5–2·5*N*) is possible, Fe and Be remaining quant. in the aq. phase. The extraction may be effected equally well using HCl (0·5–7*N*) to which the appropriate amount of KI has been added.—R. S. B.

***Removal of Impurities from Tin by Vacuum Distillation.** J. W. Price (*Nature*, 1952, **169**, (4306), 792).—A specimen of Sn heated in an evacuated silica tube at 1000° C. and a pressure of ~0·05 mm. Hg for 4 hr. showed the impurities As, Au, Pb, and Bi to be absent. Some sepn. of the evolved metal vapours can be obtained on the cooler parts of the silica tube.—R. S. B.

Vanadium: The Possibility of Its Production from Indigenous Sources. J. Sandor (*Metallurgia*, 1952, **46**, (278), 268–274).—Most British and some imported Fe ores contain small quantities of V, which tend to be concentrated in the steel-making slag. In view of the V shortage, it is suggested that an effort should be made to recover this V, and proposals are put forward for carrying out the Fe- and steel-making operations in such a way that extraction of V from the slag would be an economic proposition.—F. M. L.

Is the Chemico-Metals Technique To-morrow's Metallurgy? Howard L. Waldron (*Eng. and Min. J.*, 1952, **153**, (6), 84–87, 176).—Refining of sulphide concentrates contg. Co, Ni, and Cu by chem. leaching to produce pure metal powders is described.—K. C. R.

11 — ANALYSIS

†Procedures for the Determination of the [Hydrogen] Gas Content of Light Metals [Aluminium Alloys]. I.—Basic Principles of the Methods. II.—Routine Procedures. III.—Laboratory Methods. H. Kostron (*Metall*, 1952, **6**, (5/6), 115–123).—A critical review. [I.—] The results of various workers on the solubility of H in liq. and solid Al and its alloys, and the basic principles of its detn. by gas-bubble formation. [II.—] Routine methods for determining H content: (*A*) qual. or semi-quant., by (1) *d* detn. on solid metal by weighing, microscopic, and radiographic procedures, (2) observation of the surface of the melt before and during solidification, (3) Pfeiffer's vacuum-casting test, and (4) observation of blister formation during the heating (at 450°–550° C.) of sheet metal—a procedure which is inapplicable to ingot metal; and (*B*) quant., by Dardel's vacuum procedure—an equilibrium-pressure method which can be applied to both liq. and solid metal. [III.—] Laboratory procedures for quant. H detn. are described: (*A*) phys. methods: (1) hot extraction *in vacuo*, and (2) degassing by ionic bombardment; and (*B*) chem. methods: (1) attack by Br under CS_2 and grav. detn. of the liberated HBr and any residual H or volatile H compounds, and (2) Sauerwald's method of displacing H by flushing the molten metal with Cl and grav. detn. of the free and combined H (as HCl) formed, after conversion to H_2O. The limitations of the various methods are discussed. Accurate and reproducible results are obtained only with the laboratory procedure of hot extraction *in vacuo*, but the method is complicated as regards both apparatus and technique. It is considered that it is urgently necessary to find a broad basis of correlation between the results of the delicate laboratory methods and those obtained from routine test procedures, if rapid advances are to be made in the study of the effects of gas content on the tech. properties of light metals. 38 ref.—E. N.

***Spectral Investigation of the Influence of the Physical State in the "Haynes Stellite Alloy No. 30".** Juan Manuel López de Azcona (*Spectrochim. Acta*, 1952, **5**, (1/2), 18).—[In Spanish]. This alloy consists of Co, alloyed with C 0·33–0·50, Cr 23–29, Ni 13–17, Mo 5–7, and Fe 2%. The investigation aimed at studying the effect of heat-treatment of the electrodes on the spectrum emitted by the sample. The spectrum was recorded during the heat-treatment, which was carried out in a special furnace, employing a source which did not alter the structure of the specimen. Emission anomalies have been observed in the range 640°–680° C., and above 880°, on heating, and on cooling from 1000° C. In evaluating the spectra, the ratios of Ni, Mo, and Cr to Co have been studied. The detailed results of the investigation will appear later.—L. D. H.

***Spectrographic Copper Standards. (A Preliminary Note on the Technique of Standardization).** D. M. Smith (*Spectrochim. Acta*, 1952, **5**, (1/2), 1–4).—[In English]. A number of

Cu standards prepared by Johnson, Matthey and Co., Ltd., are described. The standards, which are at present in ingot form, have been made by diluting the purest Cu available with three series of synthetic alloys. The method of excitation and evaluation of spectra of the standards is given, and allowance is made for residual impurities in the pure Cu, the amounts of which were determined by visual inspection; and finally by extrapolation. The ranges which it is intended to cover are: 0·05%–0·0001% Al, Sb, Bi, Cr, Ga, Pb, Ag, Sn, Co, Fe, Mn, Ni, and Si, and 0·05%–0·001% As, Be, P, Se, Te, and Zn.—L. D. H.

***Specific Reagents for Germanium.** Pierre Bévillard (*Compt. rend.*, 1952, **234**, (2), 216–217).—Further investigations into sp. reagents for precipitating Ge are described. (see Tchakirian and B., *ibid.*, 1951, **233**, 256, 1033, and 1112; *M.A.*, **19**, 673).—J. H. W.

***Spectrochemical Analysis of High-Purity Lead.** Ray C. Hughes (*Spectrochim. Acta*, 1952, **5**, (3), 210–217).—[In English]. As the properties of Pb are sensitive to very small amounts of impurities, and as chem. methods require large weights of sample, and are time-consuming, spectrochem. methods were investigated. A soln. method was used, nitrates being prepared. High-purity Pb for the synthetic standards was made by digestion with Pb acetate to remove Bi, Cu, Ag; treatment of the acetate with thiourea to remove Fe, Ag, and Cu; and ignition of the acetate to metal to remove Fe and Sb, followed by vacuum distillation. A master standard was made by dissolving the pure Pb in HNO_3, and adding impurities as nitrates, and the final standards made by diluting with pure $Pb(NO_3)_2$. The soln. were added to C electrodes, pre-burned and treated with " Halowax " and excited with a 4400-V. A.C. arc. A concentration method for increasing sensitivity of detection of impurities is described. Quant. results were obtained photometrically, and semi-quant. by visual comparison. The limits of detection were: Ag 0·01; Cu and Cd 0·03; Zn 0·05; Bi and Fe 0·3; Sn 0·1 p.p.m. Accuracy ranged from ±30% to a factor of 2.—L. D. H.

***Observations on the Paper by K. Barteld and W. Hofmann: " The Determination of Oxygen in Lead ".** J. G. McMaster (*Z. Erzberg. u. Metallhüttenwesen*, 1952, **5**, (9), 363–364).—Cf. *ibid.*, 102; *M.A.*, **20**, 194. A critical investigation of B. and H.'s sampling procedure—maintenance of a relatively large mass of Pb at the appropriate temp. until equilibrium with O has been attained, rapid solidn., and cutting out of a cube ~200 g. wt. for the analytical detn.—leads to erroneous results owing to O segregation taking place during the quenching process. When dip samples of approx. the same wt. were taken from under the stream of molten Pb (at 1000 °C.) tapped from a blast furnace—thus precluding all possibilities of PbO contamination—and immediately quenched, analyses revealed that the surface layers were O-rich compared with the core. As a consequence, 30-g. samples were taken in a similar manner and used in their entirety for O detn., without any form of surface prepn. These were found to have considerably higher O contents than those reported by B. and H., although—as McM. indicates—they used pure Pb whereas his samples were of unrefined metal. It is concluded that these results confirm Fischer and Bechtel's observations (*ibid.*, 14; *M.A.*, **20**, 194) that the values obtained when determining the gas contents of metals are often influenced by the sampling procedure, this being the probable cause of the diverse values reported in the literature. 4 ref.

—E. N.

Chemical Analysis of Light Metals [Magnesium and Magnesium Alloys]. S. T. Payne (*Light Metals*, 1952, **15**, (172), 223–224; (173), 259–260).—Established methods for the detn. of Al, Cu, Si, Fe, Mn, Ni, Zn, Pb, and Sn in Mg and Mg alloys are given.—A. W. B.

***The Spectrographic Analysis of Nickel–Cobalt Alloys in the Range 0–100%.** W. Seith and A. Kottmann (*Spectrochim. Acta*, 1952, **5**, (1/2), 36–40).—[In German]. A spectrographic method was used to determine the diffusion coeff. of solid Ni and Co; S. and K. point out that the relatively large error in detn. of high amounts is in this case of the same order as the sampling error arising from the disappearance of the metallic interface. Successive layers contg. ~20 mg. of metal were dissolved in HNO_3, and the soln. analysed. C electrodes were used, with a Feussner spark source; conditions of excitation are given, as well as the lines used for evaluation.

—L. D. H.

***The Determination of Boric Acid in Nickel-Plating Baths.** Earl J. Serfass, Robert B. Freeman, and Edwin Pritchard (*Plating*, 1953, **40**, (1), 59–62).—Report No. 1 of American Electroplaters' Society Research Project No. 2. The interference by Ni with the end-point of titration of monobasic H_3BO_3 using phenolphthalein as indicator is eliminated by precipitating the Ni as Ni ferrocyanide. Filtration of the precipitate is unnecessary. The improved method consists of: (*a*) pptn. of Ni as Ni ferrocyanide, (*b*) titration with standard alkali to Bromocresol purple end-point to determine any mineral acid present, and (*c*) conversion of weak tribasic H_3BO_3 to strong monobasic acid by the addn. of mannitol, followed by titration with standard alkali to phenolphthalein end-point.—S. Ma.

***Photocolorimetric Method for Manganese and Chromium in Nickel Electrolyte.** L. B. Ginzburg and L. Ya. Livshits (*Zavod. Lab.*, 1950, **16**, (8), 918–923).—[In Russian]. The method is based on the extinction of light in the 520–550 mμ wave-length range by permanganate soln. and 510–570 mμ by Cr-diphenylcarbazide soln., compared with 560–700 mμ for Ni-salt soln. (1) *Mn.*—The oxidation is carried out by ammonium persulphate in the presence of $AgNO_3$. Chloride ions are eliminated by evaporation with H_2SO_4. In the range 2–20 mg./l. Mn, Beer's law is obeyed in pure soln. and in the presence of Ni, &c. (2) *Cr.*—The oxidation is carried out as for Mn, and the Mn colour then eliminated by the introduction of the diphenylcarbazide; the latter is added drop by drop until decolorization is complete and then the whole of the remaining quantity added at once; failure to do this affects the results. Phosphoric acid is added to delay the formation of Fe complexes, and accurate results are obtained up to 15 min. after addn. of the reagents. Cu is detrimental and is removed by boiling with Ni powder and filtering. Beer's law is again obeyed, but calibration should be carried out against a Ni background. Full details of the procedure are given, and the detn. of both Mn and Cr from sampling to calculation of results takes 1–1½ hr.—D. M. P.

***Spectral Determination of Small Quantities of Iridium, Platinum, and Rhodium in Palladium.** A. L. Kheifits (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 696–697).—[In Russian]. Standard specimens for analysis were made up in the form of soln. contg. Pd with 0·0005–0·02% Ir, Pt, and Rh. Since the Rh was added as Na_2RhCl_6, two sets of standards were made, one with Ir and Pt, the other with Ir, Pt, and Rh, in order to avoid errors from the intensifying effect of Na. The soln. were added to spectrochem. pure C electrodes, only 0·01 c.c. being used. Both spark and Sventitsky A.C. arc spectra were used, the latter being found more sensitive to Ir and Pt. For soln. contg. Ir and Pt only, the lines used were: (*a*) in spark spectra, Pt 2929·79, Ir 2924·79, and Ir 3220·78 Å., compared with Pd 2905·54 Å.; (*b*) in A.C. arc spectra, Pt 2929·79 and Ir 2924·79 Å., compared with Pd 2928·46 Å.; the pair Ir 3220·78–Pd 3234·92 Å. was used in both arc and spark spectra. For arc spectra of soln. contg. Na_2RhCl_6, the lines used were Pt 2929·79, Pt 2997·97, and Ir 2924·79 Å., compared with Pd 2961·19 Å., together with the pairs Ir 3220·78–Pd 3127·97 Å., and Rh 3396·85–Pd 3424·71 Å. The limit of detection for Ir and Pt was 0·001% in spark spectra and 0·0005% in arc spectra, with an error of ±3–5%. In the soln. contg. Rh the error was ±8–10%. Na was shown to intensify the Pd lines relative to the others.

—G. B. H.

***Determination of Sodium in Welding Flux for Aluminium Alloys.** [—I.] Hirozo Kimura (*Suiyokwai-Shi*, 1951, **12**, (1), 5–8).—[In Japanese]. From experiments K. found that Zn uranyl acetate reagent is more sensitive than Mg uranyl acetate in the detn. of Na when the soln. is not stirred or shaken. Li in soln. precipitates as a triple salt with Na, and consequently the Na value is too high. Fluorides do not interfere.—Author.

***Polarographic Determination of Tin in Tin-Plating Solutions.** Rafael Diaz (*Plating*, 1953, **40**, (1), 45–46).—The method is based on the conversion of stannate to hexachlorostannate with $HClO_4$ and NaCl and the subsequent reduction of the hexachlorostannate ion which gives a well-defined current at ~−0·47 V. Effective control of plating soln. can be achieved, as only a few min. are required for a detn.—S. M.

***Quantitative Spectrographic Analysis of the Rare-Earth Elements.—IV.–VI.** Velmer A. Fassel, Harold D. Cook, Laird C. Krotz, and Paul W. Kehres (*Spectrochim. Acta*, 1952, **5**, (3), 201–209).—[In English]. Cf. *J. Opt. Soc. Amer.*, 1949, **39**, 187; *M.A.*, **17**, 96. The detn. of the following rare earths is described: Ce, Pr, and Nd in La; La, Pr, and Nd in Ce; La, Ce, and Nd in Pr, and Pr in Nd. A D.C. arc source was used, as the very similar phys. properties of the earths gave a high degree of compensation for excitation variables. In selecting lines, owing to the complex spectra involved, and lack of intense lines, difficulty was encountered in finding lines free from interference, and sometimes in identification. A list is given of lines stated to be assigned wrongly in the M.I.T. tables. Methods of sampling, and of making corrections for residuals, are given, with a list of the line-pairs used in the analysis. Chem. checks could not be made in most cases, but some fluorescimetric detn. have been made, and agree well with the spectrochem. results. —L. D. H.

***The Influence of the Energy Liberated in the Discharge and of the Character of This Discharge on the Emission of the Spectral Lines.** G. Brucelle (*Spectrochim. Acta*, 1952, **5**, (1/2), 11–17).—[In French]. The effect of an increase of energy in a controlled spark unit by increasing capacity and voltage, or in the case of an arc by increasing the current, is to emit ionized lines of greater excitation potential; B. applies this fact to the detn. of trace amounts. Although the spark may be used, the source mainly employed is a stabilized arc. The example given is of steel analysis, but the method is of general appn.—L. D. H.

12 — LABORATORY APPARATUS, INSTRUMENTS, &c.

(See also "Testing" and "Temperature Measurement and Control".)

A High-Temperature Stage for the Polarizing Microscope. E. A. Wood (*Amer. Mineralogist*, 1951, **36**, (9/10), 768–772).—W. describes a stage that consists essentially of a heated metal mass surrounded by a mass of insulating material. The max. temp. is 1000° C., the max. heating rate is 1° C./sec., and the max. cooling rate is 10° C./min. A diagrammatic sketch and a photograph of the instrument are shown. —I. S. S.

The Differential Thermal Analyser as a Micro-Calorimeter. M. Wittels (*Amer. Mineralogist*, 1951, **36**, (7/8), 615–621).—Calorimetric measurements of the order of 10 millical. can be made by a method of differential thermal analysis in the range 300°–1100° C. A low-mass, radiation-type vacuum furnace, operated at high heating rates, is used. The instrument is calibrated by dissociating pure calcite.—I. S. S.

Electron Microscopes. Roger Simonet (*Usine nouvelle*, 1952, **8**, (42), 29–30, 47).—A short survey of electron microscopes and their uses.—H. A. H.

The Electron Microscope in Metallurgy. Hans Mahl (*Metall*, 1952, **6**, (3/4), 69–73).—Plastic, oxide film, and double-layer (polystyrene–SiO or SiO_2) replica techniques are described, and some typical appn. are illustrated by electron micrographs of steels, and Al and its alloys. A method of obtaining a plastic replica of an area previously selected by observation with an opt. microscope is discussed, as is the use of shadow-casting techniques for surface-roughness measurements. 25 ref.—E. N.

Some Improvements in the Apparatus for Spectrochemical Analysis. H. Bückert (*Spectrochim. Acta*, 1952, **5**, (1/2), 5–8).—[In German]. A survey is given of recent improvements in auxiliary equipment for medium-dispersion spectrographs. First, the increased capacity of controlled-spark generators, which in post-war models range from 10,000 to 12,000 pF., has led to an increase in accuracy, and by decreasing both the pre-sparking time and the exposure, has led to an important saving of time. The use of the interrupted arc has been advanced by several improvements; particular ref. is made to an instrument to be described in a later issue, where the arc current has been raised to 15 amp., and a large number of variations of arcing and test times is possible, and where D.C. and A.C. continuous arcing can also be used. Reproducibility of results is at least as good as that obtained with spark sources. The combined comparator and densitometer is described, in which photoelect. cells give improved sensitivity with accuracy over a large range of densities. Finally, improvements in spectrograph design and equipment are mentioned briefly, including the use of a curved prism in high-dispersion instruments, and direct-reading attachments.—L. D. H.

Direct-Reading Spectrograph: Rapid Metallurgical Analysis with a New French Instrument. ——. (*Aircraft Prodn.*, 1952, **14**, (165), 253–254).—A new direct-reading spectrograph of French design and manufacture is described. The equipment incorporates an electronic measuring device for selecting, receiving, amplifying, and recording.—W. A. M. P.

13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

Apparatus for Tension Testing at Sub-Atmospheric Temperatures. E. T. Wessel and R. D. Olleman (*Amer. Soc. Test. Mat. Bull.*, 1953, (187), 56–60).—An apparatus is described for carrying out tension tests at temp. from room temp. to −196° C., which, it is claimed, meets the requirements for low-temp. work better than previously described equipment. N vapour is used as refrigerant, the specimen being surrounded by N vapour during test, undesirable contact with liquid thus being avoided. Predetermined test temp. are readily obtained and accurately stabilized with an automatic control and recording system. The apparatus is simple and economical to operate. It may be adapted to other types of mech. test. The possibility of testing at even lower temp. is briefly discussed. 6 ref.—P. T. G.

***Bending as a Testing Method for Cast Light Alloys.** F. Gatto (*Alluminio*, 1952, **21**, (1), 15–25).—G. considers results of the transverse bending test and compares them with tensile test results. The former compare favourably with the latter, especially where alloys of low ductility are concerned. In its present form the test could be applied to the control of foundry work. In the method used by G. the test-piece was supported near its ends at points 100 mm. apart, and loaded 50 mm. apart at two symmetrical intermediate points.—I. S. M.

An Improved Creep-Measuring and Recording System. J. W. Huffmann (*Product Eng.*, 1952, **23**, (1), 172–174).—In the system described, strain is measured by a linear variable differential transformer, in which the difference in voltage of two co-axial secondary coils varies linearly with the position of a common steel core, whose movement is controlled by the specimen extension.—M. A. H.

***The Problem of the Effect of Size [of the Specimen] on the Proof Stress and the Fatigue Strength in Bending.** Alfred Schaal (*Z. Metallkunde*, 1951, **42**, (9), 279–284).—X-ray detn., using a specially designed camera, have been made of the stresses set up by tension and by bending, in specimens of C and Cr–Mn steels and a Duralumin-type alloy (contg. Cu 3·6 and Mg 1%) of square or rectangular cross-section. Annealed, strain-free specimens were found to have a higher P.S. in bending than in tension, the difference increasing with decreasing cross-section of the test-piece; internal stresses were also greater in bending than in tension. Similar phenomena were obtained with specimens possessing initial internal stresses, but the differences were found to be considerably smaller. The results are discussed, and are considered to be due to the variations of stress which occur between the surface and the interior of specimens deformed by bending; the stresses within the deeper-seated layers are lower than those at the surface and, if still within the elastic limit, tend to prevent flow taking place in the zones of higher stress. These observations are in close agreement with those of Morkovin and Moore (*Proc. Amer. Soc. Test. Mat.*, 1944, **44**, 137), who found a similar relationship between the bending fatigue strength and the specimen dia. 13 ref.—E. N.

***Determination of the Young's Modulus of Metal Bars by an Electromagnetic Method.** Jean Savornin and Georges Renucci (*Compt. rend.*, 1952, **234**, (21), 2045–2046).—A method is described which is suitable for both magnetic and non-magnetic materials. A rod of the metal, held at its middle point by a rubber ring, is set into vibration by an alternating magnetic force near one end, and its natural frequency, N, is determined by means of a piezoelect. quartz, connected to an oscilloscope, at the other end. Young's modulus (E) is then obtained from the formula: $N = \frac{1}{2}L\sqrt{(E/d)}$, where L is the length of the specimen rod and d its density. The Young's modulus of Cd, determined in this way was $5{\cdot}67(\pm 0{\cdot}05) \times 10^{11}$ C.G.S. units.—N. B. V.

***The Problem of Compensating for Inaccuracies in the Measurement of Internal Stresses [by G. Sachs's Boring Method] in Hollow Cylinders and in Tubes.** H. Bühler and W. Schreiber (*Metall*, 1952, **6**, (3/4), 77–80).—Cf. *ibid.*, 1951, **5**, 53; *M.A.*, **19**, 222. Math. Certain theoretical aspects of the method are further discussed and revised, in order to compensate for inaccuracies of measurement in the experimental technique. As an example of their practical appn., the experimental results for a hollow steel cylinder are evaluated in detail. 3 ref.—E. N.

The Testing of Fine Wires for Telecommunication Apparatus. R. C. Woods and J. K. Martin (*Proc. Inst. Elect. Eng.*, 1951, [II], **98**, (64), 529–536; discussion, 536–538).—An outline is given of the existing acceptance and quality tests for fine wires. The actual value of many of the tests is critically discussed. Particular attention is given to abrasion and elect. tests, and details are given of a pinhole test for enamelled wire. It is shown that in this test much depends on the technique. Most of the defects in the enamel were attributable to faults on the surface of the Cu, the most common being spheroidal indents up to 0·002 in. in dia. across the crater. Longitudinal scratches and rough patches also caused trouble. Examination of the wire with a 30× binocular microscope was very laborious and slow.

—D. M. L.

A New Ultrasonic Testing Apparatus and Its Practical Use. J. and H. Krautkrämer (*Métaux, Corrosion-Ind.*, 1952, **27**, (318), 89–94).—An instrument developed by the authors is described with illustrations of results. Equipment includes photographic recording. Defects at depths as little as 8 mm. below the surface are detected.—M. A. H.

Ultrasonic Flaw-Detector: Details of Recent Developments in the Echo Method of Flaw-Detection. —— (*Aircraft Prodn.*, 1952, **14**, (162), 126–129).—The technique of using ultrasonic waves for flaw detection is discussed. A new instrument—the Ultrasonoscope—is described.

—W. A. M. P.

Ultrasonic Vibrations and Their Metallurgical Applications. K. F. Alder (*Australasian Eng.*, **1952**, (Dec.), 53–59).—Non-destructive testing of metals and soldering of Al are the only present commercial uses of ultrasonics in metallurgy. Other appn. that have been investigated and that may be further developed include: degasification of molten metals, prodn. of alloys of normally non-miscible metals, and the grain refinement of castings. 28 ref.—T. A. H.

Photo-Elasticity Test-Benches. André Martin (*Métaux, Corrosion-Ind.*, 1952, **27**, (319), 135–142; discussion, 142).—The requirements to be fulfilled are listed, and descriptions of various types of set-up are given.—M. A. H.

Surface-Crack Detection: Details of the Met-L-Chek Liquid Crack-Detection Process. —— (*Aircraft Prodn.*, 1952, **14**, (170), 431).—A brief description of a non-magnetic, penetrant dye method of surface crack detection.

—W. A. M. P.

Fluorescent and Magnetic Inks for Flaw Detection. —— (*Mech. World*, 1953, **133**, (3403), 59).—The uses of three fluids are given as follows: (*a*) Supramor, electromagnetic flaw detection ink for the detection of internal and surface flaws in magnetizable metals; (*b*) Glo-mor fluorescent ink for indication of surface cracks and porosity in any metal; and (*c*) Lumor, a fluorescent magnetic ink.—S. Ma.

15 — FOUNDRY PRACTICE AND APPLIANCES

The Casting of Non-Ferrous Ingots. (Aitchison and Kondic). See col. 733.

Aluminiumguss in Sand und Kokille. (Irmann). See col. 736.

***Oxidation of Aluminium–Magnesium Alloys During Foundry Operations and the Protective Role of Small Beryllium Additions.** (Calvet and Potemkine). See col. 707.

[Current] Practice and Recent Developments in Aluminium [Alloy] Chill Casting [Gravity Die-Casting]. Gustav Lauterjung (*Metall*, 1952, **6**, (5/6), 129–133).—The gravity die-casting of the following articles of large and complicated shape is described and illustrated: housings for elect. equipment, oil-sumps, and air and water-cooled cylinder heads for all types of internal-combustion engines, radiators, &c. The prodn. of riffled tubes by continuous-casting processes is mentioned. The compn. of the alloys used, and the tech. and economic advantages of the new developments are discussed.—E. N.

Reducing Ejector Failure in Die-Casting Dies. H. K. and L. C. Barton (*Mech. World*, 1953, **133**, (3402), 34–35).—An improved ejector design, which greatly increases the operating life of die-casting dies, is described.—S. Ma.

Die-Casting Dies: The Unit System. Hiram K. Barton (*Mech. World*, 1953, **133**, (3403), 78–79).—The unit system enables a large machine, such as would be required for occasional jobs, to be kept employed at full capacity by the simultaneous use of a number of existing smaller dies mounted on adaptor bolsters. The main points that have to be considered are: secure attachment, preservation of an effective seal where the runner channels in the die and bolster meet, and provision of an adequate means of operating the existing ejector plate.—S. Ma.

Casting Drop-Hammer Dies. —— (*Aircraft Prodn.*, 1952, **14**, (160), 59–64).—The use of medium- and high-expansion plasters for producing close-tolerance patterns is described. Controlled solidification is also discussed.

—W. A. M. P.

Precision Casting or the "Lost-Wax" Process. M. Pretto (*Rev. gén. mécan.*, 1952, [N.S.], **36**, (45), 269–274).—A general survey of the "lost-wax" precision-casting technique.

—H. A. H.

Shell Moulding. —— (*Mech. World*, 1953, **133**, (3402), 22–23).—A review describing the methods used in shell moulding and giving some indication of its possibilities. Cf. *M.A.*, **20**, 658.—S. Ma.

Permanent Ceramic Moulds for Non-Ferrous Casting. D. E. Hope (*Found. Trade J.*, 1952, 93, (1887), 503–504).—H. describes the materials, apparatus, and technique required to make refractory moulds. The potentialities and limitations of the process are also discussed.—G. J. S.

Applications of Synthetic Resin [in the Foundry]. D. N. Buttrey (*Light Metals*, 1952, 15, (173), 271–272).—The characteristics of urea-formaldehyde and phenol-formaldehyde resins suitable for core binders and shell moulds are discussed with ref. to their tech. and economic possibilities in the light-alloy foundry. See also *Metal Ind.*, 1952, 81, (4), 70–71. —A. W. B.

Synthetic Resins as Sand Bonds. P. G. Pentz (*Found. Trade J.*, 1952, 93, (1895), 729–733).—After listing the properties of both urea-formaldehyde and phenol-formaldehyde resins, P. discusses the appn. of these resins to core-binding, flow problems, bench life, core drying, and some properties of the resin-bonded core are described.—G. S. S.

Synthetic Resins in the John Harper Foundry. J. W. Dews and P. H. Drury (*Found. Trade J.*, 1952, 92, (1864), 535–539, 550).—D. and D. report the results of investigations using both urea-formaldehyde and phenol-formaldehyde resins for core binding. The economic and prodn. advantages are stressed, suitable mixes recommended, and prodn. details such as knock-out characteristics, curing temp. and times, moisture content, washes and spraying, parting materials and phys. characteristics given. Several prodn. examples are discussed and illustrated.—G. J. S.

Better Cores at Lower Cost by Electronic Baking. Victor E. Hillman (*Iron Age*, 1950, 166, (13), 67–69).—The advantages of electronic baking of cores with synthetic resin binders as determined over a 2-year period at Crompton and Knowles Loom Works, Worcester, Mass., are given. A comparison is made between these cores and those oven baked. The costs and phys. properties of electronically baked cores with different types of resin binders are tabulated.—S. R. W.

Measuring the Properties of Sand and Moulds by One-To-Ten Ramming. C. A. Sanders and A. G. Clem (*Amer. Foundryman*, 1952, 22, (4), 48–50).—By subjecting the standard sand test-piece to a varying number of rams and measuring hardness, green strength, permeability, and *d* at each stage, graphs were obtained which indicate the likely behaviour of the sand in moulding.—V. K.

***Hot Strength [of Sand] at Falling Temperatures: Does It Influence Hot-Tear Formation?** D. C. Williams (*Amer. Foundryman*, 1952, 22, (5), 47–50).—The hot strength of sand was measured at rising and falling temp. up to 2400° F. (1314° C.). The results show that this strength is considerably affected through glass formation, which may account for the resistance of sand to the contracting metal leading to hot-tears.—V. K.

Fool-Proof Sand Works for a Wide Range of Castings. T. S. Schumacher (*Amer. Foundryman*, 1952, 21, (6), 54–57).—Practical tests are described showing that the moulding sand with at least 10% over 4 sieves (50, 70, 100, and 140 mesh) gives better moulding properties (flowability) as well as the least amount of casting defects.—V. K.

Easier, Surer Identification with Coloured Core Sands. L. C. Voss and W. W. Lynn (*Amer. Foundryman*, 1952, 22, (1), 44–45).—Advantages of using organic dyestuffs for colouring core sands are discussed. These include: sepn. and segregation of different mixtures, checking on the uniformity of mixing and temp. of drying. It is sufficient to add 1–2 oz. dye/ton of sand.—V. K.

Recent Developments in Foundry Equipment. — Blankenhorn ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, 1951, 347–355).—[In French]. B. discusses the latest improvements in the foundry equipment field.—G. J. S.

How to Mechanize a Small Foundry. W. A. Morley (*Amer. Foundryman*, 1952, 22, (5), 53–58).—Principles of mechanization are described, and the stages of mechanization of a foundry are explained with an illustrative example.—V. K.

Non-Ferrous Foundry. —— (*Mech. World*, 1952, 132, (3401), 262–263).—A short description of a non-ferrous foundry engaged on the manufacture of instrument parts. —R. W. R.

Simple Steps Toward Organizing a Foundry Safety Programme. W. N. Davis (*Amer. Foundryman*, 1952, 22, (4), 42–44).—Causes of accidents in foundry work are analysed and remedies suggested.—V. K.

Making the Foundry Safe for Top Production. Stanley C. Herbine (*Foundry*, 1952, 80, (11), 120–123, 277–279).—H. discusses the prevention of accidents in foundries, and emphasizes the necessity for educating all operatives in the value and use of protective equipment.—R. W. R.

Proper Ventilation Improves Foundry Working Conditions. —— (*Foundry*, 1952, 80, (11), 218, 220).—A description is given of the ventilating plant in an American foundry. —R. W. R.

Well-Engineered Ventilation System Keeps Bearing Foundry in Business. J. C. Soet (*Amer. Foundryman*, 1952, 22, (5), 43–46).—Ventilation problems of a bearing bronze foundry are described from the point of view of minimizing Pb toxicity.—V. K.

Three Ways to Control Foundry Dust: Isolation, Substitution, Wetting Agents. H. G. Meiter (*Amer. Foundryman*, 1952, 22, (6), 53–54).—Isolation of dust-producing operations, substitution of non-toxic for toxic dust-forming materials, and the use of water for dust control are discussed.—V. K.

The Job Record Card, a Valuable Tool for the Jobbing Foundry. R. W. Griswold (*Amer. Foundryman*, 1952, 22, (3), 34–36).—A method of recording process data is described. —V. K.

Costs and Quality Can Be Controlled by Wage-Incentive Plan. C. J. Pruet (*Amer. Foundryman*, 1952, 22, (6), 58–59).—Principles of a wage incentive plan are outlined.—V. K.

French Foundry Centre Completes Extensive Laboratories. Vincent Delport (*Foundry*, 1952, 80, (12), 106–107, 263).—A short description is given of the new laboratories at Sèvres of the Centre Technique des Industries de la Fonderie. —R. W. R.

Foundry Practice in the Indian Naval Dockyard, Bombay. A. E. Hook and J. J. Freitas ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, 1951, 301–312).—[In English]. H. and F. describe the layout and the techniques used by this foundry. Both ferrous and non-ferrous castings are made in sand or by the centrifugal-casting process.—G. J. S.

Technical Education in the Danish Foundry Industry. Ove Hoff ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, 1951, 81–83).—[In French]. H. describes the present stage of organization of education for apprentices, foremen, and engineers in Denmark. The principle of this education is the simultaneity of the theoretical courses at school with practical apprenticeship in the foundry.—G. J. S.

Management's Responsibility in Training. J. D. Judge (*Amer. Foundryman*, 1952, 22, (1), 47–50).—Training problems in industry are discussed.—V. K.

Are Technical Graduates Getting Adequate Training in College? Hiram Brown (*Amer. Foundryman*, 1952, 22, (3), 56–58).—A broad discussion of present-day issues in training of graduates in science.—V. K.

Meeting Competition with Good Cost Methods.—I. Albert E. Grover (*Foundry*, 1952, 80, (12), 102–103, 256–262).—The importance of a good costing system in the foundry is stressed, and the setting-up of a suitable scheme is described. —R. W. R.

Meeting Competition with Good Cost Methods.—II. Albert E. Grover (*Foundry*, 1953, 81, (1), 108–111, 229, 231).—How to proceed in actual calculation of foundry job costs and selling prices is detailed. G. also warns against certain unreliable practices which can lead to erroneous estimates in casting prices. Specimen lay-outs of cost sheets for various foundry processes are given.—G. J. S.

19 — WORKING

***Fundamental Study on the Working of Metallic Materials. III.–IV.—Cold Rolling of Duralumin and Super-Duralumin.** Joichiro Hunakoshi (*Suiyokwai-Shi*, 1950, **11**, (10), 537–542; 1951, **12**, (1), 39–42).—[In Japanese]. A report is given of an investigation into the rolling pressure, mean deformation resistance, relation between work-hardening and age-hardening, rupture, spread, and elongation during cold rolling of Duralumin and Super-Duralumin.—AUTHOR.

The Production of Aluminium and Aluminium Alloy Tubing. T. F. McCormick (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 17–31; discussion, 31).—The processes for the prodn. of Al alloy tubing in a modern plant are described, including: remelting of scrap, casting of hollow ingots, preheating, inspection, interior reaming, hot extrusion, cold reduction, cold drawing, heat-treatment, and cleaning. 11 ref.—D. M. P.

***Forging of Arc-Melted Chromium.** H. L. Gilbert, H. A. Johansen, and R. G. Nelson (*J. Metals*, 1953, **5**, (1), 63–65).—High-purity electrolytic Cr plate was crushed to −60 mesh, leached with HNO_3 to remove any Fe picked up during crushing, rinsed with H_2O until neutral, and vacuum dried. The powder was then reduced by treatment with purified dry H for 16 hr. in a furnace maintained at 1100°–1200° C.; after cooling, the ¼–⅛-in. lumps from the partially sintered powder were picked out, and arc-melted (under a He–A atmosphere) in an H_2O-cooled Cu crucible, using a W or, preferably, a consumable Cr electrode. The resulting ingots, 4 in. long and 2 in. dia., were of extremely high purity; after a H anneal for 1 hr. at 1200° C. and scalping, they could be worked open or, preferably, sheathed in an Fe pipe. Initial forging was carried out at 800°–850° C. with moderate reductions until the cast structure was broken up, when heavier reduction, by forging, rolling, or swaging, was performed at 500° C. The final desired form, after annealing in H at 1200° C., exhibited ductility at considerably lower temp. than heretofore described, although it could not be bent at room temp. without fracture. 9 ref.—E. N.

The Production of Copper and Copper Alloy Tubes. H. Y. Bassett (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 32–55; discussion, 55–56).—A description of the processes in the prodn. of tubing from the raw material to the crated article, including: melting and casting by various methods, hot extrusion, piercing and cupping, cold drawing and reduction, straightening, cleaning, plating, polishing, annealing, inspection, and packing.—D. M. P.

The Production of Lead Tubes. G. O. Hiers (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 69–82).—A review. The prodn. of Pb sheathing and pipe and the chem. and phys. requirements for satisfactory behaviour are described. 44 ref.—D. M. P.

The Production and Properties of Magnesium Alloy Tubing. C. J. Huffman and G. Ansel (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 102–107).—Mg alloys suitable for tubing and the extrusion processes are described. The properties of the extrusions are considered, and it is noted that these can be improved by cold working on a laboratory scale. 2 ref. —D. M. P.

The Production of Nickel and High-Nickel Alloy Seamless Tubing. W. A. Dickinson and H. F. Hendershot (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 83–100; discussion, 100–101).—The processes in the prodn. of Ni alloy tubing are described, including: forging of ingots, hot extrusion, cold reduction and drawing, annealing and inspection. Tools, lubricants, and the characteristics of the materials at various stages are considered. 4 ref.—D. M. P.

Machining Titanium: Preliminary Investigations into Machining Qualities. —— (*Aircraft Prodn.*, 1952, **14**, (159), 28–29).—A brief note discussing turning, milling, and cut-off operations based on data obtained from a report published by the Curtiss-Wright Corp., U.S.A.—W. A. M. P.

Grinding Titanium: Interim Results on Research into Grinding Wheels and Techniques of Grinding. —— (*Aircraft Prodn.*, 1952, **14**, (165), 238–239).—A discussion of the interim results obtained from some experimental work on the grinding of Ti and Ti alloys. The variables in a precision grinding operation are studied, and suitable grinding wheels are recommended for various grinding operations. —W. A. M. P.

When to Use Impact Extrusions [of Al or Zn]. —— (*Product Eng.*, 1950, **21**, (6), 106).—S. R. W.

Upset Blades: Application of the Omes Electro-Forging Process to the Production of Gas-Turbine Rotor and Stator Types. —— (*Aircraft Prodn.*, 1952, **14**, (165), 227–229).—An electro-upset forging process is described, in which the stock is automatically upset in an anvil and vice after heating by a low-voltage current. The subsequent forging is carried out adjacent to the upsetting machine without further heating of the stock. The appn. of the process to the prodn. of gas-turbine rotor and stator blades using materials such as Nimonic 90 and 80A, stainless steel, and Al bronze is discussed.—W. A. M. P.

High-Pressure Rubber-Die Forming. —— (*Aircraft Prodn.*, 1952, **14**, (165), 224–226).—A description of a new-type rubber-die, directly-acting hydraulic press, which operates at 5000 lb./in.².—W. A. M. P.

Rotary Stretch-Forming: Forming Circular Components from Section or Sheet Material. —— (*Aircraft Prodn.*, 1952, **14**, (163), 183–184).—A description of a contour-forming machine and a radial draw-forming machine for forming circular components from section or sheet material. —W. A. M. P.

The Application of the Theory of Plasticity to the Shaping of Metals: The Rolling of Thin Strips. J. Lerebours-Pigeonnière (*Métaux, Corrosion-Ind.*, 1952, **27**, (319), 119–134).—The results of researches on the mechanism of deformation in rolling, and the theories advanced by various workers, are reviewed. 20 ref.—M. A. H.

A Note on the Back-Pull Factor in Strip-Drawing. R. Hill (*J. Mechanics Physics Solids*, 1953, **1**, (2), 142–145).—The results of a recent paper by Green and H. (*ibid.*, 1952, **1**, (1), 31; *M.A.*, **20**, 377) are extended to the case where back-tension is applied. The conclusions show that values of the die pressure obtained differ appreciably from those given by other current theories, particularly when the reduction or coeff. of friction is large.—E. O. H.

The Theory of Tube-Producing Methods. E. J. Ripling (*Tube Producing Practice* (*Amer. Inst. Min. Met. Eng., Inst. Metals Div. Symposium Series No.* 4), **1951**, 1–16; discussion, 16).—A qual. discussion of the factors governing the piercing, extrusion, cupping, and sinking processes in tube forming. Tube drawing is dealt with on a more quant. basis. 15 ref. —D. M. P.

Wax-Filamented Model Billets Aid Study of Material Flow. J. L. Holmquist (*Product Eng.*, 1952, **23**, (4), 158–163).—The flow of material in the Mannesmann process is studied by piercing in a special mill a wax billet contg. coloured wax filaments, and by observing their changes of shape and distribution.—M. A. H.

The Wire-Drawing of Metals. M. Bonzel (*Métaux, Corrosion-Ind.*, 1952, **27**, (320), 143–149).—B. describes the mechanism of flow through the die, the stresses involved, energy requirements, die forms, tests of drawability, and details of practice.—M. A. H.

Procedures and Metals in Chipless Manufacturing. E. R. Tichaur (*Australasian Eng.*, **1951**, (July), 84–92).—After an historical review of the development of metal forming, the theory of plastic deformation, the various factors that affect plastic deformation, and the theoretical explanation of some manufacturing processes, e.g. rolling, forging, drawing, pressing, and extrusion, are discussed.—S. R. W.

A New Heavy-Duty Milling Machine with Cutters Attached to Swinging Arms, for the Removal of Casting or Rolling Skins from Ingot or Strip. K. Schiffer (*Metall*, 1952, **6**, (1/2), 22–24).—An illustrated description of the construction, operation, and appn. of this Junkers machine. It has twin milling cutters, each of which is 30 mm. across, driven electrically, and fixed to the bottom of an inclined arm which is pivoted to the top housing, and hydraulically loaded to control the depth of cut (up to 1 mm.); each arm can be operated as an independent unit. Milling is carried out on the top surface of the ingot or strip, and chips are removed by suction. Operating data are presented for a variety of non-ferrous alloys.—E. N.

20—CLEANING AND FINISHING

(See also " Protection " and " Electrodeposition ".)

The Anodic Surface Treatment [Electrolytic Brightening] of Light Metals. A. Benoit (*Usine nouvelle*, 1952, **8**, (41), 53).—Anodizing and the electrolytic brightening of light metals are briefly discussed, together with some of the associated troubles. It is suggested that many of these can be avoided by consulting member firms of the French Syndicat National du Revêtment et du Traitement des Métaux.—H. A. H.

Aluminium Paints. R. H. Warring (*Indust. Finishing* (*Lond.*), 1952, **5**, (54), 334, 336–338).—A survey of the advantages of Al as a pigment for use with vehicles such as varnishes, lacquers, bitumens, or silicones in the manufacture of paints for appn. to metal or wood surfaces.—H. A. H.

Pretreatment Etch Primers of the Single-Solution Type [for Brass and Aluminium].—I.–III. E. E. Halls (*Indust. Finishing* (*Lond.*), 1952, **5**, (52), 185–186, 188–189; (53), 286–292; (54), 367–372).—Wet-heat and salt-spray test results are given for 3 proprietary single-soln. and 1 two-soln. etch primers applied to brass and Al. Considerable differences in behaviour were found for the single-soln. etch primers, which generally were not as good as the two-soln. material, although they showed considerable promise.—H. A. H.

Finishing Zinc and Aluminium Die-Castings: Some Comparisons of Methods in Britain and America. —— (*Product Finishing* (*Lond.*), 1952, **5**, (11), 83–87; also *Indust. Finishing* (*Lond.*), 1952, **5**, (53), 296–299).—A summary of the report published by the Anglo-American Council on Productivity on a visit to America of a specialist team from the British Zn and Al die-casting industry (see *M.A.*, **20**, 542).
—H. A. H.

Zinc Metal-Finishing Systems. —— (*Product Finishing* (*Lond.*), 1952, **5**, (12), 83–89).—Notes on the cleaning, pretreatment before painting, and protective and decorative chem. finishing of Zn-base alloy die-castings.—H. A. H.

Finishes for Zinc [Die-Castings]. —— (*Die Castings*, 1950, **8**, (5), 49–50).—A brief review of the methods used to finish Zn die-castings.—S. R. W.

Clock Components: Polishing, Plating, Painting, Lacquering. —— (*Product Finishing* (*Lond.*), 1952, **5**, (12), 42–49).—A short account of the Cu, brass, Cr, and Ni (dull and bright) electroplating and painting of alarm-clock components.
—H. A. H.

Effective Pretreatment [of Metals] Is Vital. —— (*Product Finishing* (*Lond.*), 1952, **5**, (11), 51–55).—Notes on chem. and mech. methods for the treatment of ferrous and non-ferrous metals before painting.—H. A. H.

Cleaning Metals. Frank Spicer (*Machinery* (*Lond.*), 1952, **81**, (2075), 328–331).—A practical survey is given, describing chem. methods: alkaline and acid; phys. methods: solvent degreasing and emulsion cleaning; and mech. methods: grinding, scratch-brushing, barrelling, shot-blasting, and abrasive cleaning.—C. P. F.

***Comparative Studies on the Polishing Power and Economics of Polishing Papers.** G. Pahlitzsch (*Metalloberfläche*, 1952, [A], **6**, (4), 53–61).—The properties of various grades of polishing papers have been studied in relation to: (1) the type, grain-size, and dispersion of the abrasive medium, the nature of the cementing agent, the type of backing (e.g. paper, fibre, &c.) and its stiffness, and (2) the operating conditions, i.e. the size of the paper, the speed and time of rotation, the pressure, the angle and arc of contact between the paper and the work, and the nature and type of surface of the work-piece. A machine for the experimental study of the polishing process under the prescribed conditions is described and illustrated. The results are discussed and analysed mathematically in relation to the cost of polishing. They show that since the quality of paper varies greatly from one manufacturer to another, routine laboratory testing (including the detn. of the optimum operating conditions)—although giving only an approx. indication of service performance—is of great importance for maintaining costs at a low level in an industrial polishing establishment.—E. N.

Abrasive Finishing. R. H. Warring (*Product Finishing* (*Lond.*), 1952, **5**, (9), 38–43).—An elementary account of barrel finishing, rumbling, tumbling, &c.—H. A. H.

Backstand Grinding and Polishing [of Metal Surfaces]. J. Rhodes (*Product Finishing* (*Lond.*), 1952, **5**, (9), 48–54).—A general account of the backstand-idler method for the polishing and grinding of ferrous and non-ferrous metal surfaces.—H. A. H.

Barrel Finishing [of Metals]. Frank Spicer (*Indust. Finishing* (*Lond.*), 1953, **5**, (55), 399–400, 402–404, 406).—A review.—H. A. H.

Barrel Finishing: A Case History. —— (*Product Finishing* (*Lond.*), 1952, **5**, (12), 64–72).—The development is described of satisfactory barrelling techniques for the prodn. of a mech. finish suitable for subsequent bright Ni plating on some Zn-base die-cast, brass, and stainless steel motor-car components.—H. A. H.

Vacuum Evaporation [of Metals]: Some Practical Considerations. —— (*Indust. Finishing* (*Lond.*), 1952, **5**, (52), 206, 208–209).—Some of the practical difficulties associated with the appn. of metals to plastics, glass, pottery, and Zn-base alloy die-castings by the vacuum-evaporation technique are described.—H. A. H.

Vacuum Deposition Becomes a Trade. —— (*Product Finishing* (*Lond.*), 1953, **6**, (1), 62–67).—A short description of the processes carried out by V.M.C., Ltd., who specialize in the vacuum coating of plastics or metallic objects by evaporation.—H. A. H.

Dewatering Fluids [in Metal-Finishing Processes]. —— (*Product Finishing* (*Lond.*), 1952, **5**, (10), 80–88).—A fairly comprehensive account of the use of dewatering fluids in metal-finishing processes such as phosphating, anodizing, electroplating, &c. Their properties, methods of appn., and benefits are described.—H. A. H.

Protective [Hand] Creams in the Finishing Shop. E. E. Halls (*Product Finishing* (*Lond.*), 1952, **5**, (9), 56–59).—The use of barrier creams as a safeguard against dermatitis is described.—H. A. H.

21 — JOINING

Metal Bonding: Developments in the Redux Process: Application of the Vacuum Technique. G. S. Newell (*Aircraft Prodn.*, 1952, **14**, (165), 220–223).—N. gives a brief account of the development of the Redux metal-bonding process, and describes in detail a new technique in which atmospheric pressure is used for applying the bonding pressure. The components to be joined are laid on steam-heated platens and covered by a rubber blanket which is evacuated. This technique is satisfactory for joints which have good contact and do not require high pressure to bring the components together. Test results are given. 17 ref.—W. A. M. P.

The " Al-Fin " Process. E. Herrmann (*Aluminium Suisse*, 1952, (4), 115–125).—[In French and German]. The " Al-Fin " process depends upon immersing a ferrous-base component in molten pure Al at 830°–880° C. for a time sufficient to form an Fe–Al layer. To improve its ductility and to facilitate subsequent bonding, the component is reheated, preferably in a molten Al–Si alloy, in the range 540°–675° C. The treated shell is then placed in a sand mould or metal die and the outer shape cast. A number of examples of steel liners and inserts bonded to Fe by this process are given.—A. W. B.

Flux-Bath Dip-Brazing of Aluminium Alloys. D. Wallace (*Product Eng.*, 1952, **23**, (2), 173–175).—Details are given of the procedure and the filler materials and alloys suitable for the dip-brazing of Al alloys.—M. A. H.

Brazed-Aluminium Heat-Exchangers. —— (*Modern Metals*, 1951, **7**, (9), 25–28).—The appn. of a brazing method in the prodn. of Al heat-exchangers is described. Special brazing Al sheet, consisting of a core of pure Al coated on one or both sides with a brazing alloy (92·5% Al and 7·5% Si) is used. The exchangers are assembled and held in a steel jig, and after preheating to a temp. just below the m.p. of the brazing alloy (100° F. (38° C.) lower than the m.p. of Al) they are plunged into a flux bath held at ∼1100° F. (595° C.). The brazing alloy melts and bonds the several separate components into a rigid structure.—S. R. W.

Brazing Costs as Affected by Seven Basic Brazing and Weld-Brazing Methods. —— (*Product Eng.*, 1950, **21**, (6), 132–133).—S. R. W.

Materials and Procedures for Soldering Aluminium. G. W. Birdsall (*Materials and Methods*, 1952, **35**, (4), 116–118).—Ultrasonic, flame, and soldering-iron methods are discussed, and a range of solders contg. Sn (50–75%), balance Zn, are recommended, together with suitable fluxes. The importance of thorough cleaning is stressed, and a number of techniques for cleaning and soldering are described. The complete removal of flux after soldering is recommended, and methods for the prevention of electrolytic corrosion enumerated.
—R. P. H. F.

Finishes for Soft Soldering.—I.–II. E. E. Halls (*Product Finishing* (*Lond.*), 1952, **5**, (10), 68–79; (11), 76–79).—[I.—]. The solderability of hot-dip Sn or solder coatings; electrodeposited Cd, Sn, Ag, Pb, Sn–Pb, and Sn–Zn alloys; fused electrodeposited Sn or Pb–Sn alloy coatings; and a combination of electrodeposit and hot-dip solder are discussed. Their appn. to Zn- and Al-base materials receives special attention. [II.—] It is not good practice to solder directly to Zn castings, and it is recommended that such castings be Ni-plated before applying the solderable finish—electrodeposited Sn, Ag, or solder dip. Direct appn. of Sn or Cu coatings to the Zn surface is not favoured.—H. A. H.

Spot Welding in Highly Stressed Aluminium and Magnesium Assemblies. A. Schoellerman and S. Jenkins (*Product Eng.*, 1952, **23**, (3), 183–185).—Data are given on design factors, including accessibility to electrodes, matching materials of various compn., relation between thickness and elect. resistance of dissimilar materials, and positioning and size of welds.—M. A. H.

The Technique of Constructing Large Aluminium Containers [by Welding]. —— (*Alluminio*, 1952, **21**, (1), 65–88).—A report from the Istituto Sperimentale dei Metalli Leggeri. Design, fabrication, jointing, surface protection, installation, and testing of Al containers are described, with special ref. to welding technique. 42 ref.—I. S. M.

***Determination of Sodium in Welding Flux for Aluminium Alloys.** [—I.] (Kimura). See col. 720.

Lead Welded by Inert-Gas-Shielded Arc Process. —— (*Materials and Methods*, 1952, **36**, (1), 97).—The Heliarc welding method for joining Pb flashing units, with and without the use of a filler rod, is described.—R. P. H. F.

Welding [of Nimonic 75] and Assembly of Jet-Propulsion Units. —— (*Machinery* (*Lond.*), 1950, **76**, (1956), 555–562).—A description of various A-arc welding operations on Nimonic 75 sheet.—S. R. W.

Welding [Nimonic 75] Sheet-Metal Components for Jet-Propulsion Units. —— (*Machinery* (*Lond.*), 1950, **77**, (1969), 79–82).—A description is given of various precision-welding operations on Nimonic 75 sheet involved in the prodn. of components for jet-propulsion units.—S. R. W.

Resistance and Fusion Welding of Titanium and Its Alloys. E. F. Holt and W. H. Moore (*Weld. J.*, 1952, **31**, (3), 213–216).—Techniques for the spot, flash, and inert-arc welding of Ti and its alloys are reviewed. 14 ref.—K. B.

Flash-Welding Titanium Alloys. I. A. Oehler (*Materials and Methods*, 1952, **36**, (3), 206, 208, 210).—Typical welding cycles and mech. properties of welded joints in Ti alloys are given. Materials having a uniform structure are said to have suitable welding characteristics. The use of techniques similar to those for Al are advocated, and the annealing of parts immediately after welding to prevent cracking is recommended.—R. P. H. F.

Thoriated-Tungsten Electrodes: Their Welding Characteristics and Applications. G. J. Gibson and R. O. Seitz (*Weld. J.*, 1952, **31**, (2), 102–108).—The theory of arc welding is examined to determine the difference between thoriated- and pure W electrodes. The use of thoriated electrodes reduces time losses due to contamination, poor arc striking, and electrode adjustment. Welding conditions are given for Cu and Al with He and A shielding gases. 3 ref.—K. B.

Cold-Pressure Welding of Wire. Edmund D. Sickels (*Wire and Wire Products*, 1952, **27**, (12), 1298–1299, 1323–1324).—Describes a process developed by the General Electric Co. which is applicable at present only to non-ferrous metals whose grains flow easily under deformation stresses, especially Cu, Al, and many of their alloys. Some success is also claimed with Cd, Pb, Ni, and combinations thereof. Wires butt-welded under cold pressure can be drawn, and after annealing will resume a completely normal grain structure.
—C. P. F.

Welded and Brazed Parts. H. R. Clauser (*Materials and Methods*, 1952, **36**, (5), 123–138).—The welding and brazing of Cu and its alloys, Al and its alloys, Mg and its alloys, Ni alloys, and Fe and steel are reviewed. The strength, corrosion-resistance, appearance, machinability, and thermal and mech. treatment of welded joints in these materials are discussed. The considerations of joint design and cost of welding are also reviewed.—R. P. H. F.

Welding of High-Heat-Resistant Materials. J. L. Solomon (*Weld. J.*, 1952, **31**, (3), 233–238).—Resistance-welding techniques for Nimonic, Inconel, Ti, and other heat-resistant metals are discussed. Welding conditions are governed by their high U.T.S., high resistivity, and low thermal conductivity. Three-phase equipment is recommended. 4 ref.
—K. B.

Schlieren Analysis of Inert-Gas Arc Shields. W. B. Moen and G. J. Gibson (*Weld. J.*, 1952, **31**, (3), 208–213).—The Schlieren method of making gases visible is applied to a comparison study of He and A as arc shields with non-consumable electrode equipment. The results indicate that either gas will provide adequate protection. Choice of gas depends only upon the desired elect. characteristics and economics.—K. B.

25 — BOOK REVIEWS

The Casting of Non-Ferrous Ingots. By Leslie Aitchison and Voya Kondic. 8½ × 5½ in. Pp. xxi + 370, with 71 illustrations. 1953. London: Macdonald and Evans, Ltd. (42*s.*)

An up-to-date and comprehensive text-book on the casting of non-ferrous ingots has long been a notable gap in the literature, and the authors have attempted to fill it with the present book.

In their Section I, on the theoretical background to the subject, they deal briefly with the properties of liquid metals and at greater length with the manner in which metals and alloys solidify in practice. Accounts of volume changes and the associated phenomena occurring during solidification, including segregation effects, are followed by a short treatment of the factors affecting the cooling rate of ingots and the theoretical treatment of the heat-transfer problem.

The second section of the book deals with certain metallurgical aspects of the casting of ingots for working, including brief accounts of the melting and casting characteristics of non-ferrous metals and alloys, and descriptions of various types of moulds and mould materials and of methods of pouring. The section concludes with accounts of the macrostructures, soundness, and other metallurgical characteristics of ingots and of methods for examining these.

The final section deals with the production aspect, including the design of moulds, selection and control of pouring operations, and current practice in ingot casting in the non-ferrous industry. This section includes accounts of the layout of casting shops and of inspection and control procedures.

In their preface to the book the authors state that: "The absence of large numbers of references to published work may be noted and may possibly raise questions in the minds of readers. . . . We believe that our digestion of the publications and our presentation of the contemporary knowledge should relieve our readers from the necessity of referring to the original sources which we have consulted exhaustively." There is no doubt in the reviewer's mind that in so far as student readers are concerned the authors have largely achieved their object, and Section I will be appreciated also by those already engaged in this field of non-ferrous metallurgy. For the latter readers, however, the treatment of metallurgical and production aspects in Sections II and III is too superficial, and the reviewer is convinced that these readers will not be relieved of the necessity of referring to original published work. To quote an example, these readers might reasonably hope to find some critical appraisal of published work on the heat-flow problem in modern continuous or semi-continuous casting processes, and to be referred to the very sketchy account of heat-flow theory in an earlier chapter of the book, when the question arises as to what amount of cooling water must be used, will not be found very helpful. Again such readers may feel that the four pages (pp. 257–260) devoted to an elementary account of the use of thermocouples and millivoltmeters, and the two (pp. 262–263) devoted to a description of a tundish, might have been devoted more profitably, for example, to a critical consideration of the factors determining the thickness of ingot moulds.

The authors acknowledge that their treatment of the subject necessarily involves a certain amount of repetition in Sections II and III and this could scarcely be avoided, but more terse statements on some elementary points in the latter sections would have sufficed in the balanced and concise account which the authors aimed to provide. Finally, the reviewer, at least, is irritated by the practice of printing occasional words in heavy type, such as: "The demand for these various grades of copper arises from the fact that their **properties**—physical and electrical—**are different**" and "The diameter of the **holes in the tundish is regulated** in order **to secure that the flow of the metal** into the mould shall be accomplished with the lowest attainable accompaniment of turbulence".

Notwithstanding these criticisms, however, the book is a useful general introduction, and will be read with interest and profit by newcomers to the subject.

W. A. Baker.

Chemische Technologie. Band 5. **Metallurgie und Allgemeines.** Herausgegeben von Karl Winnacker und Ernst Weingaertner. 24·5 × 17 cm. Pp. xxvii + 846, with 311 illustrations and 117 tables. 1953. Munich: Carl Hanser-Verlag. (Brosch., DM 71.50; geb., DM 74.50.)

The first portion (some 550 pages) of this book may be regarded as a handbook of extraction metallurgy supplemented by notes on the uses of the metals and the properties and applications of the more important alloys of the various metals discussed. Theoretical considerations are dealt with relatively briefly, and the main object appears to have been to provide a record of recovery and refining practice.

The section on mineral-dressing is short, partly because much of the equipment and many of the processes have been dealt with in earlier volumes of the series. Though not sufficiently detailed for the specialist, the outline is more informative and useful than the accounts found in many books of similar scope.

The chapters dealing with extraction metallurgy have been contributed by a number of different authors and, as might be expected, vary considerably in the amount of detail included. The apportionment of space between the different groups of metals is unusual, and to British readers may be unattractive. In particular, the extraction and refining of the heavy non-ferrous metals and the precious metals receive relatively little attention. For example, lead smelting and refining practice is disposed of in some 17 pages, and zinc in less than 12 pages. On the other hand, the chapters devoted to the alkali and alkaline-earth metals, magnesium, and aluminium are excellent. They contain a remarkable amount of interesting and valuable data, are well-illustrated, and are supplemented by many references to other sources of information.

The remainder of the book is devoted to matters of interest to the industrial metallurgist and chemical engineer. The subjects discussed include the protection of metal surfaces, measuring and controlling instruments, and the supply, handling, and treatment of essential services and sources of energy. Planning of plant layout and "safety" in chemical-engineering plant are treated, and the final chapter is devoted to patent law and related matters.

"Winnacker–Weingaertner" is well-produced and is a most valuable addition to metallurgical literature. If the chapters on the heavy non-ferrous metals and the precious metals were as comprehensive as some of the earlier chapters, it would rank as one of the most satisfactory handbooks of extraction metallurgy available.

G. L. Evans.

Korrosionstabellen metallischer Werkstoffe, geordnet nach angreifenden Stoffen. Von Franz Ritter. Dritte Auflage. 25 × 17·5 cm. Pp. [iv] + 283, with 29 illustrations. 1952. Wien 1: Springer-Verlag. (58*s.* 6*d.*)

In these corrosion tables Dr. Ritter has collected together quantitative and qualitative data on the corrosion of solids by different media. The original edition appeared in 1937 and although a further printing was made in 1942, most of the copies were destroyed by fire. In this third edition a new impression has been made of the second edition, the results of new corrosion investigations being collected in an appendix.

The tables are arranged alphabetically according to the corroding media. While most of the references are to solutions of inorganic salts, organic liquids and gases, data are also given for corrosion in a wide variety of media including molten metals and salts, alcoholic drinks, and foodstuffs such as butter and margarine. In the list preceding the tables, the materials are arranged according to their composition and not according to their commercial names, although these are also included. Over nine hundred metals and alloys are listed and referred to in the tables by their number as well as their commercial name. Besides these, however, information is also given in the tables on the corrosion of solids such as rubber, wood, and glass, as well as thermoplastics and resins.

All numerical results have been calculated in units of g./m.^{2}/day and mm./year. Where numerical results are lacking, qualitative estimates are given, with brief descriptions of the corrosion behaviour.

The collection and tabulation of this mass of corrosion data was obviously a monumental task, and Dr. Ritter is to be congratulated on the result of his work. The reviewer has had cause to refer to this book on several occasions and has rarely been disappointed with the information provided. There are, however, no references to bacterial corrosion nor to the effectiveness of protective coatings, although the author has promised to remedy the latter omission in a future edition. Most of the thousand references given at the end of the book are to German authors, no mention being made, for example, of the elegant work of Bengough and his colleagues on corrosion in chloride solutions.

The book should be of value to chemists and engineers in the selection of working materials. It has been beautifully printed and bound by Springer-Verlag, and its cost will be amply repaid in the saving of time required to obtain the desired information.

George Butler.

Technische Temperaturmessungen. Von H. Lindorf. 21 × 14·5 cm. Pp. 168, with 123 illustrations and 21 tables. 1952. Essen: Verlag W. Girardet. (Geb., DM 15.60.)

Here is a technical book on an important technical subject, written by a technical scientist, as I think it should be, for use by works' technicians concerned with the measurement of temperature. In works, I think it will be read until the covers fall off and the pages crumble, but that won't happen for many, many years, as the book is very substantially got up. It is a book for the thoughtful and likewise for the thoughtless, for it can help these latter, very materially, to grow up in science, and to apply it soundly in their everyday applications to temperature measurements, which, with flow measurements, are, too often, worse carried out than any determinations I came across in about 30 years' industrial experience.

The book comprises 4 sections, devoted to: (1) contact thermometers of all kinds, mercury in glass, vapour pressure, expanding metal, electric resistance, thermocouples, and a few others, (2) radiation pyrometers, (3) the Seger-cone type of instrument, including a series of metallic cylinders for use in the temperature range 130°–1100° C. in steps of 10°, and paints for use as temperature indicators between 100° and 400° C., when one indicating paint passes through the following striking series of colours (p. 140), white, 140°; yellow brown, 220°; grey, 290°; olive green, 325°; and dark olive, 405° C.; and (4) recorders.

There are two points to which I would draw particular attention in this book. First it is not encumbered with any theoretical matter. The only formulæ I have come across are those on p. 9 relating to the interconversion of Centigrade, Fahrenheit, and Réaumur temperatures. If we must have these formulæ in a book devoted to technical temperature measurements, well I don't think less space in any book has ever been devoted to the matter. Secondly, I want to commend to everyone concerned with the industrial measurement of the temperatures of streams of gases and vapour, the short section (pp. 96–99) devoted to this subject. Errors of about 50° C. or even much more are quite commonly made in industrial practice by failure to appreciate the principles here very clearly set out. A simple device can reduce a possible error of 25° C. to 1° C. or even less in the case of a stream of hot gas at 200° C. moving with a velocity of 5 metres/sec. It just isn't good enough to stick a thermometer into a stream of hot gas, *et voilà*, the temperature. Nor does one get the velocity of the stream by simply sticking a Pitot tube into the stream, *et encore voilà*, the velocity. And yet how very often have I seen these very measurements made in just such ways.

There is one other matter to which I would refer. St. Thomas Aquinas is not generally numbered amongst the physicists, but, nevertheless, his statement that " the more a thing is found to be self-sufficient, the better it is ", has many applications in physics, and instrument makers would be well-advised to bear his saying in mind. Now I consider that makers of resistance pyrometers have taken a retrograde step in departing from the original Callendar compensation method for eliminating leads resistance, and substituting therefor a system of compensation (see, e.g. p. 47) which is not, as Aquinas would say, self-sufficient, and therefore not for the best. The same criticism holds for attempts at eliminating lead errors in thermoelectric pyrometry, though for this again, a self-sufficient method was available some years ago but has now fallen into disuse.

I think I have said enough to indicate that this well-printed, well-illustrated, strongly bound book has my very heartiest commendation. All the sources of information given in the index are German, and the same is true of all the recorders described (they are fortunately not very numerous). An adequate subject index is provided.

J. S. G. Thomas.

Aluminiumguss in Sand und Kokille. Von Roland Irmann. Fünfte, neubearbeitete Auflage. 25 × 17 cm. Pp. xii + 302, with 318 illustrations. 1952. Düsseldorf: Aluminium-Zentrale e.V., Alleestrasse 31. (Brosch., DM 28.20; Ganzleinen, DM 30.60.)

This book appeared for the first time in 1935 as a short summary of aluminium alloy foundry practice. Since then the growth of the aluminium foundry industry has been so rapid that a complete revision of the original book became imperative. It is very fortunate that the author was a witness, as well as a contributor, to these modern developments and thus was in a position to approach, if not to attain, that very elusive target of writing an up-to-date book on foundry practice.

The book is clearly intended for those who are engaged in the aluminium foundry industry. The problems concerning the main stages of sand and gravity die-casting manufacture—raw materials, melting furnaces and practice, moulding in sand and designing metal moulds, fettling and finishing processes, mechanical testing, and casting design—are all well treated and, what is very helpful, illustrated with numerous examples. The metallurgical story which forms the background of these processes is also included, but it lacks somewhat in both completeness and freshness in comparison with the technological side of the book. It is mainly for this reason that a student of metallurgy will find this book less useful than will the foundryman, although he will find some general data that will guide him to look for the original references.

This book represents a notable contribution to foundry literature. It is to be hoped that the next edition will be even more concentrated on foundry practice, with the metallurgical sections forming the subject of a separate volume.

V. Kondic.

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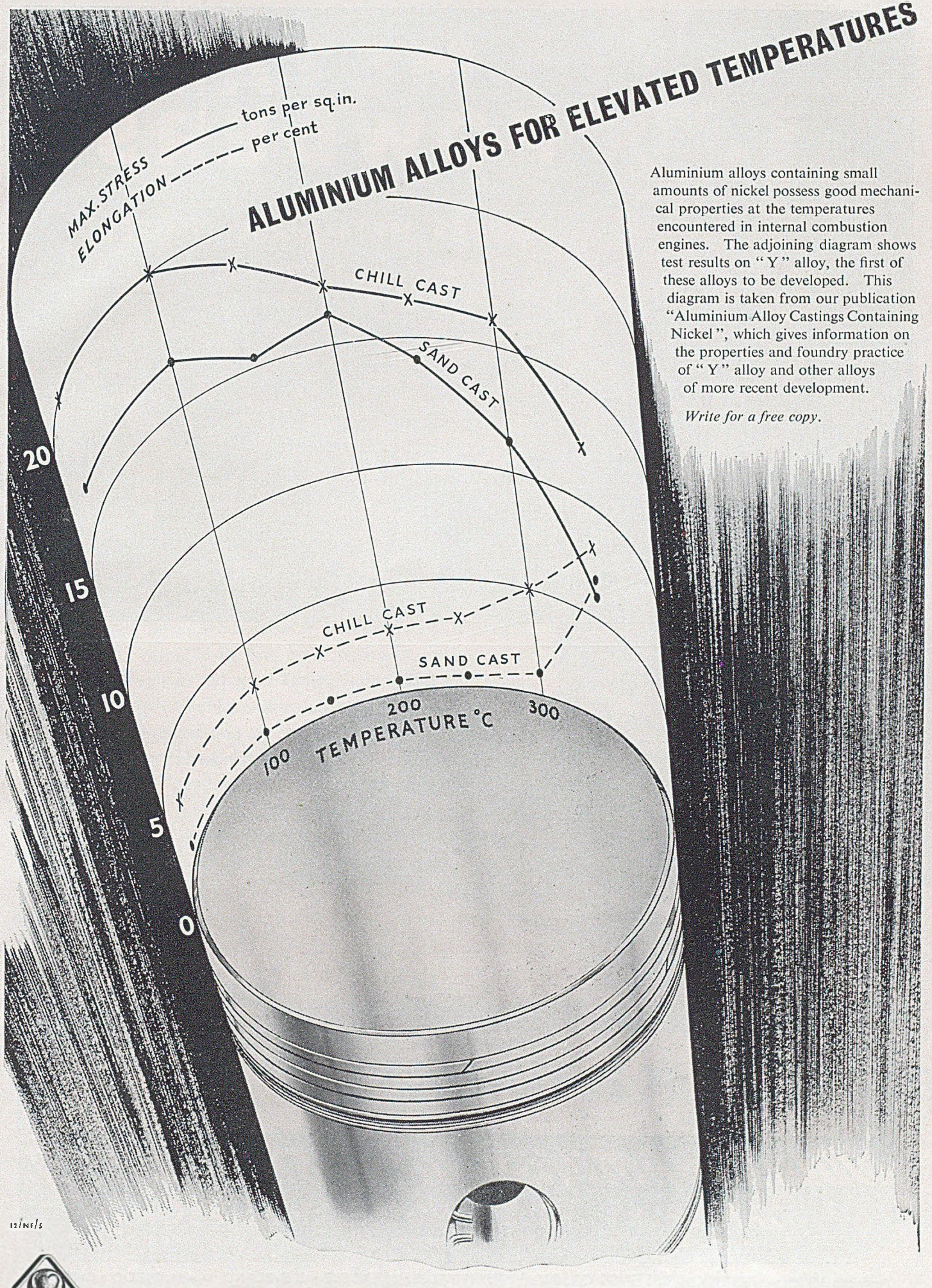

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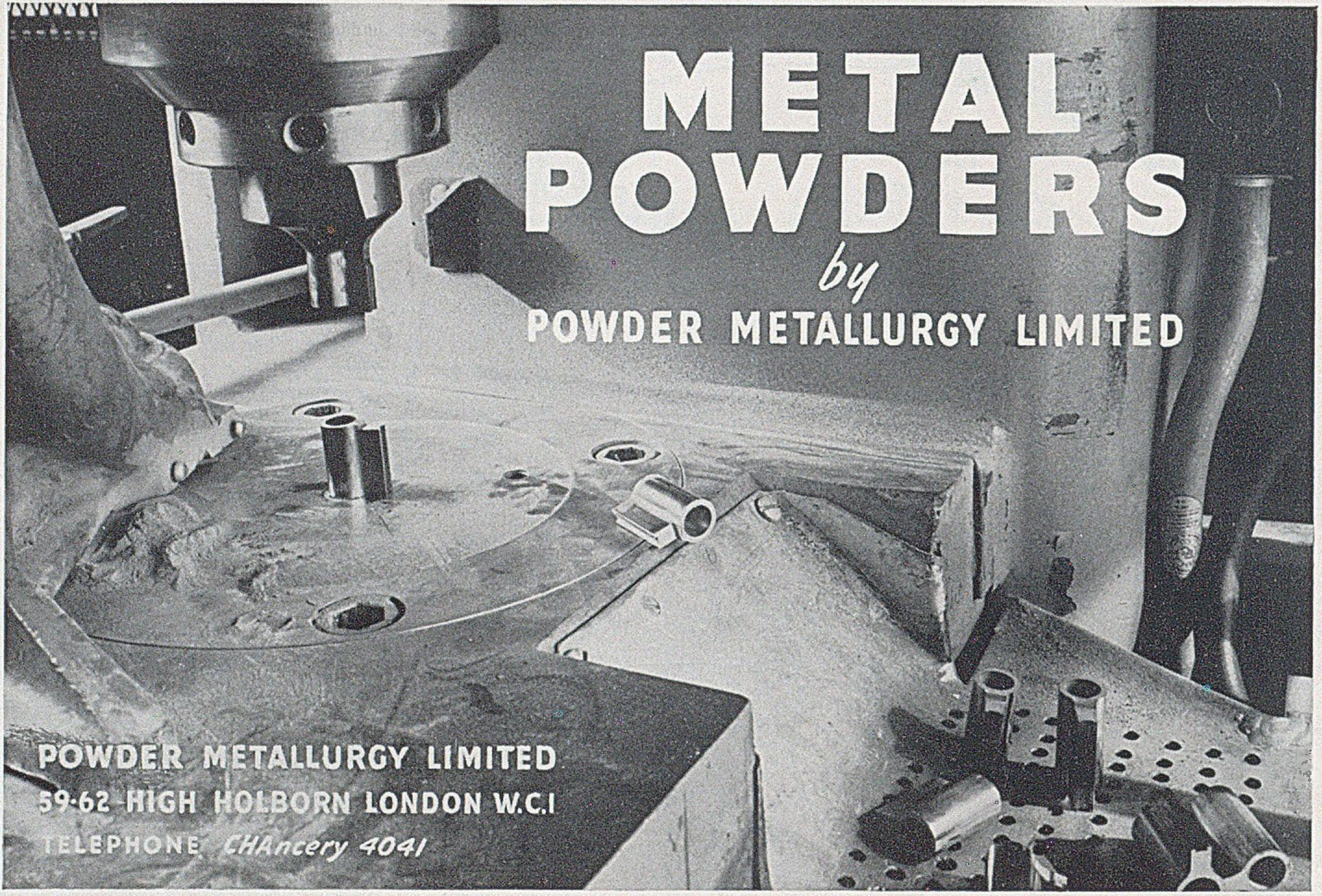

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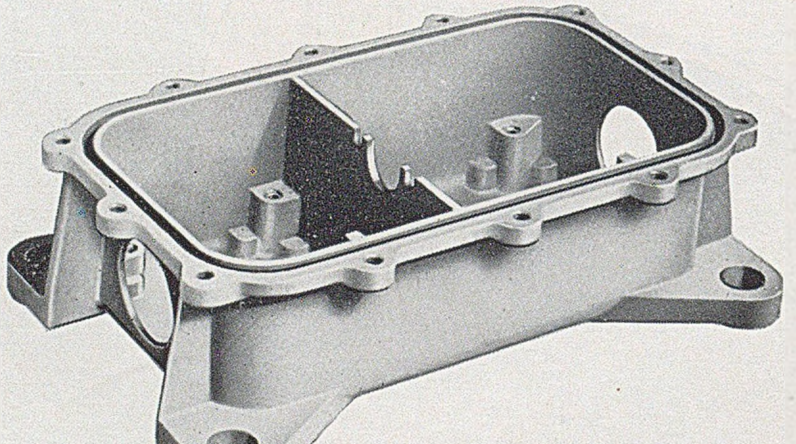

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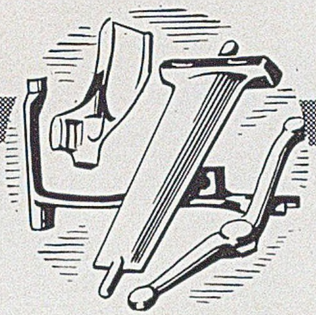

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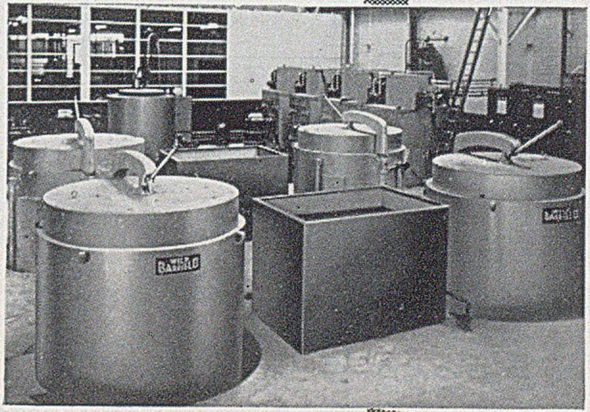

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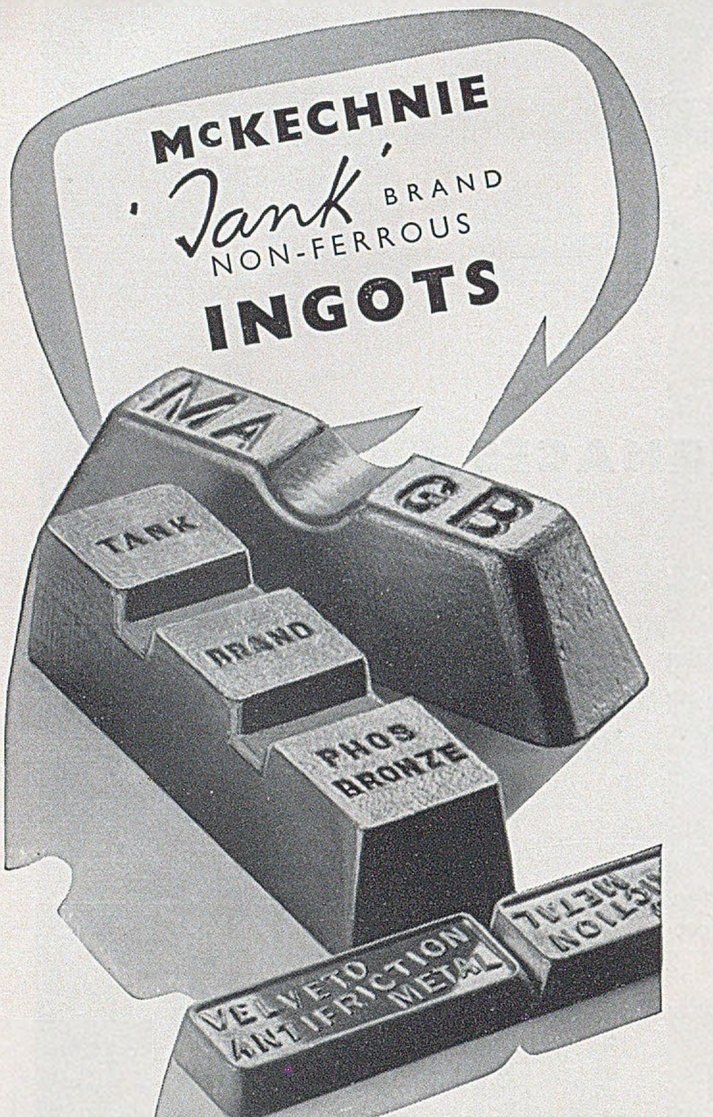

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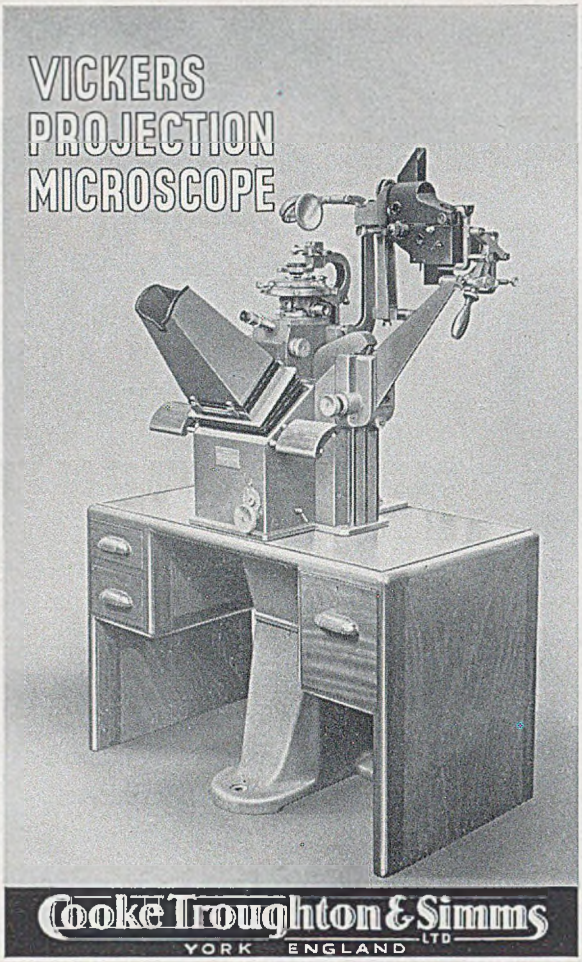

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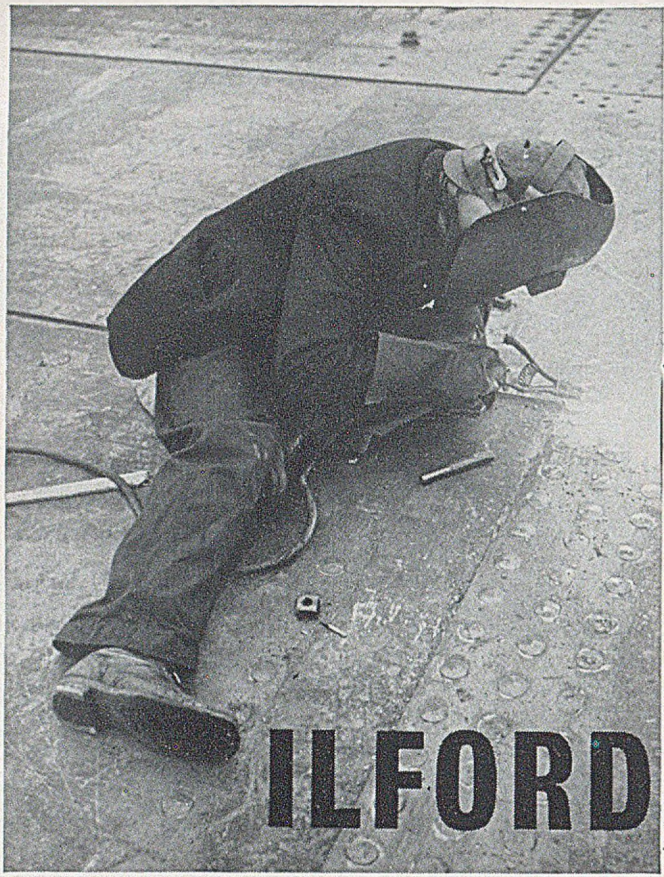

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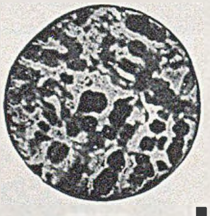

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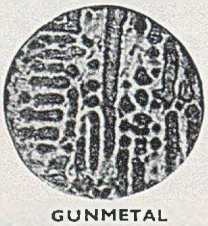

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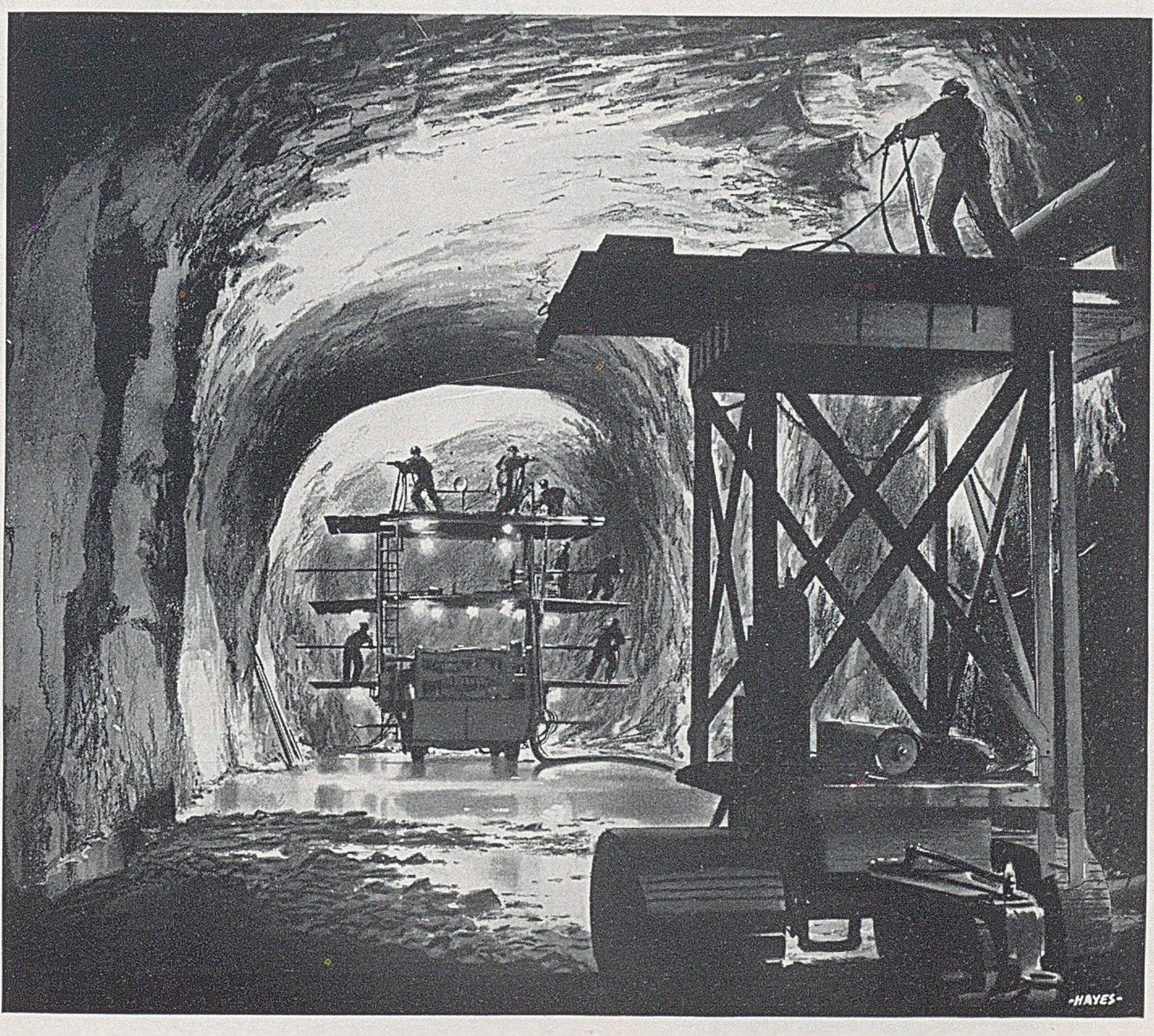

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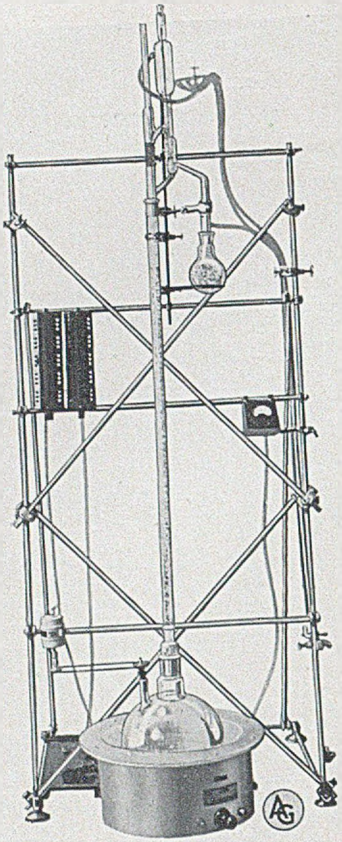

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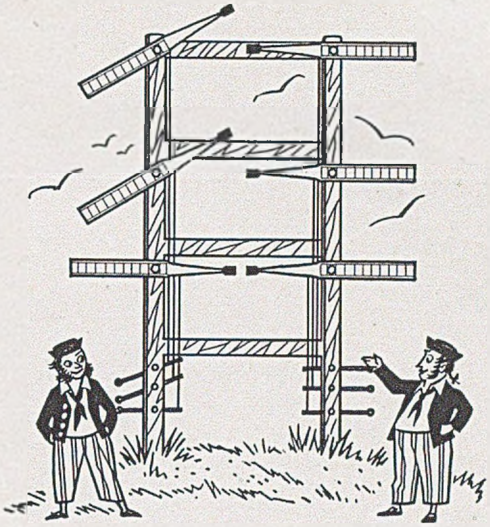

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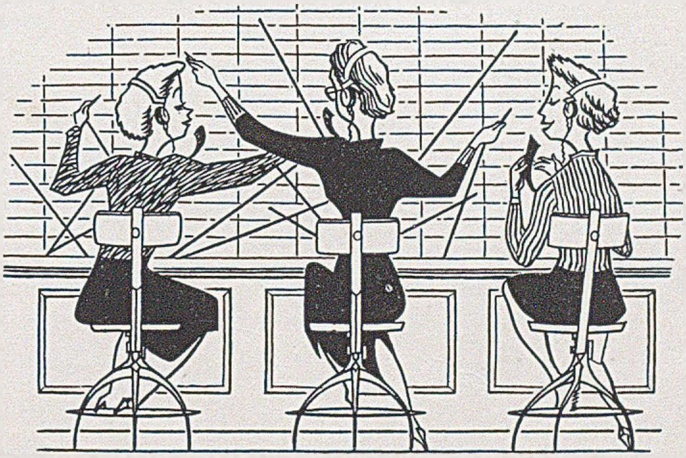

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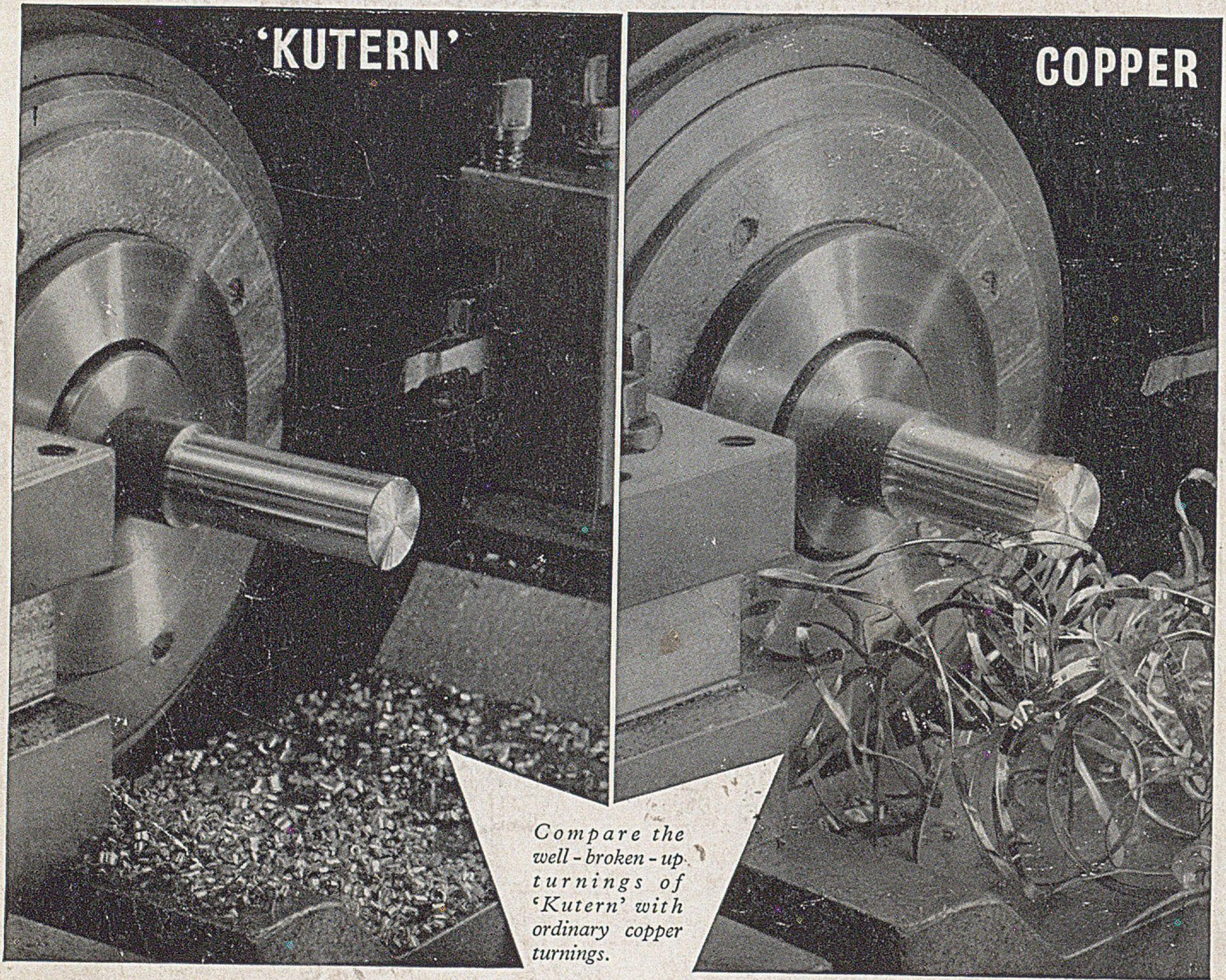

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