THE JOURNAL OF THE INSTITUTE OF Metals

WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents:

PRICE: TEN SHILLINGS AND SIXPENCE

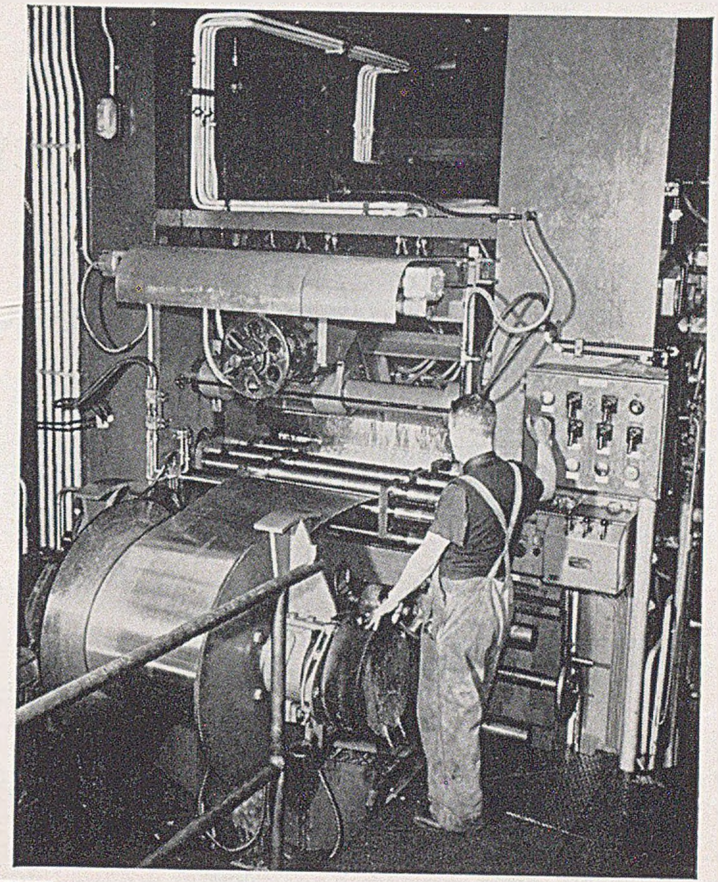
APRIL 1953

DAVY-UNITED — DEVELOPMENTS IN STRIP HANDLING EQUIPMENT

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

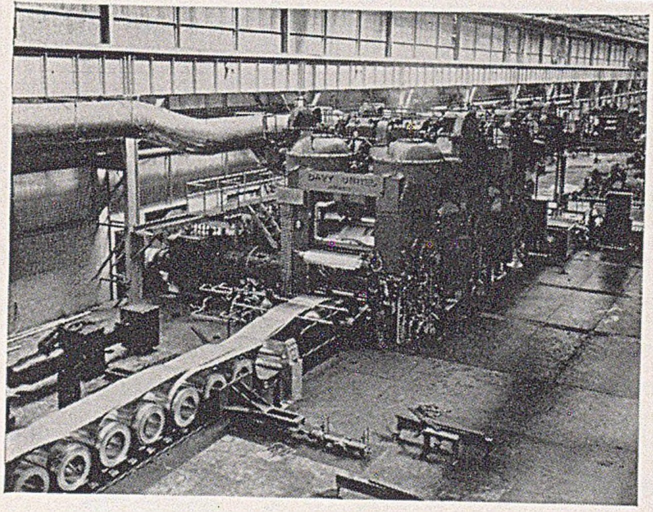
Combined Uncoiler and Sticker Roller Bridle

for aluminium strip



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

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



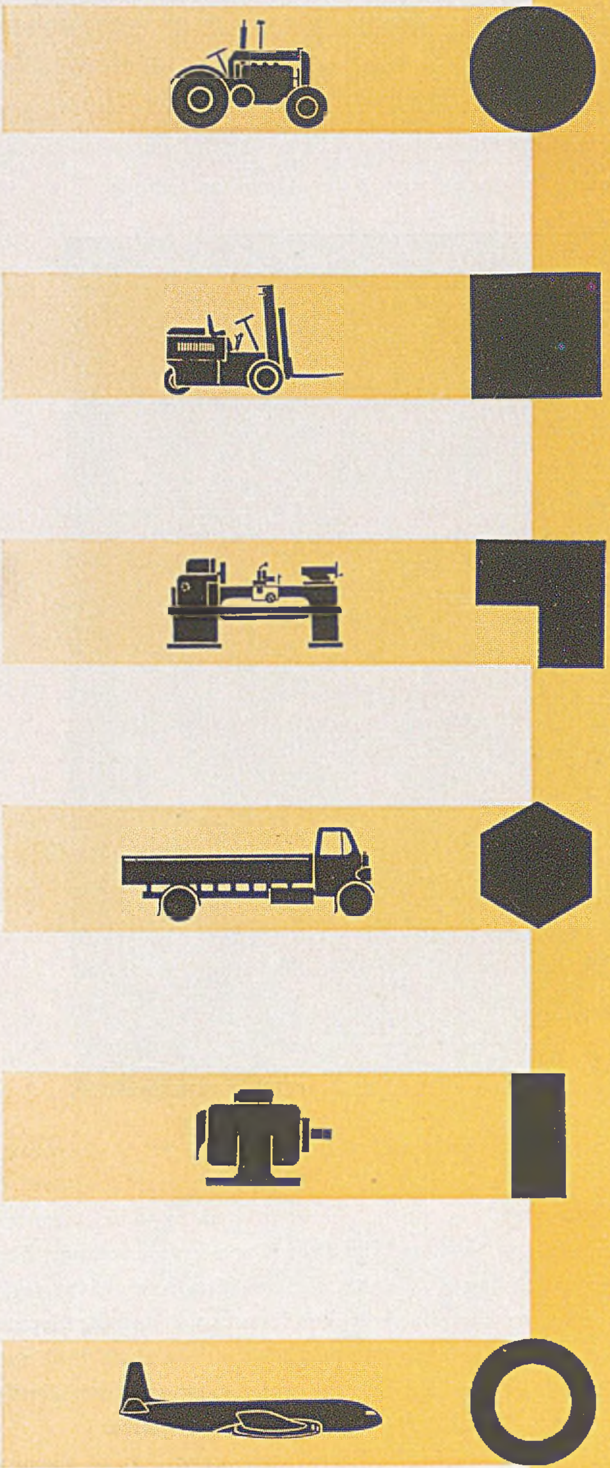
Proved in continuous operation on the world's fastest aluminium cold mill—2,000 f.p.m. finishing speed—this combined unit is designed to handle 5,000 lb. coils of strip up to 56 in. wide, the heaviest coils yet envisaged in British non-ferrous rolling. The opening and closing of the cones, the clamping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

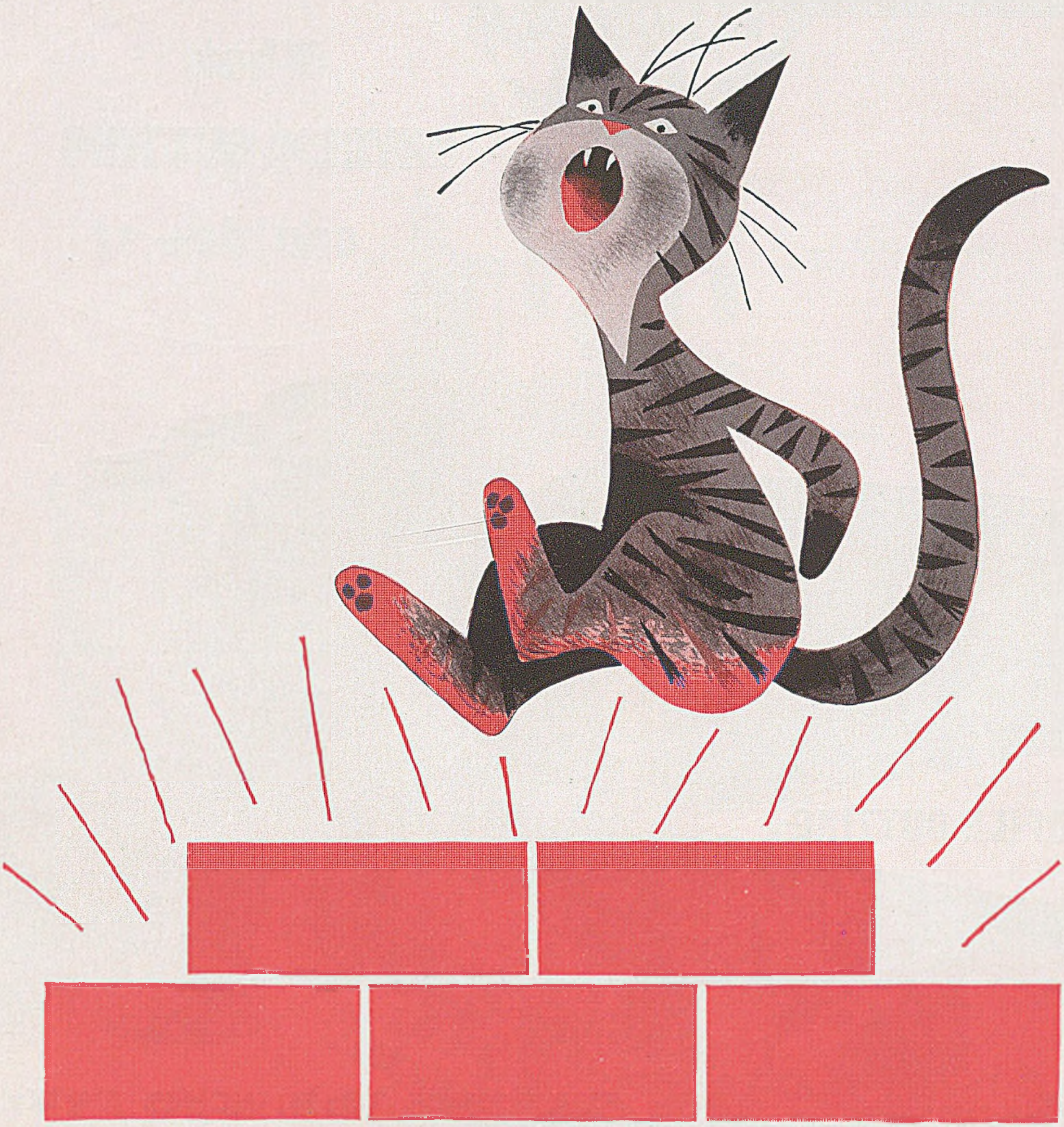
Apart from quickening and simplifying the motions of accurate coil entry the combined unit is also designed to impart a measure of back tension to the strip during rolling.

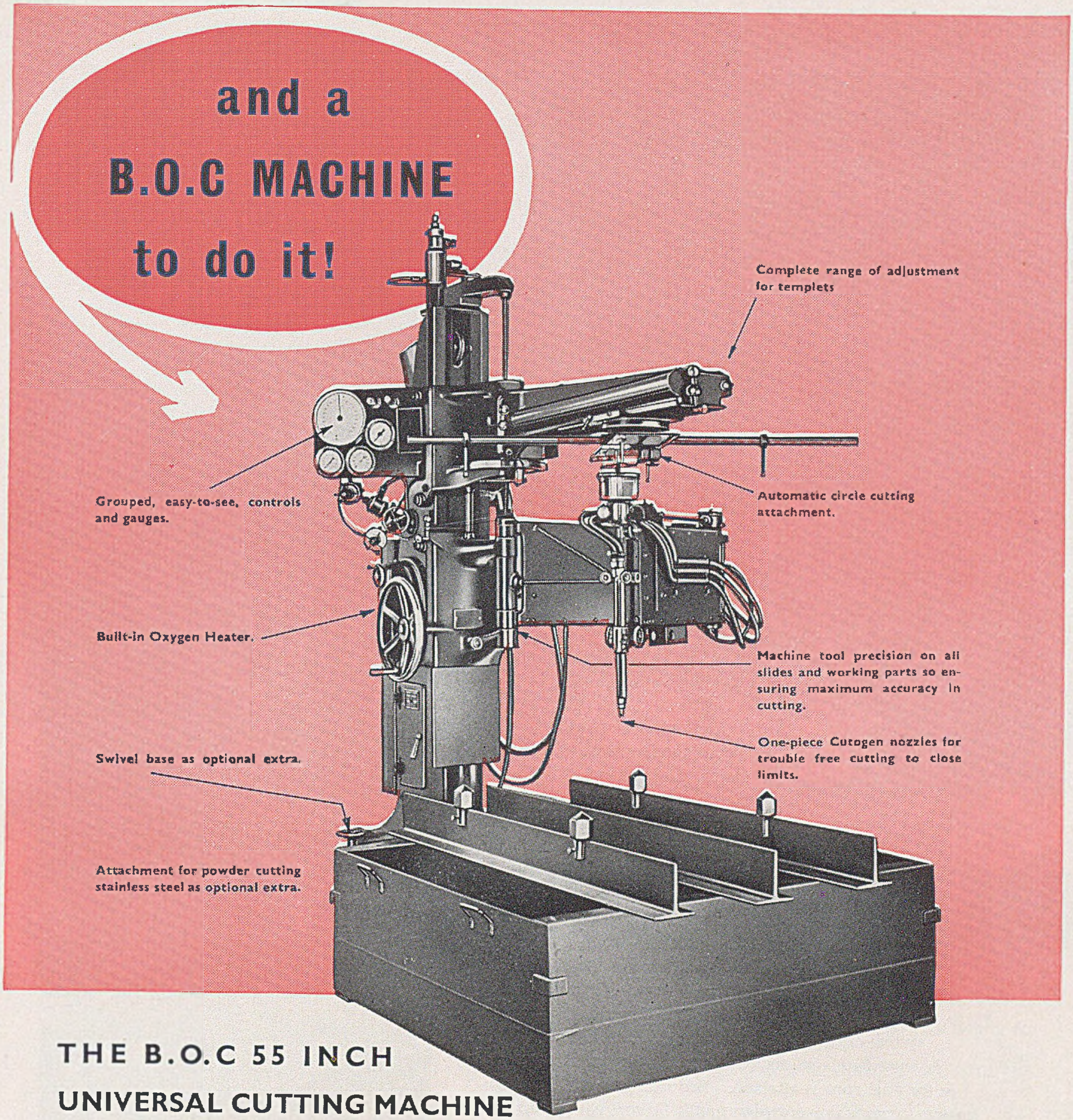
Progressive engineering at the service of the Metals Industries.

DAVY-UNITED

SHEFFIELD · MIDDLESBROUGH · GLASGOW







THE B.O.C 55 INCH
UNIVERSAL CUTTING MACHINE

cuts any shape within 55″ radius including circles up to 55″ diameter, rectangles up to 16 sq. ft. surface area and straight cuts up to 9 ft. 3 ins. long. The automatic circle cutting attachment makes circular cuts up to the maximum diameter of 55″ without the use of templets.
The 55″ machine is fitted with the M.C.12 Cutter of 12″ capacity using the Cutogen One-piece nozzles as standard but other cutters are available up to a capacity of 24″.

The machine can also be supplied with a swivel base so that four cutting tables may be employed.

For profiling stainless steel a special powder cutting attachment is available.
Write for full particulars to your B.O.C. Branch

THE B.O.C 36 INCH
UNIVERSAL CUTTING MACHINE

Within its capacity of 36″ cutting radius and maximum cutting depth of 6 ins. the B.O.C. 36″ U.C.M. is identical in performance with the 55 inch model.

A Powder cutting attachment is available for profiling stainless steel.

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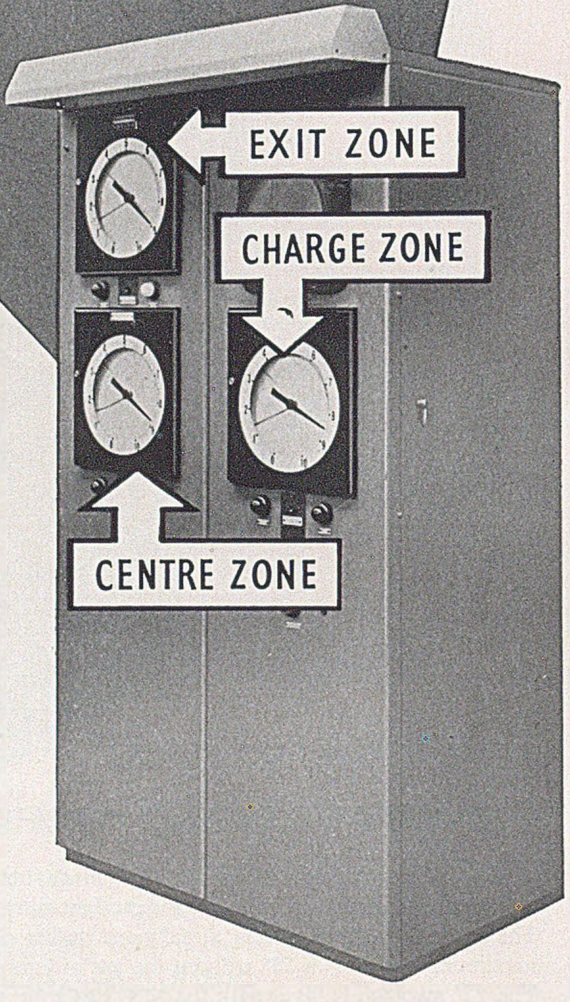
THE BRITISH OXYGEN CO LTD
LONDON & BRANCHES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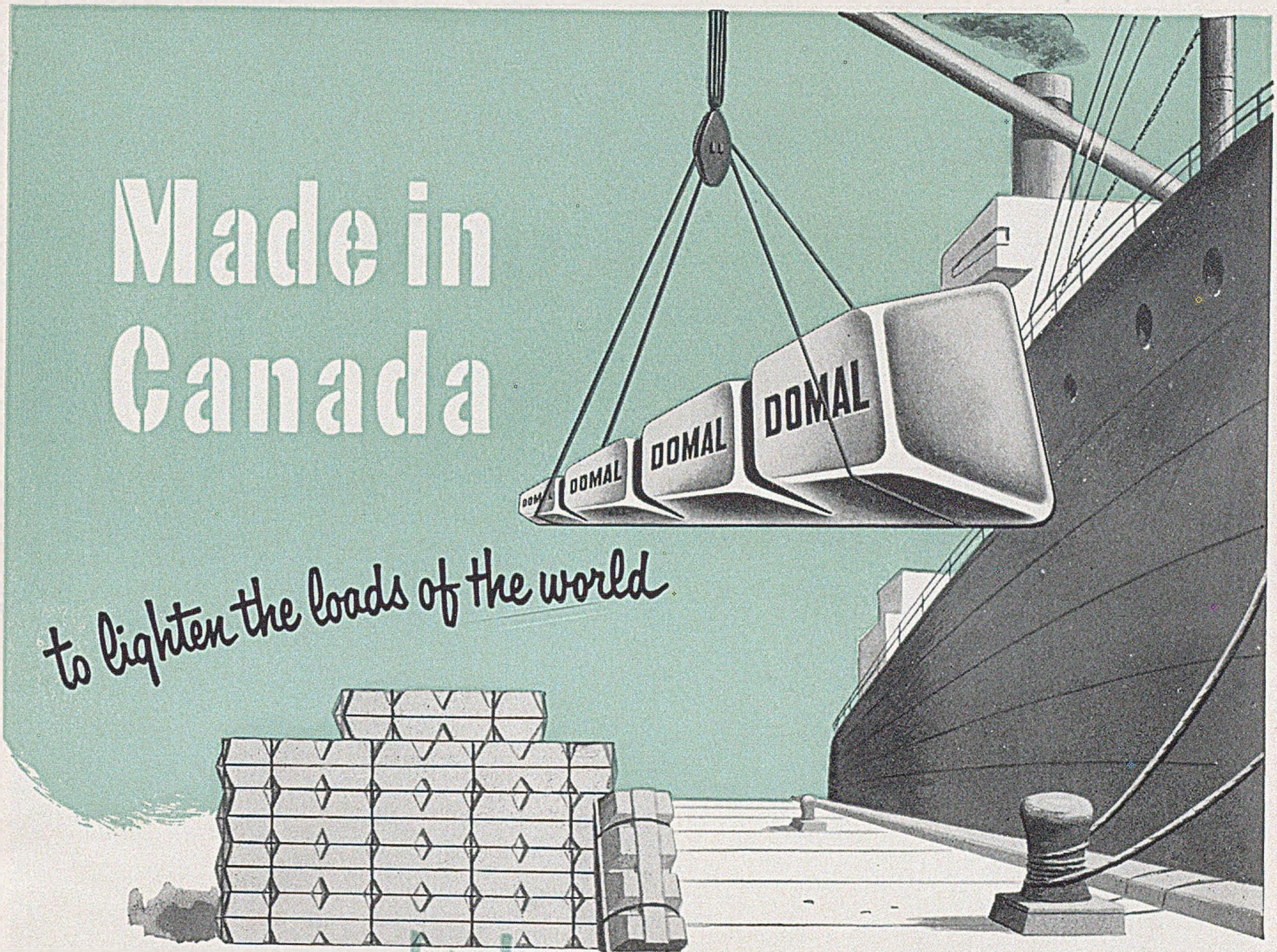
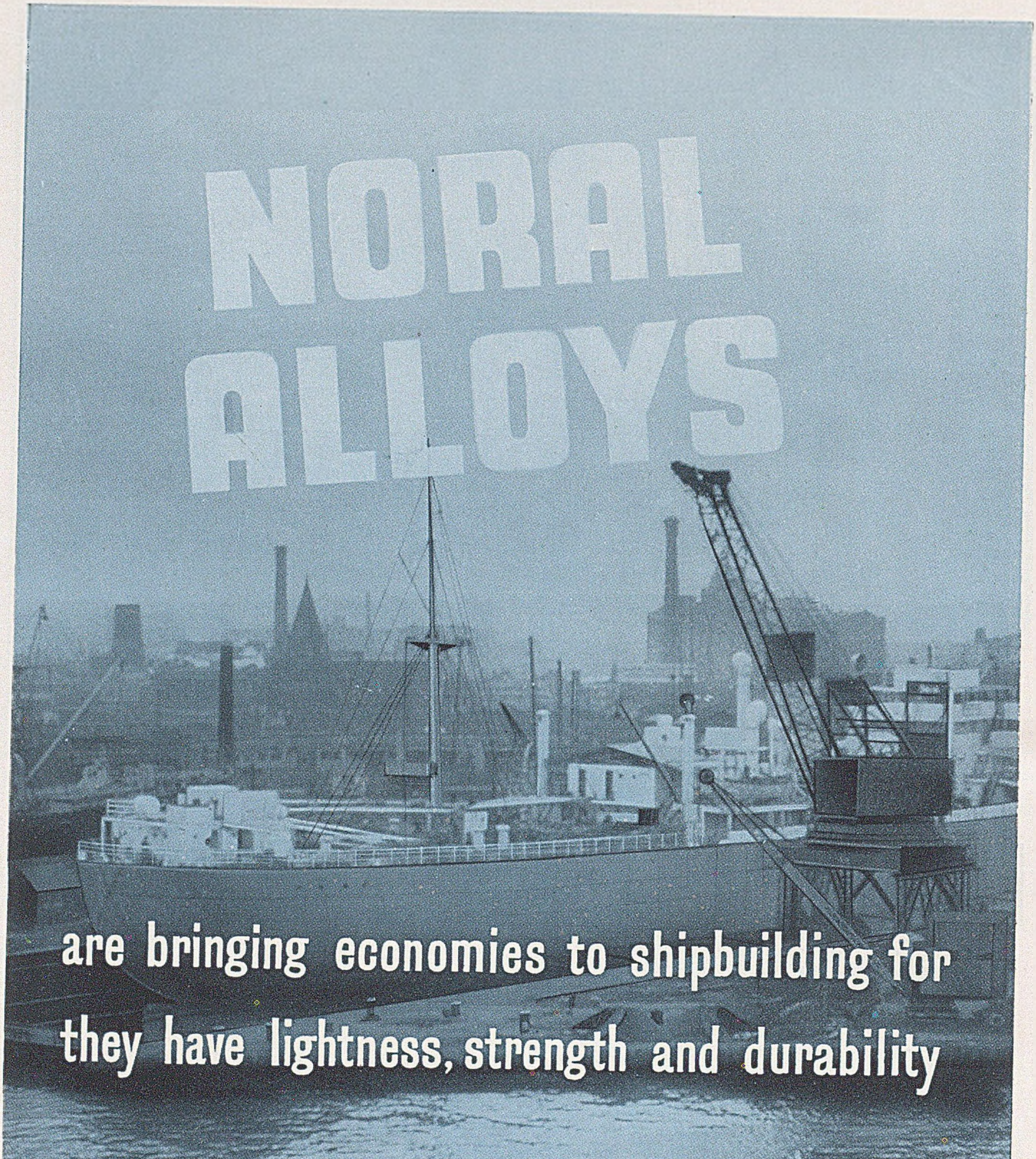

Made in Canada
to lighten the loads of the world
DOMAL



NORAL ALLOYS
are bringing economies to shipbuilding for they have lightness, strength and durability
Northern Aluminium COMPANY LIMITED
ALUMINIUM LIMITED
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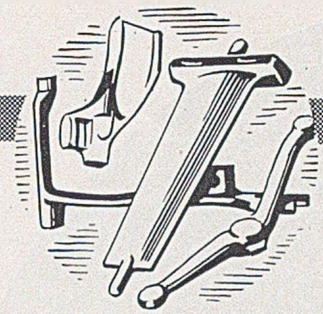

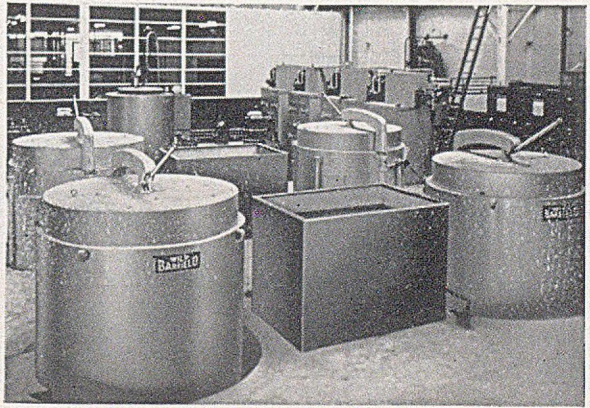

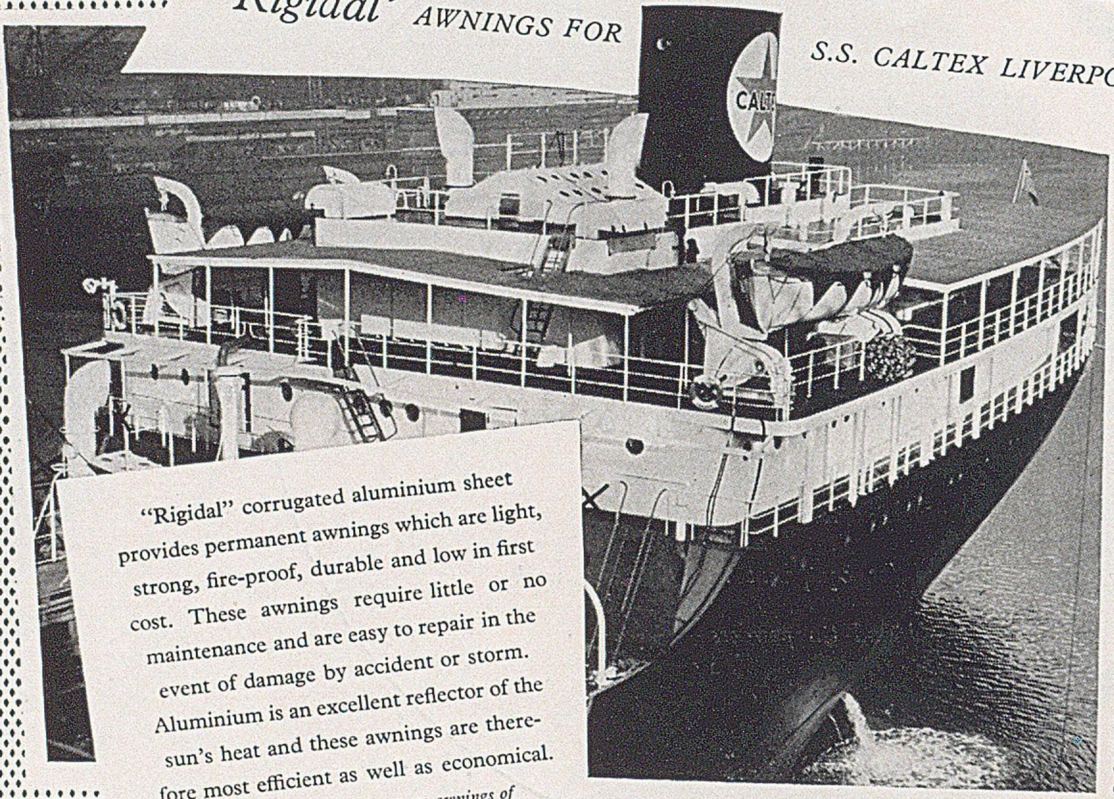
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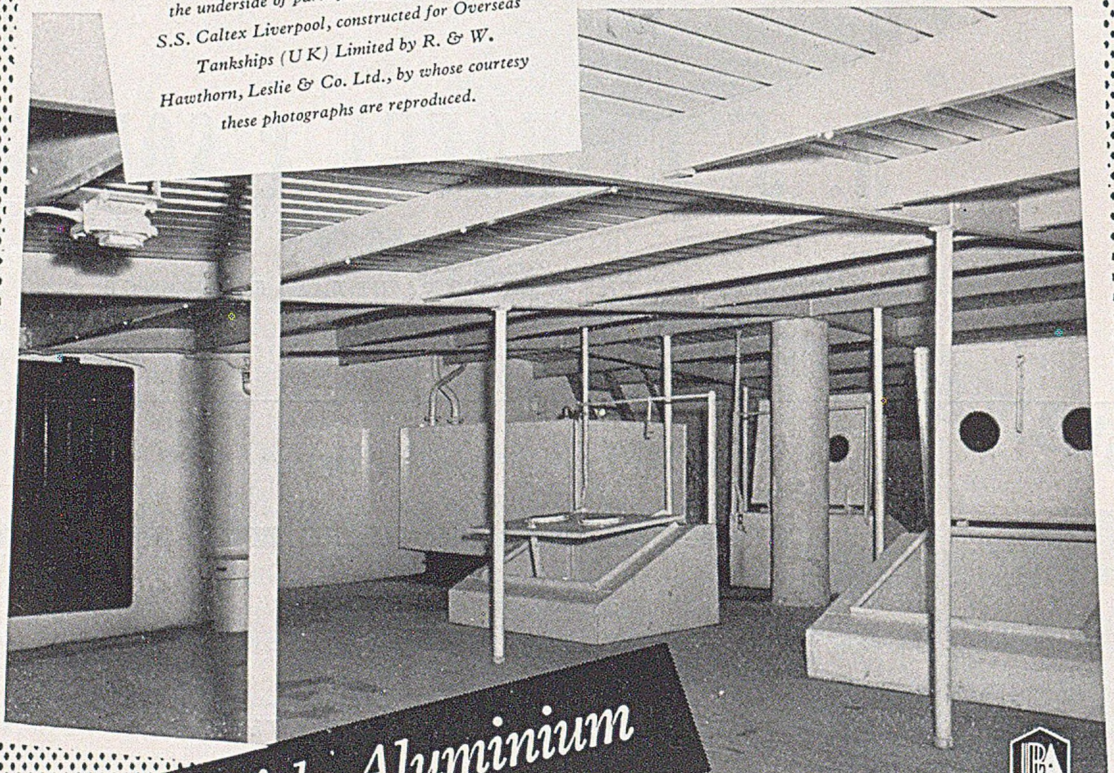
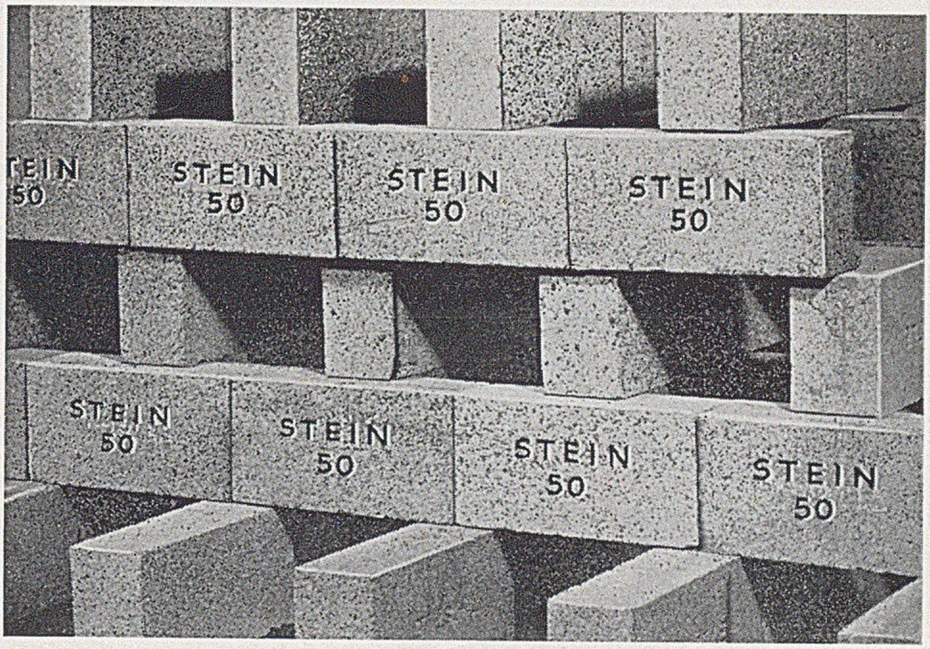


STEIN *Refractories*



STEIN 50

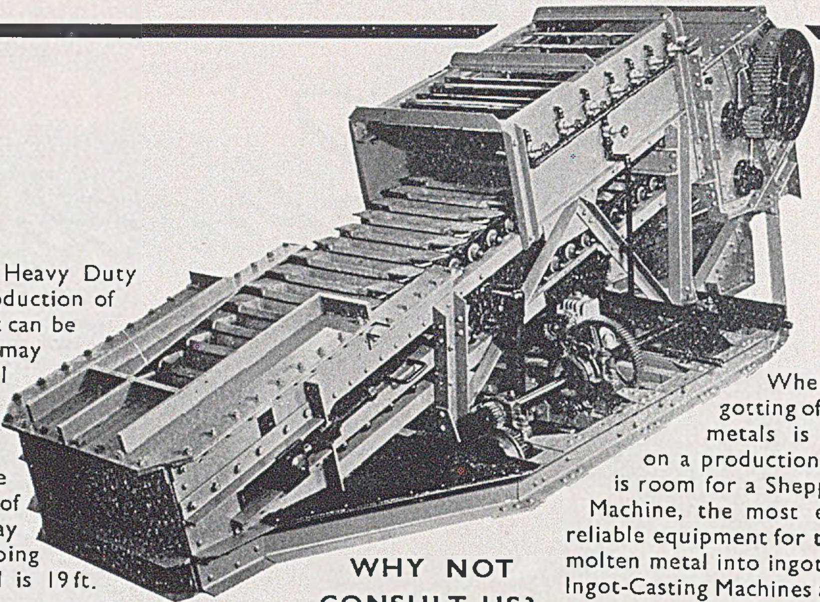
APPROXIMATE TECHNICAL DATA

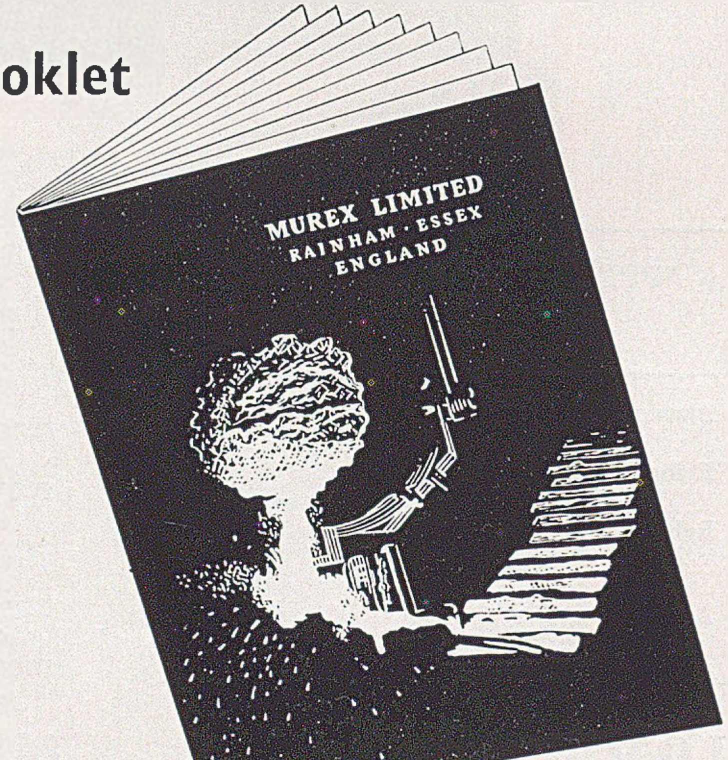
Alumina	51-52%
Refractoriness	Cone 35/36 1780°C
Refractoriness-under-load 28 lbs./in^2	5% Deformation at 1595°C
After-Contraction 2 hrs. at 1500°C	Nil

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

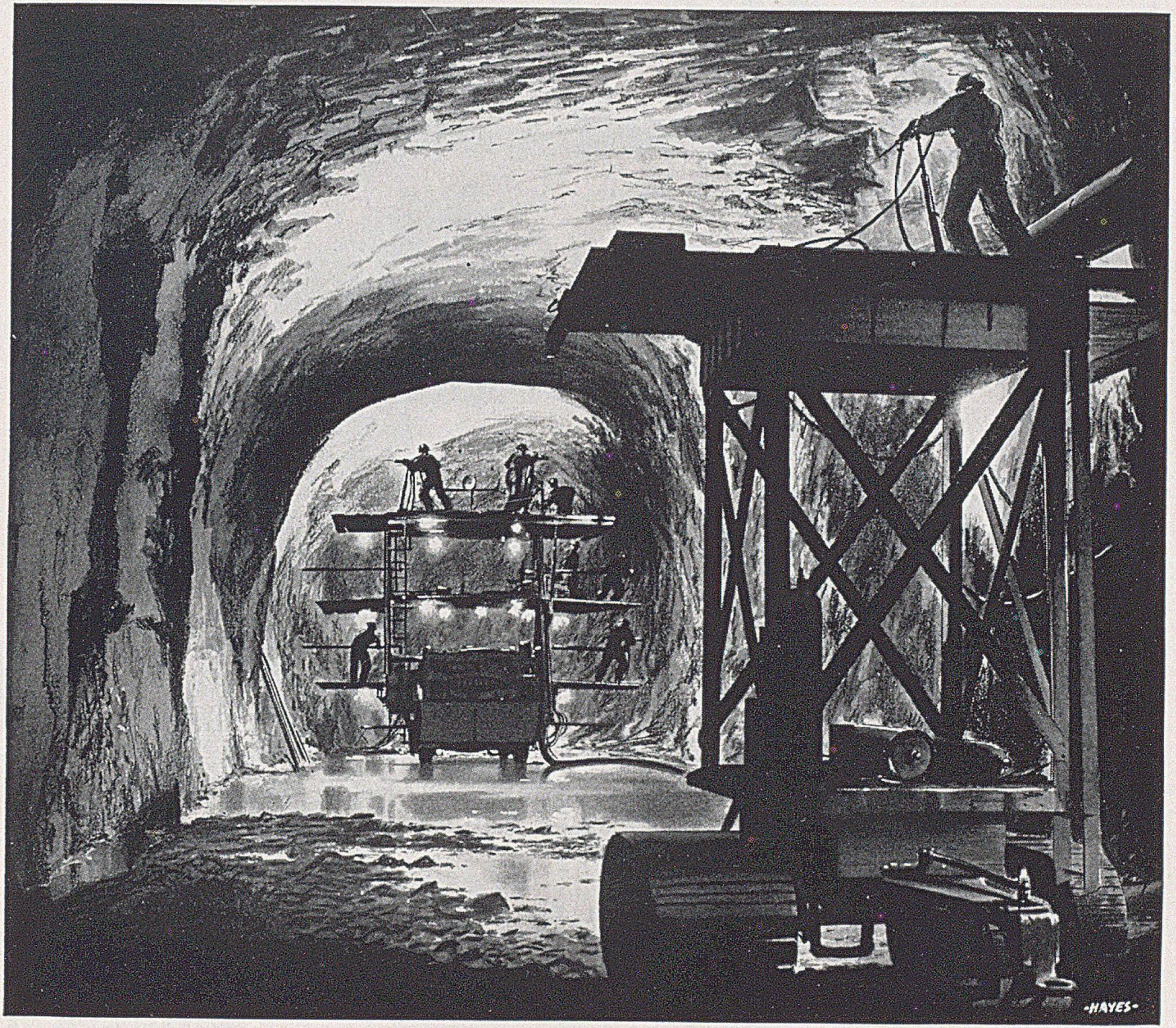
JOHN G. STEIN & C^{o} L^{TD} Bonnybridge. Scotland

TEL: BANKNOCK 255 (3 LINES)




MUREX LIMITED
RAINHAM · ESSEX
ENGLAND

aluminium adventure



tunnellers in a MOUNTAIN

Ten miles beyond this tunnel entrance and half a mile down, roaring waters will be tamed and harnessed. A finger on a switch will set great dynamos turning — dynamos that will generate power capable of providing the free countries of the world with nearly half a million tons of aluminium a year when fully developed.

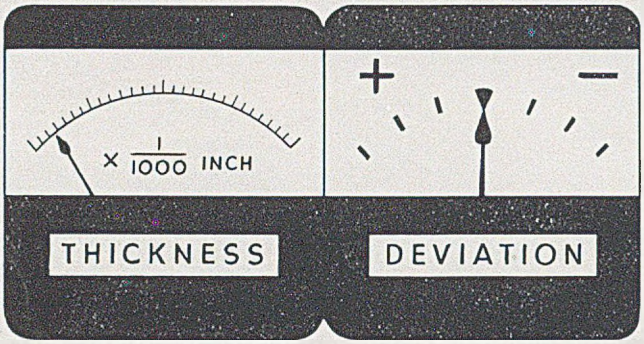
The scene is British Columbia, where the Aluminum Company of Canada Ltd. (an Aluminium Limited Company) is engaged on a great industrial expansion in the face of rugged opposition — a barrier of mountains.

Already the course of a chain of lakes is being reversed by damming and the drainage area converted into a huge reservoir. Water from this reservoir will have but one outlet — a tunnel gouged out of the mountains and falling steeply to what will be the largest underground power station in the world. All this to one end — the growth of large-scale production and distribution of aluminium and its alloys, from mine to market.

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

Aluminium Union Limited

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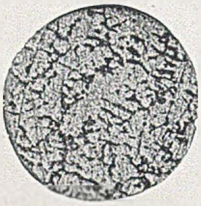

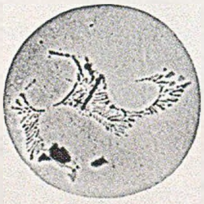

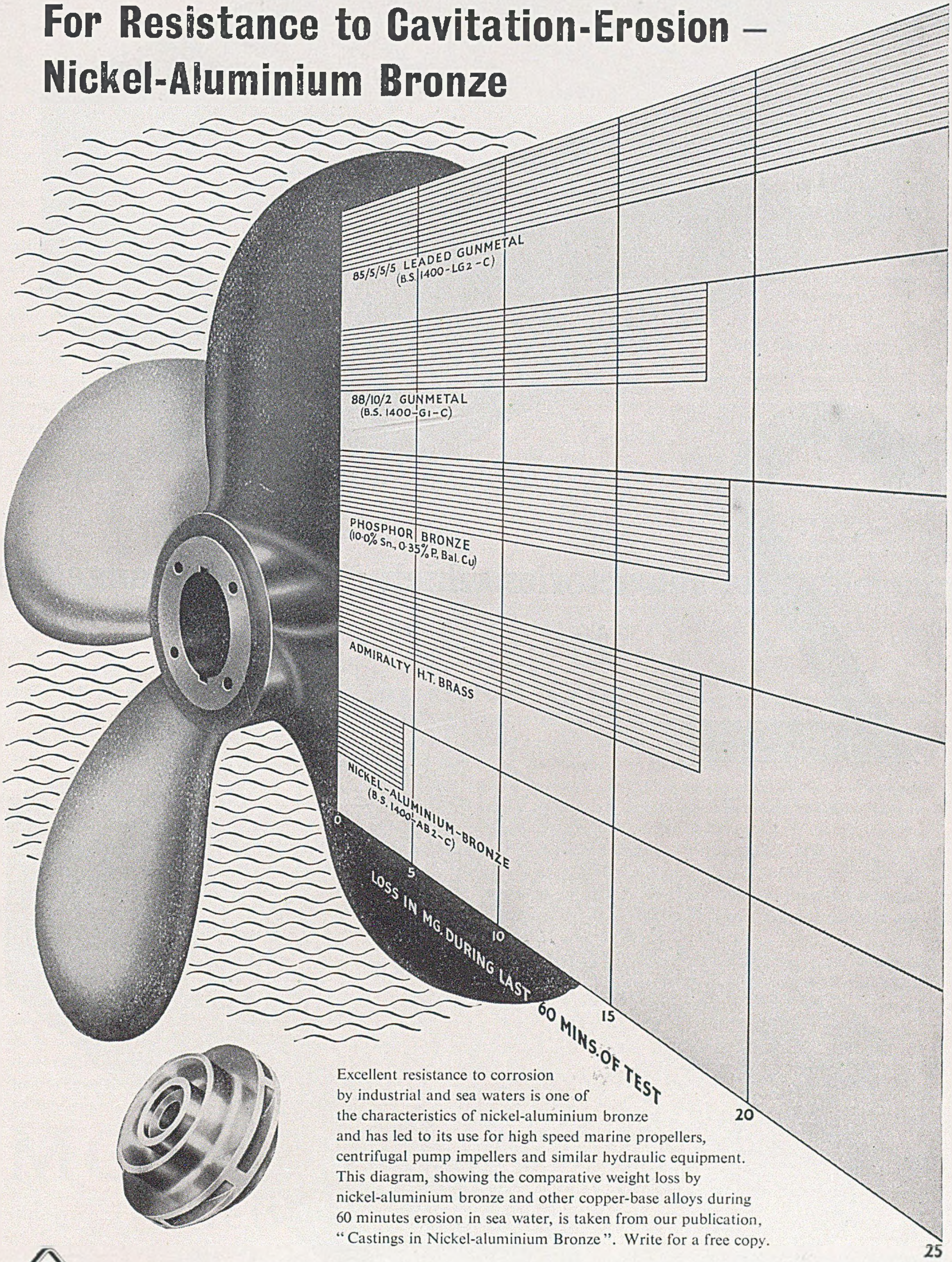
85/5/5/5 LEADED GUNMETAL
(B.S. 1400-LG2-C)
88/10/2 GUNMETAL
(B.S. 1400-G1-C)
PHOSPHOR BRONZE
(10·0% Sn., 0·35% P., Bal. Cu)
ADMIRALTY H.T. BRASS
NICKEL-ALUMINIUM-BRONZE
(B.S. 1400-AB2-C)
0
5
10
15
20
25
LOSS IN MG. DURING LAST 60 MINS. OF TEST

2

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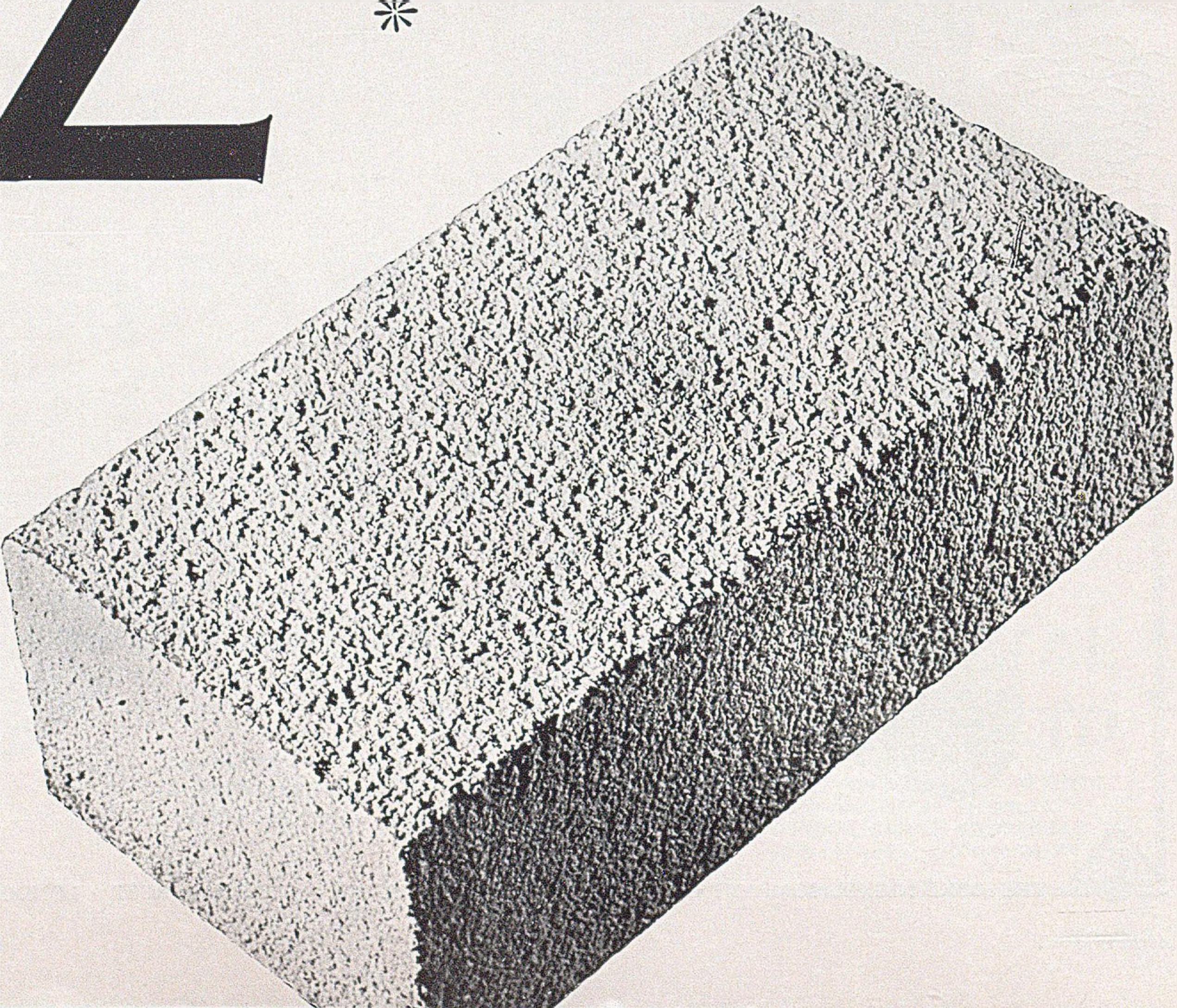
refractories which may well change the

*



FURNACE MAINTENANCE

* THE MORGAN M.R.1

—a brick that carries the ordinary high quality firebrick into entirely new fields of usefulness. It can be used, for example, at temperatures as high as 1600°C—far beyond the capacity of other refractories of similar alumina content. In fact, with these bricks, this standard of comparison is no longer valid. They can be judged only on performance and in that they are comparable only with special purpose refractories having a very high alumina content indeed.

How is it done? The answer is in the way they are made: in the selection and purification of the clay; in the unusually hard burning and careful grading of the grog; above all in the very high temperature of the final firing. The manufacturing process is a continuous one—which in itself makes for uniformity—and it is carried out under rigorous quality control.

All this costs money—but consider what its results mean in practice. Remarkable rigidity up to 1600°C; an after-contraction at this temperature of only 1% (The usual temperature for measuring after-contraction is 1410°C); negligible after-contraction below 1600°C. The bricks are close textured and strong, their resistance to abrasion is high; and they have unusually good resistance to both thermal and physical spalling.

Bricks of this type, although not previously manufactured in this country or in Europe, have been in use for some years in the U.S.A. where they have decisively proved their economy in terms of reduced furnace maintenance.

TYPICAL PROPERTIES OF M.R.1

Approximate Chemical Analysis

Silica	(SiO_2)	52/53%
Alumina	(Al_2O_3)	43/44%
Iron Oxide	(Fe_2O_3)	less than 1%
Titanium Oxide	(TiO_2)	less than 1%
Magnesia	(MgO)	less than 2%
Lime	(CaO)	
Potash	(K_2O)	
Soda	(Na_2O)	

Physical Characteristics

Refractoriness ...	Cone 35 (1770°C)
Refractoriness under load (25 lb./sq.in.):	
Commencement of subsidence	1600°C:
10% subsidence	1700°C
Bulk density ...	132 — 137 lb./cu.ft.
After-contraction (2 hrs. 1600°C)	less than 1.0%
Thermal expansion ...	4.5 x 10-6 per °C.

whole conception of furnace maintenance and efficiency

* THE MORGAN LOW STORAGE REFRACTORY M.I.28

—a brick that can double furnace output. It is a hot-face insulating refractory which can be used at furnace (or interface) temperatures up to 2800°F (1538°C).

At these temperatures it has a lower conductivity than any other type of refractory; and therefore provides a greater reduction in the convection and radiation losses from the outside of the furnace. But that is less than half the story: even more important is the reduction in heat storage. The M.I.28 is little more than a third of the weight of an ordinary refractory; consequently only a third of the heat is required to raise the whole of it to the same temperature. But, with the same furnace temperature, the average temperature of an M.I.28 is much lower (owing to its lower conductivity) and this still further reduces the heat it takes up. With the same heat input, therefore, furnaces built from M.I.28 bricks heat up rapidly: on batch furnaces this can amount to a doubling of the furnace output—to say nothing of the saving in fuel. Low conductivity also has a bearing on the effective strength of the brick. For, even when working at high face temperatures, the greater part of the brick is relatively cool. Not only this but their light weight reduces the load on the lower courses and, incidentally, the weight of the whole furnace structure.

There have been hot-face refractories before. What, then, is new about the M.I.28? In theory nothing. Again it is a matter of the way they are made. Morgans have put the whole of the theory into practice. The plant is entirely new in design and operates under rigorous quality control from the purification of the clay to the final grinding to size.

As in the case of the M.R.1, bricks of this quality have been available for some years in the U.S.A. and the improvements they can make in furnace efficiency have been firmly established.

TYPICAL PROPERTIES OF M.I.28

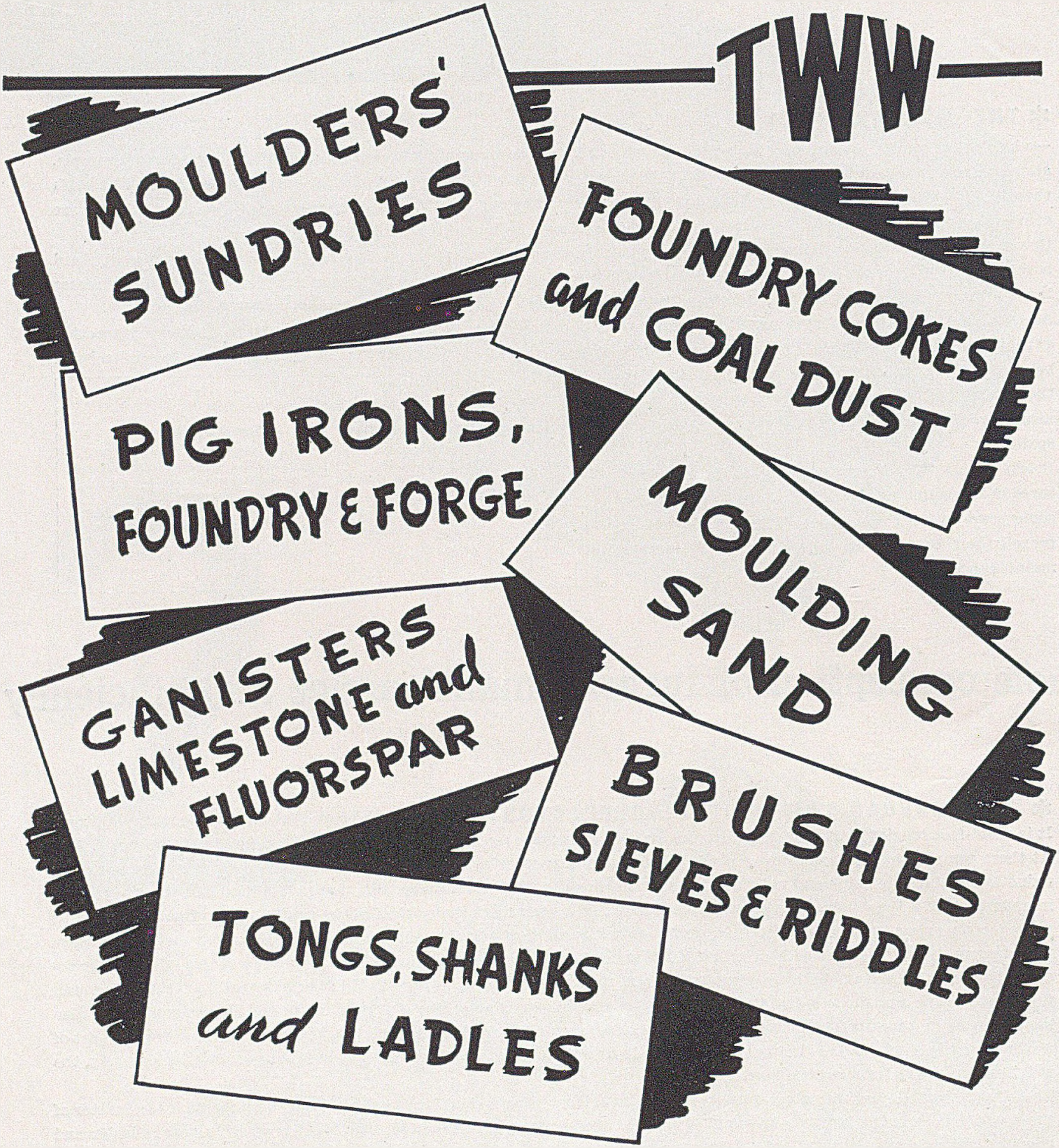
Maximum Service Temperature		... 1538°C (2800°F)
Thermal Conductivity:		
Mean Temperature	538°C (1000°F)	... 2.4 B.Th.U hr.(sq.ft.)(in.)(°F)
	816°C (1500°F)	... 2.9 B.Th.U hr.(sq.ft.)(in.)(°F)
Bulk Density		... 47.5 lb. cubic ft.
Refractoriness		... 1710°C (3110°F)
Modulus of Rupture		... greater than 120 lb./sq.in.
Heat Capacity Factor		... 0.105

(the ratio of the heat stored in a M.I.28 furnace wall relative to that stored in a firebrick wall of the same area, and of a thickness giving similar hot and cold face temperatures)

MORGAN Refractories ARE WORTH FAR MORE THAN THEY COST

THE MORGAN CRUCIBLE COMPANY LTD.
Battersea Church Road, London, S.W.11. Tel: Battersea 8822

FURNACE EFFICIENCY


TWW
MOULDERS' SUNDRIES
FOUNDRY COKES and COAL DUST
PIG IRONS, FOUNDRY & FORGE
MOULDING SAND
GANISTERS LIMESTONE and FLUORSPAR
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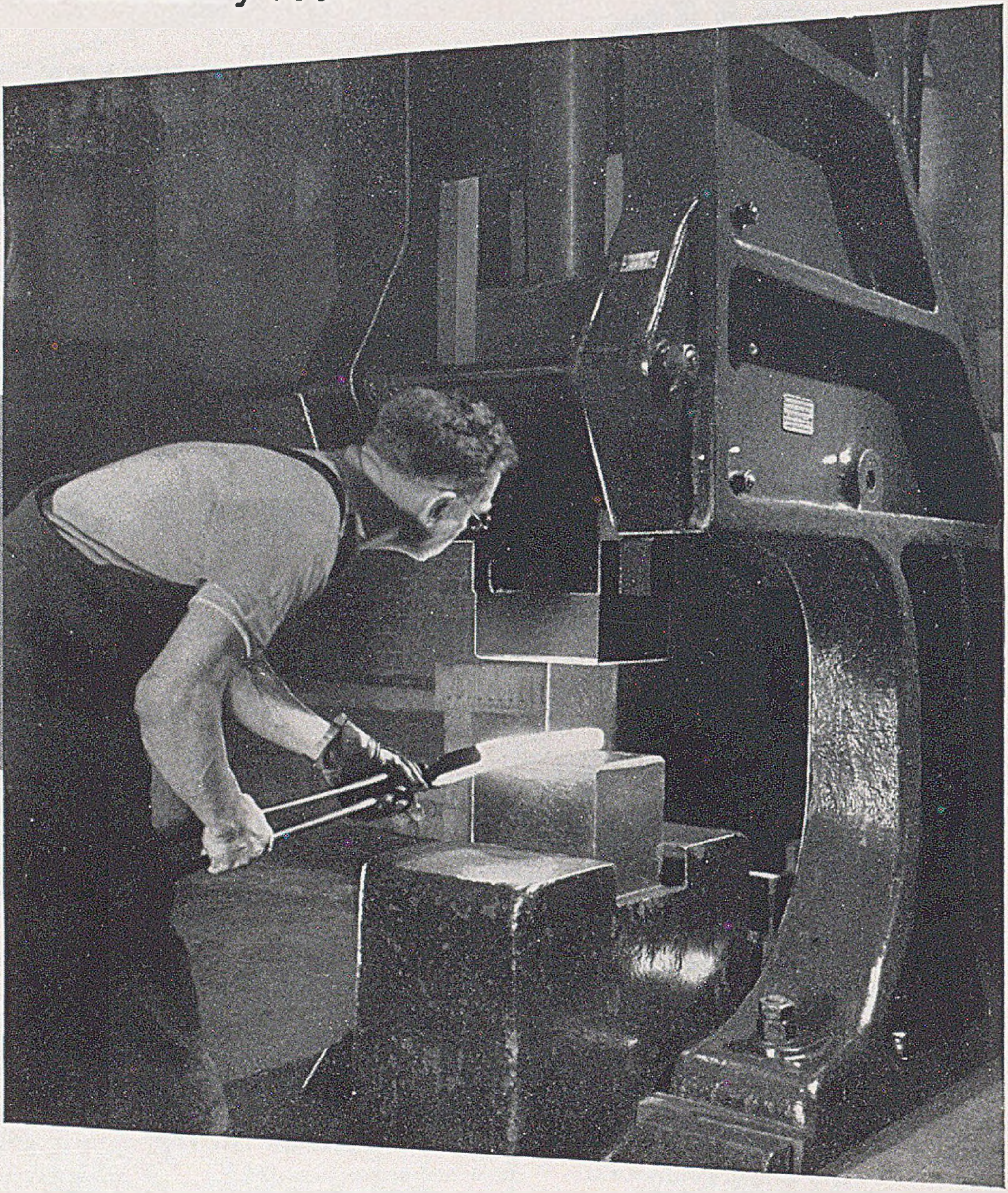

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METALLURGICAL ABSTRACTS

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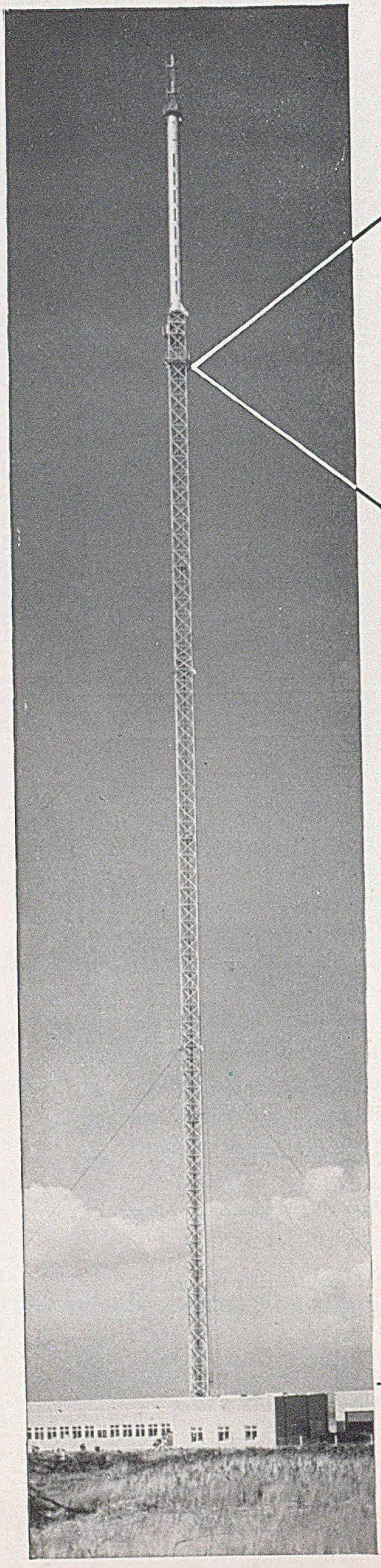

BULLETIN
OF THE INSTITUTE OF METALS

VOLUME I APRIL 1953 PART 20

INSTITUTE NEWS

W. H. A. Robertson Medal and Premium

The Council has awarded the W. H. A. Robertson Medal for 1952, and a Premium of Fifty Guineas, to Mr. John Francis Waight for a paper on "Gas Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys", published in the *Journal*, 1951–52, Vol. 80, pp. 269–285.

1954 Autumn Meeting in Switzerland

The Council has very gratefully accepted an invitation from the Swiss Association of Machinery Manufacturers and the Swiss Association for Testing Materials to hold the 1954 Autumn Meeting in Switzerland.

An announcement regarding dates will be published in due course.

Rolls and Their Maintenance in the Non-Ferrous Metals Industry

It is hoped to publish in the May issue of the *Bulletin* a summary of the Informal Discussion on this subject, which was held in Birmingham on 8 January 1953.

Informal Discussion on "Lubricants for Metal-Working Operations in the Non-Ferrous Metal Industries"

An all-day Informal Discussion on "Lubricants for Metal-Working Operations in the Non-Ferrous Metal Industries", organized by the Metallurgical Engineering Committee, will be held in Birmingham on Wednesday, 6 January 1954.

Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys

"Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys" (No. 14 in the Institute's Monograph and Report Series) has now been published. It contains seven papers presented at a Symposium held during the Annual Spring Meeting last year, together with an abridged account of the discussion on them. The published price is 15*s.* ($2.50), post free. Each member may obtain one copy at the privileged price of 7*s.* 6*d.* ($1.50), post free.

Reduced Rate of Subscription for Members Retired from Active Professional Life

Members who have retired from their main professional occupation, are not less than 65 years of age, and have been members of the Institute for not less than 25 years, may claim the privilege of paying a reduced rate of annual subscription on application to the Secretary.

Publication of Short Notes in the Journal

Members are reminded that the Publication Committee is now willing to consider original communications of a length not exceeding 800 words for publication as one-page notes in the *Journal*. Such notes will be judged by the same standards as those applied to ordinary papers, but, if acceptable, they will be given a high priority in publication and will appear with a minimum of delay. An example of such a note will be found in the present issue (p. 432).

Students' Essay Prize Competition, 1952

The Council of the Institute of Metals has made two awards, of Ten Guineas each, for essays submitted in connection with the Institute's annual Students' Essay Prize Competition. The awards are to Mr. R. D. Stacey of the University of Birmingham, for an essay on "Some Experimental Evidence for Dislocations" and Mr. G. Thomas, B.Sc., of Cambridge University, for an essay on "Martensitic Transformations in Non-Ferrous Metals and Alloys". Both are Student Members of the Institute.

Chairmen of Committees for 1953–54

The Council has appointed the following Chairmen of Committees for 1953–54:

Finance and General Purposes Committee.—Mr. Alfred Baer.
Local Sections Committee.—Dr. L. B. Pfeil.
Medal Committee.—The President.
Metal Physics Committee.—Professor G. V. Raynor.
Metallurgical Engineering Committee.—Mr. W. J. Thomas.
Nominations Committee.—The President.
Publication Committee.—Mr. Christopher Smith.

1953 Autumn Lecture

Dr. Maurice Cook, Joint Managing Director, Metals Division, Imperial Chemical Industries, Ltd., has accepted an invitation from the Council to deliver the 1953 Autumn Lecture on the occasion of the 45th Autumn Meeting of the Institute, which will be held at Southport, Lancs., from Monday to Friday, 21–25 September.

The W. H. A. Robertson Medal and Premium

The W. H. A. Robertson Bronze Medal and a Premium of Fifty Guineas are awarded annually by the Council to the author or authors of the paper adjudged to be of the highest merit contributed to the *Journal of the Institute of Metals* on engineering aspects of non-ferrous metallurgy. No award is made if, in the Council's opinion, no paper published during the year is of sufficient merit to justify the award.

The award is open to persons of any nationality and of either sex, and to non-members as well as to members of the Institute. MSS. of papers should be addressed to the Secretary of the Institute in the normal way. All published papers are considered for the award annually by the Medal Committee, with the advice of the Metallurgical Engineering Committee.

The medal and award are placed at the disposal of the Council by the Directors of W. H. A. Robertson and Co., Ltd., of Bedford, who, having in mind the few papers contributed to the *Journal* dealing with the engineering side of non-ferrous metal production, and believing that such papers would be to the advantage of the Institute and the industry, offered the sum of £100 a year, for a period of not less than seven years, to stimulate the writing of papers of this kind.

Fig. A.—THE W. H. A. ROBERTSON MEDAL (Full-size).
By arrangement with the Deputy Master of the Royal Mint. Artist: Mr. Percy Metcalfe, C.V.O.

The objects which the donors had in mind are warmly approved by the Council, which hopes that the awards will stimulate the submission of such papers for publication by the Institute. The Metallurgical Engineering Committee will be grateful to members of the Institute if they will encourage suitable members of their staffs to submit papers of a high standard for the consideration of the Publication Committee.

The medal, which was designed by Mr. Percy Metcalfe, C.V.O., is reproduced—full size—in Fig. A.

The figure on the medal represents Hephaistos, the god of fire and smiths, who was called Vulcanus by the Romans. According to Homer he was the son of Zeus (Jupiter) and of Hera (Juno). Later traditions (Hesiod) state, however, that he had no father and that Hera gave birth to him independently of Zeus, of whom she was jealous. Homer says he was born lame and weak and that he dwelt with the marine divinities Thetis and Eurynome for 9 years in a grotto beneath Oceanus, where he fashioned a number of exquisite works of art, among them a golden throne with invisible chains. Later he was recalled to Olympus, where he became the great artist of the gods.

The sojourn of Hephaistos in the cavern under the sea and his fondness for the island of Lemnos is, in all probability, based on volcanic phenomena, the submarine activity of volcanic fires, and the natural features of the island of Lemnos itself. Here there were annual festivals in his honour. In one, all fires were put out for nine days, during which rites of atonement and purification were performed. Then fresh fuel was brought on a sacred ship from Delos, the fires were rekindled, and a new life began. In October the smiths and smelters celebrated the Chalkeia, a feast of metal workers; at the Apatūrĭa sacrifices were offered to him, among other gods as the giver of fire; torches were lit and hymns sung, and at the Hephaistĭa, finally, there was a torch race in his honour.

Hephaistos appears to have been originally simply the god of fire, but as fire is indispensable in working metals he was afterwards regarded as an artist. He made palaces of brass for himself and the other deities. His palace in Olympus was imperishable. It contained his workshop with an anvil and 20 bellows, which worked spontaneously at his bidding "blowing upon the melting pots, sending forth deft blasts of varying strength now to further his labours and now again, however Hephaistos might will it, according as his work went on. And on the fire he threw tough bronze and tin and precious gold and silver." He made the armour of Achilles, the ægis of Zeus, the sceptre of Agamemnōn, the fatal necklace of Harmonia, the fire-eating bulls of Aeëtes, King of Colchis. He had a forge also on Mount Mosychlos in Lemnos, possibly one under Etna in Sicily and on the sacred island of Hephaistos in the Lipari Islands where the Cyclopes were his workmen and servants. To help his lameness he made, according to

Homer, two golden maidens with the power of motion, on whom he leaned when he walked.

The lettering on the medal was the subject of considerable discussion, because the name was spelt very differently at different periods in Greek history and in different parts of the Greek world. Ultimately, on the advice given by Dr. L. R. Palmer, Professor of Greek at King's College, London, it was decided to adopt the spelling which appears on the inside of a basin of the 5th century B.C. (see Fig. B), which is illustrated in Roscher's " Lexikon der Mythologie ", p. 2058.

Fig. B.

Election of Members

The following 20 Ordinary Members, 7 Junior Members, and 17 Student Members were elected on 13 February 1953:

As Ordinary Members

Barclay, Donal James, General Manager, Hudson and Wright, Ltd., Halberton Street, Birmingham 18.

Baxter, Laurence Cecil, General Manager, The Hockley Chemical Co., Ltd., Hockley Hill, Birmingham 18.

Bourne, Bernard Stather, Partner, Newall and Bourne, Consulting Engineers, Post Office Chambers, 26 Brown Street, Manchester 2.

Cleaver, Frederick Thomas, Wire Mill Superintendent, British Insulated Callender's Cables, Ltd., Prescot, Lancs.

Desirant, M. C., Ph.D., Directeur du Laboratoire des Recherches Physiques, Ateliers de Constructions Electriques de Charleroi, Charleroi, Belgium.

Evans, Sydney Joseph, Works Engineer, D. F. Tayler and Co., Ltd., Newhall Works, George Street, Parade, Birmingham 3.

Gerritsen, Alexander Nicolaas, D.Sc., Scientist, Stichting Fundamenteel Onderzoek op het gebied van de Materie (F.O.M.), Werkgroep Metalen, Bÿlhouwerstraat 6, Utrecht; and Honorary Scientific Officer, Kamerlingh Onnes Laboratorium, Rijksuniversiteit te Leiden, Nieuwsteeg 18, Leiden, Netherlands.

Kirby, Noel Spencer, B.E., General Superintendent, Electrolytic Zinc Company of Australasia, Ltd., Box 634B, G.P.O., Hobart, Tasmania, Australia.

Saulnier, Adrien, Chef de Section Micrographie et Rayons X, Service Recherches et Essais Physiques, Compagnie Péchiney, 1 Avenue Lorraine, Chambéry (Savoie), France.

Simons, Arthur John, A.C.A., Managing Director, Arthur E. Heckford, Ltd., Frederick Street and Regent Street, Birmingham 1.

Somville, Charles A., Managing Director, Société Anonyme Laminoirs de l'Escaut, Burcht-Anvers, Belgium.

Spence, Leonard Rolls Bowness, Works Manager, British Insulated Callender's Cables, Ltd., Prescot, Lancs.

Stevens, John Hugh, M.A., Project Manager, Canada Wire and Cable Co., Ltd., Toronto 17, Ont., Canada.

de la Vallée-Poussin, Raoul, Ing. Civil des Mines, Ing. Civil Elect., Directeur de la Section du Hainaut du Centre National de Recherches Métallurgiques, Université du Travail, Boulevard Solvay, Charleroi, Belgium.

Vazquez Lopez, Jerónimo, Metallurgical Engineer, Nueva Montaña Quijano S.A., Paseo Pareda 32, Santander, Spain.

Verchiani, *Commendatore* Augusto, Ing., Ispettore Tecnico, Società Metallurgica Italiana, Borgo Pinti 99, Firenze, Italy.

Wilman, Harry, Ph.D., D.Sc., F.Inst.P., Lecturer, Laboratory of Applied Physical Chemistry, Imperial College of Science and Technology, London, S.W.7.

Wilson, Graham Rudge, B.Sc., A.M.I.E.E., Chief Engineer, Metal Industries Division, English Electric Co., Ltd., Stafford.

Winning, Herbert, B.Sc., D.R.T.C., A.R.I.C., Technical Manager, James Mitchell and Son (Greenock), Ltd., 7 West Stewart Street, Greenock.

Yates, *Major* Leonard James, Director, B. Halstead and Co., Ltd., Praed Road, Trafford Park, Manchester 17.

As Junior Members

Beynon, David Joshua, Deputy Chief Chemist, Richard Thomas and Baldwins, Ltd., Wern Aluminium Works, Briton Ferry, Neath, Glam.

Crawford, Henderson Black, B.Sc., A.R.T.C., Metallurgist, The Glacier Metal Co., Ltd., No. 3 Factory, Kilmarnock.

Frehner, Roy Charles, Research Assistant, Research Laboratories, The British Aluminium Co., Ltd., Chalfont Park, Gerrards Cross, Bucks.

Jones, Bernard Joseph, L.I.M., Metallographer, South Wales Aluminium Co., Ltd., Rheola Works, Resolven, near Neath, Glam.

Meredith, Keith Edward George, Assistant Metallurgist, Northern Aluminium Co., Ltd., Rogerstone, Newport, Mon.

Morris, Russell R., Experimental Department, Non-Ferrous Die Casting Co., Ltd., Nonferdica Works, North Circular Rd., London, N.W.2.

Waddington, Peter Hatten Crouch, Standards Engineer, The English Electric Co., Ltd., Stafford.

As Student Members

Bidwell, Henry Thomas, Metallurgical Apprentice, The Bristol Aeroplane Co., Ltd., Filton, Bristol.

Blair, Ian Scott, Student, Department of Metallurgy, University of Manchester.

Britton, Roger David, Metallurgical Apprentice, The Bristol Aeroplane Co., Ltd., Filton, Bristol.

Callaghan, John, Student of Metallurgy, University of Liverpool.

Dalton, Patrick Leonard, Laboratory Assistant, British Non-Ferrous Metals Research Association, London, N.W.1.

Dudley, John J., Bayliss, Jones and Bayliss, Ltd., Gable Street, Wolverhampton.

Eggleton, Maurice Arthur, Metallurgical Analyst, Earle, Bourne and Co., Ltd., Heath Street South, Birmingham 18.

Furmidge, John Edward, Student of Metallurgy, King's College, Newcastle-on-Tyne.

Garstone, John, B.Met., Post-Graduate Research Student, University of Sheffield.

Graham, Charles D., Jr., B.Met.E., Research Student, Department of Metallurgy, University of Birmingham.

Graham, Leslie Wilfred, Student, Department of Metallurgy, University of Manchester.

Hart, Roy Vincent, Student, Department of Metallurgy, University of Manchester.

Hill, Roger Bryan, B.Sc., Research Student, Department of Metallurgy, University of Manchester.

James, John Melvyn, Junior Chemist, Richard Thomas and Baldwins, Ltd., Wern Aluminium Works, Briton Ferry, Neath, Glam.

Kellert, Bernard George, Metallurgical Laboratory Assistant, 23 Aylands Road, Enfield, Middlesex.

Phillips, Derek John, Student, Department of Metallurgy, University College, Swansea.

Salkield, John Alan, Student, Department of Metallurgy, Sir John Cass College, Aldgate, London, E.C.3.

The following 17 Ordinary Members, 1 Junior Member and 8 Student Members were elected on 25 February 1953:

As Ordinary Members

Anderson, Alan McNaught, Director and Production Manager, The Wolverhampton Metal Co., Ltd., Wednesfield, Wolverhampton.

Bridge, Francis Downie, B.Sc., A.R.T.C., A.R.I.C., A.I.M., Head, Metallurgy Section, G.K.N. Group Research Laboratory, Lanesfield, Wolverhampton.

Davis, Anthony John, B.Sc., Metallurgist, Standard Telephones and Cables, Ltd., Footscray, Kent.

Garwood, Ronald David, M.Sc., Lecturer, Department of Metallurgy and Fuel Technology, University College, Cardiff.

Hackett, Walter, Jr., Joint Managing Director, Accles and Pollock, Ltd., Oldbury, Birmingham.

Hirsh, Walter L., Metallurgist, Research Laboratories of The General Electric Company, Ltd., Wembley, Middlesex.

Indulski, Jerzy, G.I.Mech.E., Technical Assistant, Standards and Specifications, The English Electric Co., Ltd., Rugby.

James, Thomas Claude, Chairman and Managing Director, The Wolverhampton Metal Co., Ltd., Wednesfield, Wolverhampton.

Johnson, Edwin Done, O.B.E., A.I.E.E., Production Director, The Indian Cable Co., Ltd., Golmuri, Jamshedpur 3, India.

Jones, (*Mrs.*) Joyce Eveline, B.Sc., D.R.T.C., Senior Scientific Officer, Metallurgical Branch, Armament Research Establishment, Woolwich, London, S.E.18.

Miller, John Leonard, D.Eng., Ph.D., M.I.E.E., F.Inst.P., Chief Consultant Engineer, British Insulated Callender's Cables, Ltd., Prescot, Lancs.

Munir, Mohamed Zaki, B.Sc., Ph.D., Senior Lecturer, Technology and Metallurgy, Cairo University, Egypt.

Pinheiro, Vítor Pinto, Lecturer in Metallurgy, Laboratório de Metalurgía, Instituto Superior Técnico, (affiliated to Universidade Técnica de Lisbõa), Lisbon, Portugal.

Richardson, Henry Taylor, B.Sc., Works Metallurgist, W. J. Fraser and Co., Ltd., Monk Bretton Works, Barnsley, Yorks.

Sheppard, John Alan, Technical Manager, The Metallizing Equipment Co., Ltd., 58 Victoria Street, London, S.W.1.

Weldon, Brian Arthur, B.Sc., Investigator, British Non-Ferrous Metals Research Association, Euston Street, London, N.W.1.

Young, John Joseph, General Manager, Minworth Metals, Ltd., Minworth, Birmingham.

As Junior Member

Savas, John, B.S.(Met.Eng.), Metallurgist, Packard Motor Car Company, Detroit 32, Mich., U.S.A.

As Student Members

Adams, Howland K., Jr., B.Met.Eng., Graduate Student, Department of Metallurgy, Rensselaer Polytechnic Institute, Troy, N.Y., U.S.A.

Belk, John Anthony, B.Sc., Research Student, Department of Physics, Royal Holloway College (University of London), Englefield Green, Surrey.

Burdon, Peter James, Scientific Assistant, Mitcham Works, Ltd., Mitcham Junction, Surrey.

King, Errol George, B.Sc.(Eng.), Scientific Officer, Ministry of Supply Factory, Springfields, Salwick, near Preston, Lancs.

Llewellyn, David Thomas, Student of Metallurgy, University College, Swansea.

Martin, John Cory, Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Skinner, John, Student of Metallurgy, Sir John Cass College, London, E.C.3.

Tangri, Krishan K., M.Sc., Research Fellow, Metallurgy Department, Missouri School of Mines and Metallurgy, Rolla, Mo., U.S.A.

PERSONALITIES

Professor Georg Masing

(Institute of Metals (Platinum) Medallist 1953)

Georg Masing was born on the 2 February 1885, in St. Petersburg, Russia, where his father was a chemist. He was educated at the Humanistische Gymnasium there, and then studied chemistry in St. Petersburg before going, in the autumn of 1905, to Göttingen University as a pupil of Professor G. Tammann. He received his Doctorate in 1909 for a thesis "On the Formation of Compounds under Pressure and the Reactions of Metals in the Solid State".

He then entered industry in Berlin, and in 1922 was made head of the Metallographic Department of the Research Laboratory of the Siemens concern. In that capacity he carried out extensive experiments in many fields. As a result of his work, Masing was offered the Directorship of the K.W. Institut für Metallforschung, Stuttgart, when it was founded in 1933, but he declined the post. In 1937 the Deutsche Gesellschaft für Metallkunde awarded him the Heyn medal, and in the same year he accepted the appointment of Director of the Institut für allgemeine Metallkunde at Göttingen University. He has held that position, in succession to Professor Tammann, ever since. In December 1952 the Technical University of West Berlin made him an honorary Doktor-Ingenieur.

Masing's work covers a great range of metallurgical subjects, both theoretical and practical, and he is the author of a large number of papers in the technical and scientific press. In 1926 he developed the alloys of beryllium in co-operation with O. Dahl. After two smaller books, "Ternäre Systeme" and "Grundlagen der Metallkunde", he published his comprehensive "Lehrbuch der allgemeinen Metallkunde" in 1950.

Dr. C. E. Ransley

(Rosenhain Medallist 1953)

Charles Eric Ransley was born in London in 1910 and educated at Wembley County School. On leaving school he joined the Research Laboratories of The General Electric Co., Ltd., as a student-assistant, and graduated in chemistry at London University in 1932. He took an M.Sc. degree in 1936, and gained a Ph.D. in 1940 for his work in the field of gas–metal reactions. As a member of the metallurgical staff of the G.E.C. Laboratories, he was engaged in research on various problems connected with the manufacture of lamps and thermionic valves, but during the period 1940–1945 he worked mainly on the development of special devices for radar applications. In 1945, he took up his present position with The British Aluminium Co., Ltd., at their then newly-established Laboratories at Gerrards Cross, Bucks, where he is in charge of a group studying various aspects of the casting and fabrication of aluminium and its alloys.

Dr. Ransley has published a number of papers on metallurgical subjects in the *Journal* of the Institute of Metals and elsewhere, and serves on several committees concerned with various aspects of metallurgical research. He is the present Chairman of the London Local Section of the Institute.

Mr. J. F. Waight

(W. H. A. Robertson Medallist 1953)

John Francis Waight received his technical education at the Woolwich Polytechnic and graduated at London University in 1935, with a degree in Engineering.

He was employed then by the South Metropolitan Gas Company, where he was engaged in the Industrial Department and the Physical Laboratory, on the design and construction of gas-fired heat-exchange equipment.

In 1945 he transferred to the City of Birmingham Gas Department as Assistant Superintendent of the Industrial Heating Section. He is now Industrial Engineer to the West Midlands Gas Board.

PERSONAL NOTES

Mr. E. G. Bayliss, Chief Metallurgist to John Garrington and Sons, Ltd., has been made an Honorary Associate of the Birmingham College of Technology.

Mr. W. F. Brazener, Managing Director of The Mint, Birmingham, Ltd., has been made an Honorary Associate of the Birmingham College of Technology.

Mr. E. H. Bucknall has been appointed Director of the recently established National Metallurgical Laboratory, Jamshedpur, India, and left England on 12 February. Since 1944 he had been Principal Assistant to the Superintendent of the Birmingham Laboratory of The Mond Nickel Co. Ltd. For a number of years Mr. Bucknall was Secretary of the Birmingham Local Section of the Institute.

Mr. D. Davies has left Aluminium Die-Castings (Birmingham), Ltd., in order to take up an appointment as Chief Metallurgist to The Mint, Birmingham, Ltd.

Mr. T. Fitzgerald has left the National Gas Turbine Establishment, Leicester, and returned to the Defence Research Laboratories, Melbourne, Australia.

Mr. O. B. J. Fraser, Assistant Manager, Research and Development Division, The International Nickel Co., Inc., has received the Samuel Wylie Miller award of the American Welding Society, for meritorious achievements in the art of welding.

Professor André Guinier, Professor of Physics at the Sorbonne, has been appointed visiting Professor of Physical Metallurgy in the Department of Mining and Metallurgical Engineering, University of Illinois.

Dr. D. S. Oliver has relinquished his post in the Physics Department, University of Bristol, and is now employed by the Ministry of Supply at the Metallurgical Laboratories, Division of Atomic Energy (P), Culcheth, near Warrington, Lancs.

Mr. E. W. Roberts has left England for Bahrain, Persian Gulf, to take up an appointment with the Bahrain Petroleum Co., Ltd.

Mr. Stanley Robson has been appointed Secretary of the Royal Institution.

Mr. C. F. Rogers is now engaged as Metallurgist with Bound Brook Bearings (G.B.), Ltd., Lichfield.

Mr. James S. Vanick, Research Metallurgist, The International Nickel Co., Inc., New York, has received the annual award of the Grey Iron Founders' Society for contributions to the grey iron-founding industry.

Deaths

The Editor regrets to announce the deaths of:

Mr. Edward M. Foster, Manager of Beryllium and Copper Alloys, Ltd., on 28 January 1953.

Mr. William Arthur Fowler, Production Manager of The British Aluminium Co., Ltd. Mr. Fowler died suddenly at the Company's offices in St. James's Square, London, on 16 February.

OBITUARY

Mr. A. Pinkerton

Andrew Pinkerton, who died on 11 January 1953, was the elder son of Alexander Pinkerton of Glasgow and was born on the 10 June 1888. He received his early education at the Allan Glen School, Glasgow, and from there passed to the Royal Technical College. His first appointment was with the Glasgow City Analyst, with whom he spent three or four years. His next post was with the Rio Tinto Copper Mines in Spain, where he remained until the outbreak of the First World War, returning to join the Forces. From the ranks of an infantry battalion he obtained a commission, and with his technical qualifications he was posted to the special gas company of the Royal Engineers. His military service was distinguished, and he was awarded the Military Cross early in 1918. On demobilization he joined the metallurgical staff of Earle, Bourne and Co., Ltd., Birmingham, and in a year or so became Chief Metallurgist on the resignation of the then chief. He rendered notable service to the Company, and was responsible, in large measure, for the technical advances made by the Company in its varied range of products.

Mr. Pinkerton enjoyed a considerable reputation in the non-ferrous metal industry, and his contributions and services to the metallurgical societies were outstanding. He had been a Member of the Institute since 1923, and had served as President of the Birmingham Metallurgical Society. He had a very strict sense of personal honour, and all who had the benefit of his friendship held him in the highest esteem.

LETTERS TO THE EDITOR

Picturesque " Whisker " Growth on Copper

In view of the scanty published data on whisker growth on metals in general[1] and on copper in particular,[2,3] it may be worth putting on record some recent observations, together with some tentative suggestions as to the possible mechanism of formation of whiskers.

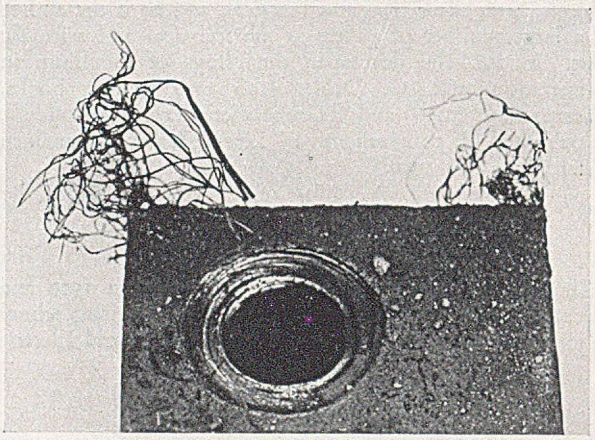

Fig. 1.—Whisker Growth on Block of High-Conductivity Copper. × 2 approx.

Compton, Mendizza, and Arnold,[1] in an investigation of filamentary growth, mainly on plated surfaces, stated that copper specimens develop whiskers only in the presence of sulphur (at temperatures up to 110° F. (43° C.)). Percy,[2] in 1861, recorded the formation of the so-called " moss copper " during the smelting of certain copper ores. Although moss copper forms in the region of 1100° C., the phenomenon appears to be similar to that described below. Liversidge[3] considered the formation of moss copper, silver, and gold as occurring " spontaneously ", and classed the phenomenon as " crystallization by thermo-reduction ".

Fig. 1 shows an example of the growth observed on a block of commercial high-conductivity copper. The copper specimen was placed inside a furnace used for annealing large steel sheets. The heat-treatment consisted of heating slowly to 850° C., maintaining this temperature for approximately

24 hr., and cooling slowly; the total time of heat-treatment was about 3 days. The atmosphere was "its own", i.e. atmospheric air initially, changing, by reaction with the silicon-iron, to a complex mixture consisting mainly of N_2, CO, CO_2, CH_4, and H_2.

Unlike the whiskers found on cadmium, zinc, tin, &c.,[1] the copper whiskers formed in the present investigation are neither single crystals nor twinned, but are polycrystalline, as was shown by Laue X-ray photographs. The diameter of the whiskers varies from 0·01 to 0·001 cm., which is much larger than those found hitherto. They are not always cylindrical, but have an "extruded" appearance, with numerous parallel grooves along their length.

Micrographs of the copper block were examined before and after the heat-treatment. Numerous cavities, both spherical and intergranular, in the heat-treated specimen suggested that, owing to segregation of impurities, low-melting-point regions are formed locally, and subsequently ooze out. This theory, however, is disproved by the fact that the whiskers consist of copper of the same lattice dimensions (within experimental error) as the base. A large percentage of alloying element would be required to depress the melting point of copper by 200° C., and furthermore such an alloy would tend to form a globule immediately after its protrusion through the surface. Such globules can be observed in the photograph, in addition to the whiskers. Spectrographic analysis indicated the presence of silicon, aluminium, manganese, and tin in the whiskers, though these elements were not present in the base metal, and the amount of iron present had apparently increased. These facts are most probably due to the formation of an evaporated film of these elements on the whiskers. Their presence in solution was not detected by X-ray techniques.

Bearing in mind the extruded appearance, an alternative explanation is that hydrogen in the atmosphere diffuses into the copper and reacts with the oxide inclusions to form steam. This steam will be under high pressure, and as the copper at the temperatures concerned will be mechanically weak, some of the copper will be caused to spurt out in the form of whiskers.

The precise mechanism of whisker growth is, however, by no means clear, and the tentative suggestions proposed above are open to criticism. This is further apparent when it is borne in mind that growth occurs only when the specimens are above earth potential (in this case 200 V.). The manner in which the whiskers tend to concentrate at the corners of a specimen suggests that the growth is connected with electrical discharge. The author would welcome suggestions to explain this interesting phenomenon.

R. J. WAKELIN

G.K.N. Group Research Laboratory,
Wolverhampton.

REFERENCES

1. K. G. Compton, A. Mendizza, and S. M. Arnold, *Corrosion*, 1951, **7**, (10), 327.
2. John Percy, "Metallurgy", p. 359. London: 1861.
3. A. Liversidge, *Chem. News*, 1877, **35**, 68.

Intercrystalline Cracking of Metals

In connection with recent correspondence on intercrystalline failure,[1,2] I should like to draw attention to an unusual example, namely, delayed intercrystalline failure at room temperature in a "stiff" metal.

This failure occurs in cast, ternary β-brass containing 4% aluminium. In the ordinary tensile test the brass has approximately the following characteristics: ultimate tensile stress, 40 tons/in.^{2}; elongation, 14%; limit of proportionality, 10 tons/in.^{2}; 0·1% proof stress, 24 tons/in.^{2}. In this test the fracture is transcrystalline. However, under a sustained tensile stress of 25 tons/in.^{2}, in air, intercrystalline cracking and failure occur within a few days. After the initial loading, no appreciable extension has been detected before fracture. Similar failure also occurs when the specimen is immersed in liquid paraffin and machine oil.

The weakness was discovered during an investigation into the intercrystalline failure of high-tensile β-brasses immersed in sea-water.[3] Although no aggravation by sea-water of the failure of the aluminium brass under sustained stress could be detected, immersion in sea-water caused lower figures in the ordinary tensile test and fractures of intercrystalline inception.

There is some information indicating that the weakness of the brass may be due to the presence of extremely fine particles of the brittle γ phase. On this basis there are likely to be two factors at work. First, in air under sustained stress the growth of cracks from the brittle phase is indicated; in this connection, it should be noted that the time for failure is not constant, ranging from a few hours to several days. The second feature is the aggravating effect of the sodium chloride in the ordinary tensile test. This may be an electrochemical action producing stress-corrosion, or alternatively a surface-tension effect.[4]

Experiments are in progress to test these ideas and to determine whether the mechanical brittleness is manifested in other ways. Meanwhile, it would be interesting to have views on how the failure in question fits in with the theories proposed in the earlier correspondence.

A. R. BAILEY

Constantine Technical College,
Middlesbrough.

REFERENCES

1. J. N. Greenwood, *Bull. Inst. Metals*, 1952, **1**, (12), 104; (14), 120.
2. D. N. Frey, *ibid.*, 1952, **1**, (15), 126.
3. A. R. Bailey, *Metal Ind.*, 1952, **80**, (26), 519.
4. C. Benedicks, *Rev. Mét.*, 1948, **45**, 9.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

London Local Section

On 21 January the London Local Section held a joint meeting, at University College, with the London and South Eastern Counties Section of the Royal Institute of Chemistry and the London Section of the Society of Chemical Industry. The speaker was Dr. A. C. MENZIES and his subject was:

Analysis of Metals by Spectroscopy

The lecturer began by outlining and demonstrating the production of spectra by gratings and prisms, using as his practical subject mercury vapour. The development of emission instruments was then traced from earliest times to the present day. Modern developments are now principally concerned with improved and more rapid methods of line-density measurement, but improved source units, such as the triggered arc circuit, are being developed.

Dr. Menzies discussed the difficulties attending accurate determinations of certain elements, and quoted as examples phosphorus and sulphur in steel. In extreme cases resort may have to be made to evacuated and short-path instruments to reduce absorption.

The theory of emission and absorption of spectral lines was described in some detail. A study of the electronic changes in energy level accompanying emission of radiation enables the optimum line-pairs for analysis of particular combinations of elements to be selected, and the factors influencing such a selection were discussed.

A practical demonstration, given with a three-electrode arc, served a twofold purpose: it showed migration of ions in a D.C. arc and the preferential emission of radiation requiring a lower excitation potential; in the example used, emission from copper was partially suppressed by potassium.

Finally, the newer direct-reading instruments were described. A demonstration of the possibilities of this type of instrument was given, using a medium quartz spectrograph with photo-multiplier equipment.

At a meeting of the Section held at 4 Grosvenor Gardens, London, S.W.1, on 5 February, Dr. N. P. ALLEN, Superintendent of the Metallurgy Division, National Physical Laboratory, Teddington, gave a lecture:

Titanium

Dr. Allen said that the manufacture of titanium and its alloys had developed in the U.S.A. during the last six years and had now reached a scale such that masses of over a ton weight were handled, and the output was of the order of 6–8 tons per day. Production on a somewhat smaller scale had been developed in England. The commercially pure metal had the strength of a mild or medium-carbon steel, and had good hot-working properties. The working properties in the cold, on account of the hexagonal structure of the metal, were not so good. Oxygen, nitrogen, carbon, and hydrogen had pronounced effects on its ductility. The alloys of titanium could be classified and understood by considering the effect of the added element on the temperature of the α/β change. Some alloys had the strength and toughness of heat-treated high-tensile alloy steels. Their properties were not yet fully under control, however, and study of the impurities in the alloys was important.

The titanium alloys had outstanding resistance to corrosion and corrosion-fatigue. Their strength was well maintained at temperatures up to 400° or 500° C., though above this temperature they declined seriously. Alloys were available that up to 250° C. or higher were superior to austenitic steels and aluminium alloys on a strength-for-weight basis. Titanium alloys were likely to replace high-tensile alloy steels and stainless steels to a considerable extent in aircraft, in positions where the temperature was not above 500° C., but, on account of their low modulus of elasticity, they would not replace aluminium alloys in aircraft structures for some time. The high cost of titanium was likely to hinder its wide general application, unless improved and cheaper methods of extraction could be found.

South Wales Local Section

At a meeting held at University College, Swansea, on 3 March, Mr. E. S. W. EARDLEY, of The Incandescent Heat Co., Ltd., gave a lecture on:

Recent Advances in Furnace Design

The lecturer began by describing the developments which had taken place in the past few years in the component parts that went to make up the furnace structure proper. These included: the concentrated combustion burner; the dual-type burner, suitable for use with either oil or gas; the jet-type recirculating radiant tube; developments in flat arch construction; and the centrifugally cast roller in heat-resisting alloys.

Mention was made of the way in which dust collection, the hot-blast cupola, continuous dielectric core drying, and low-temperature annealing, had improved foundry practice.

The paper outlined the salient features of the walking-beam furnace, its application to bar and pack heating in mechanized mills, and to plate heating.

Mr. Eardley went on to point out the advantages of continuous strand annealing of ferrous and non-ferrous materials, and also described recent advances in direct-fired lift-off furnaces for the annealing of steel sheets, high-speed billet heating, gas carburizing, low-temperature recirculation for galvanizing baths, &c., continuous heat-treatment lines for armour plate, and a new furnace for heating bars.

Finally, a description was given of equipment for the production of controlled atmospheres.

JOINT ACTIVITIES

Joint Committee for National Certificates in Metallurgy

The Joint Committee of the Iron and Steel Institute, the Institution of Mining and Metallurgy, the Institute of Metals, and the Institution of Metallurgists, in conjunction with the Ministry of Education, has issued its report on the progress of the scheme for the award of National Certificates in Metallurgy for the year 1951–52. Dr. J. W. Jenkin is the Chairman.

The Joint Committee has approved five new Schemes for National Certificates in Metallurgy during 1952; other Schemes have been revised, and Schemes are also being organized in districts not yet covered, with a view to suitable courses being started as soon as possible.

Final examinations of the Course for the Ordinary National Certificate have been held during 1952 at 28 Technical Colleges, and for the Higher National Certificate at 16 Technical Colleges, for candidates who have satisfied the conditions laid down in Ministry of Education Rules 111, under which the Schemes are operated.

The Committee is pleased to report that: (*a*) 178 candidates have qualified for the award of an Ordinary National Certificate in Metallurgy and 103 candidates have qualified for the award of a Higher National Certificate as a result of final examinations held in 1952. (*b*) The records of a further 5 candidates for the Ordinary Certificate and 3 for the Higher have been approved by the Joint Committee, subject to the fulfilment of certain conditions. (*c*) Eight candidates for the Ordinary and six candidates for the Higher Certificate, who entered the final examinations in 1951, and whose records were approved by the Joint Committee subject to certain provisions, have qualified for the award of Certificates in 1952. (*d*) The Committee has also approved the award of the Ordinary Certificate to 8 candidates, and the Higher Certificate to 17 candidates, who entered the final examinations in 1951, but whose records were not available at the date of publication of the Report for 1950–51. (*e*) Distinctions have been gained by 38 candidates who have shown an exceptional grasp of their subjects, indicating a high degree of training and knowledge in the particular subject in which the Distinction has been gained.

Comparative figures of entries and successful candidates for the seven years since the institution of the Scheme for National Certificates in Metallurgy are as follows:

(i) *Ordinary National Certificate*

Year	Technical Colleges	Total Number of Candidates			
		Entered	Awarded Certificates	With Distinction	With Prize Award
1946	4	28	23	4	7
1947	11	102	69	14	17
1948	15	123	74	12	16
1949	21	222	107	24	27
1950	20	281	142	23	26
1951	25	314	138	17	24
1952	28	334	186 *	22	31

(ii) *Higher National Certificate*

Year	Technical Colleges	Total Number of Candidates			
		Entered	Awarded Certificates	With Distinction	With Prize Award
1946	—	—	—	—	—
1947	2	22	18	—	1
1948	8	38	22	1	2
1949	10	74	59	20	22
1950	11	62	42	8	10
1951	16	110	54	5	11
1952	16	132	120 *	16	22

* The results of 8 candidates for the Ordinary Certificate and 17 candidates for the Higher Certificate whose records were not available in 1951 have been included in 1952.

Prizes have been awarded in 1952 to 53 successful candidates who have shown particular merit in the final examinations, from the Prize Fund established by the Iron and Steel Institute, the Institution of Mining and Metallurgy, and the Institute of Metals for the purpose.

Attention is drawn to the revised Regulations for Admission to Membership of The Institution of Metallurgists, published in September, 1951, which provide for admission to Licentiateship of the Institution by examination of candidates holding an Ordinary National Certificate in Metallurgy in which Chemistry and Physics are assessed subjects in the final year, and Mathematics an assessed subject in the pre-final year. The Regulations also make provision for admission to Licentiateship and for exemption from the Institution's examination for candidates producing evidence as follows:

> "D.—(1) Having obtained Higher National Certificate in Metallurgy and having been notified in the pass list issued by the Joint Committee for National Certificates in Metallurgy that exemption standard has been reached in at least three subjects of the Final Examination (including endorsement subjects). Holders will be exempted from the Institution's examinations on a subject for subject basis. Exemption standard will approach the standard required for distinction. Candidates claiming exemptions under this regulation must have obtained their Certificate at a College whose syllabuses have been approved by the Institution for the purpose of this regulation.
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> (2) All candidates granted exemption under this regulation will be required to sit the Institution's examinations in the remaining sections (if any) for the Licentiateship (see page 13 of the Regulations), and must (*a*) produce evidence of having done metallurgical work of sufficient merit and (*b*) attend for an oral examination or interview."

The Joint Committee has agreed to an amendment of paragraphs 2 and 3 of Section 17 of Ministry of Education Rules 111. The two paragraphs in question will be re-worded to read:

> "As soon as possible after the scripts of the examination of the final year have been marked by the examiners, and in any case not later than 1 July in any year, the examinations officer must forward them to the Committee, together with Forms 329 F.E. and 330 F.E. completed as far as applicable. The Forms 329 F.E. and 330 F.E. will exhibit the records of the candidates for the several years of their course in respect of—
>
> (i) Attendance.
> (ii) Marks for homework, for laboratory work and records, and for drawings, if any.
> (iii) Examination marks."

A short Refresher Course in Metallurgy organized by the Ministry of Education for teachers of Metallurgy in Techincal Colleges was held during the Easter Vacation, 1952. The Joint Committee is pleased to note that the course, which dealt with modern metallurgical techniques, was very well attended and successful, and the Chairman, who visited the Course, was much impressed by the enthusiasm of those attending.

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee now invites applications for the award of Mond Nickel Fellowships for the year 1953. Awards will be made to selected applicants of British nationality educated to University degree or similar standard, though not necessarily qualified in metallurgy, who wish to undergo a programme of training in industrial establishments; they will normally take the form of travelling Fellowships; awards for training at Universities may be made in special circumstances. There are no age limits, though awards will seldom be given to persons over 35 years of age. Each Fellowship will occupy one full working year. The Committee hope to award up to five Fellowships each year to an approximate value of £900 to £1200 each.

Mond Nickel Fellowships will be awarded in furtherance of the following objects:

(*a*) To allow selected persons to pursue such training as will make them better capable of applying the results of research to the problems and processes of the British metallurgical and metal-using industries.

(*b*) To increase the number of persons who, if they are subsequently employed in executive and administrative positions in the British metallurgical and metal-using industries, will be competent to appreciate the technological significance of research and its results.

(*c*) To assist persons with qualifications in metallurgy to obtain additional training helpful in enabling them ultimately to assume executive and administrative positions in British metallurgical and metal-using industries.

(*d*) To provide training facilities whereby persons qualified in sciences other than metallurgy may be attracted into the metallurgical field and may help to alleviate the shortage of qualified metallurgists available to industry.

Applicants will be required to state the programme of training in respect of which they are applying for an award, as well as particulars of their education, qualifications, and previous career. Full particulars and form of application can be obtained from: The Secretary, Mond Nickel Fellowships Committee, 4 Grosvenor Gardens, London, S.W.1.

Completed application forms will be required to reach the Secretary of the Committee not later than 1 June 1953.

NEWS OF KINDRED SOCIETIES

Institution of Metallurgists Examinations

The next Examinations for the Licentiateship and Associateship of The Institution of Metallurgists will be held from 24 August to 1 September 1953. Candidates must submit their applications for permission to enter the examinations before 1 May 1953.

Each application must be made on a form to be obtained from the Registrar-Secretary, The Institution of Metallurgists, 4 Grosvenor Gardens, London, S.W.1, and must be accompanied by a registration fee of one guinea. The balance of the examination fee will be payable before 1 August 1953.

Details of the various Centres at which the Examinations will be held, and arrangements for the Examinations will be forwarded to all candidates in due course.

Examinations for the Fellowship will also be held between 24 August and 1 September 1953, and intending candidates should apply for permission to enter for the examination, submitting, for the approval of the Council, a statement of the subject of the dissertation or the branch of metallurgy in which they offer themselves for examination.

The Physical Society Exhibition of Scientific Instruments and Apparatus

The 37th Annual Exhibition of the Physical Society will be held at the Imperial College of Science and Technology, Imperial Institute Road, London, S.W.7, from Monday, 13 April to Friday, 17 April 1953. A return has been made to the practice of former years of locating the whole of the exhibition in the main buildings of the College, and this should be of great convenience to visitors.

Following the normal policy of the Society, the great number of exhibits appeals primarily to physicists, and much new and original equipment will be on show. Among the many modern aspects of science dealt with in the exhibition are the problems arising out of high-speed flight, industrial use of radio-isotopes, and the applications of telecommunications and radar in modern travel.

As in previous years, the comprehensive Handbook of the Exhibition will be available at the Exhibition, and copies can be obtained on application to the Secretary-Editor, the Physical Society, 1 Lowther Gardens, Prince Consort Road, London, S.W.7. The price of the publication is 6*s*. (by post 7*s*. 3*d*.).

OTHER NEWS

Research on the Constitution of Alloys

Several significant gaps in current research on alloy constitutions are commented on in a statement by the Constitution of Alloys Group (Chairman: Professor G. V. Raynor) of the British Iron and Steel Research Association. The Group also indicates some ways in which it is willing to help fill the gaps.

In calling attention to these possible lines of research, the Group emphasizes its desire to encourage new and existing work on the subject of the constitution of alloys. It is prepared to suggest systems requiring investigation, and to put enquirers in touch with other workers in the same or similar fields. In particular cases, where the results of such work are likely to be of direct interest to the iron and steel industry, the Group is able to make recommendations for some financial assistance. The Group would also be glad to know of existing equilibrium-diagram work which has not been reported to it, and of projected research programmes involving the constitution of alloys. Those interested should communicate with the Secretary, Constitution of Alloys Group, British Iron and Steel Research Association, 11 Park Lane, London, W.1.

The gaps in research work that the Group's reviews have indicated can be summarized as follows:

(i) Although there appears to be no obvious gap in research work on high-temperature materials, a greater effort might well be expended in this field, owing to the relatively slow rate of progress dictated by the difficulties. In particular, more work is needed on the higher-melting-point alloys of Groups V and VI. Concerning materials to resist wear and high temperatures, it is felt that much more information is required on systems containing borides and that equilibrium relations in ferrous alloys containing silicides are probably equally in need of study.

(ii) There is a surprising lack of research in the general field of ferrous alloys, except in so far as the constitution of several alloy systems involving phases with the structure of the Fe–Cr sigma phase is being adequately studied. It is felt that certain systematic work on the binary alloys of iron would be of value, both from the theoretical point of view and to serve as a basis for studies of the complex equilibria involved in ternary and quaternary alloys. Further, the relationships between the various carbides present in alloy steels offer a wide field.

(iii) The constitution of magnesium alloys appears to be receiving little attention, and in view of the importance of these materials to aeronautical engineering, a more concentrated study may be necessary, particularly of ternary and more complex alloys. The response to heat-treatment of magnesium alloys is, in general, not good, and constitutional work on this subject would be of value.

(iv) The constitutional background required for the proper development of materials of the high-tensile-brass type and the complex bronzes is still obscure.

(v) On the theoretical side, it appears to the Group that the contribution made by those interested in the determination of the crystal structures of intermetallic phases and in accurate lattice-spacing measurements is of much importance, and that greater effort in this field would not be inappropriate. Further problems connected with the theory of metals and alloys arise in the purely physical field (e.g. X-ray emission and absorption experiments on alloys, magnetic properties, low-temperature specific heats), and a more intimate contact between the physicist and the metallurgist interested in the general theory of alloys would be an advantage.

It should be noted that in this statement no attempt is made to review the present position with regard to all types of alloy. No mention has been made of cases in which the Group felt that the research effort was reasonable, having regard to the various problems involved. There is no implication that subjects and systems not mentioned are of lesser interest.

Protection from Radiation

In view of the growing use of X-rays and radiation from radioactive isotopes in industry and elsewhere, the Board of the Institute of Physics has again directed attention to the need to take simple precautions against risks to health arising from the absorption of radiation. A short bibliography of information on the subject is given below. Further information can be obtained from any of the following:

The Director, National Physical Laboratory, Teddington, Middlesex.

The Hospital Physicist at the nearest hospital.

Regarding industrial uses

The Chief Inspector of Factories, 8 St. James's Square, London, S.W.1.

Regarding the use of radioactive isotopes

The Director, Atomic Energy Research Establishment, Harwell, Didcot, Berkshire.

BIBLIOGRAPHY

1. *Memorandum on Gamma-Ray Sources for Radiography* (obtainable from The Institute of Physics, 47 Belgrave Square, London, S.W.1. Price 3s. 6d. net).
2. *Handbook of Industrial Radiology* (published for The Institute of Physics by Edward Arnold and Co., 41 Maddox Street, London, W.1. Price 21s.)
3. *X-ray Protection Design*, by H. O. Wycoff and L. S. Taylor, being Handbook 50 of the National Bureau of Standards (published by the Government Printing Office, Washington 25, D.C., U.S.A. Price 15 cents).
4. *Factories (Luminising) Special Regulations* 1947, No. 865 (published by H.M. Stationery Office. Price 2d.)
5. *Recommendations of the International Commission on Radiological Protection*, 1950 (copies obtainable (gratis) from The Faculty of Radiologists, 45 Lincoln's Inn Fields, London, W.C.2.)
6. *Introductory Manual on the Control of Health Hazards from Radioactive Materials, Issue No.* 2, 1949 (copies obtainable (gratis) from the Medical Research Council, 38 Old Queen Street, London, S.W.1.)
7. *Recommendations of the British X-ray and Radium Protection Committee*, 7th Revised Report, October 1948 (obtainable from the Director, National Physical Laboratory, Teddington, Middlesex).

Summer School and Conference on the Theory of the Plastic Deformation of Metals

The H. H. Wills Physical Laboratory and the Department of Adult Education of the University of Bristol, in cooperation with The Institute of Physics, will be conducting a short Summer School followed by a conference on "The Theory of the Plastic Deformation of Metals, with Special Reference to Creep and to Fatigue" from 13 to 16 July next in Bristol. The provisional programme of the course which will precede the conference includes lectures by Professor N. F. Mott, Dr. A. J. Forty, and Dr. F. C. Frank. The course is similar in conception to those held in the University of Bristol on this and similar subjects; it is intended mainly for research students at universities and for members of the staffs of government and industrial laboratories. The particular aim is to see to what extent the observed phenomena can be explained in terms of present theories, and to guide future work.

The fee for the Summer School, which will be on 13 and 14 July, is £1 10*s.* 0*d.*; but there will be no fee for the conference.

Further particulars and forms of application, to be returned before 31 May, can be obtained either from the Director of the Department of Adult Education, The University, Bristol 8, or from the Secretary of The Institute of Physics, 47 Belgrave Square, London, S.W.1.

B.W.R.A. Summer School of Welding

The third Summer School of Welding organized by the British Welding Research Association will be held at Ashorne Hill, near Leamington Spa, Warwickshire, from Monday, 27 April, until Saturday, 2 May 1953.

This year the general theme of the School is welding fabrication and production, and the various aspects will be covered in logical sequence during the course. The School will start with introductory lectures giving critical and comparative studies of welding processes, plant, and basic materials. Students will then be given full instruction in welding design, followed by lectures on the study of works planning and welding costs and the more efficient use of up-to-date processes such as automatic welding, &c. Finally, the use of standardization, time-and-motion study, and other means of improving productivity will be dealt with. The course will also include lectures on the non-destructive testing and inspection of welds.

To assist industrial organizations whose staff may be more particularly interested in one aspect of welding, the School has been divided into certain groups. Students will be given lectures of general interest, usually in the morning, followed in the afternoon by lectures on more specialized subjects under five group headings:

Group A: Aircraft Industry
Group B: Shipbuilding Industry
Group C: Structural Engineering
Group D: Automobile and Sheet Metal Industries
Group E: General Engineering.

The School will be of great use to designers, technicians, engineers, and indeed to all those interested in improving welding methods and production efficiency. As vacancies are limited, it is suggested that those wishing to attend the School, or to send representatives from their organizations, should communicate as soon as possible with the Secretary, B.W.R.A. Summer School of Welding, British Welding Research Association, 29 Park Crescent, London, W.1, requesting the necessary application forms and full particulars.

Errors in the A.S.T.M. Index

The A.S.T.M. Index of X-ray Powder Diffraction Data has now been in use for about ten years, and it is recognized that it contains a number of wrong identifications which are likely to lead to confusion. The X-ray Analysis Group of The

Institute of Physics therefore proposes to collect information on the errors which have been observed by users of the Index and to publish as soon as possible in some suitable way a list of all those errors which seem to be substantiated. It is believed that many, if not all, of these errors have by now been noted by various workers and may in some cases have been mentioned casually in their publications, but in order to be of real use to others it is necessary that they should all be published collectively. Accordingly, everyone who has noted such errors is asked, whether he has already published them or not, to send the details to Dr. A. J. C. Wilson, University College, Cathays Park, Cardiff.

DIARY

9 April. Liverpool Metallurgical Society. Annual General Meeting, followed by "The Basic Processes Involved in the Tempering of Plain Carbon and Low-Alloy Steels", by Dr. W. S. Owen. (The Temple, Dale Street, Liverpool, at 7.0 p.m.)

15 April. Society of Chemical Industry, Corrosion Group. "The Corrosion of Aluminium and Its Alloys in Supply Waters", by C. F. Porter. (Chemical Society, Burlington House, Piccadilly, London, W.1, at 6.30 p.m.)

16 April. Institution of Mining and Metallurgy. General Meeting. (Geological Society, Burlington House, Piccadilly, London, W.1, at 5.0 p.m.)

17 April. Institute of Physics. Symposium on "Practical and Theoretical Aspects of Ultrasonic Testing". (Institute of Physics, 47 Belgrave Square, London, S.W.1, at 6.30 p.m.)

30 April and 1 May. Iron and Steel Institute. Annual General Meeting. (4 Grosvenor Gardens, London, S.W.1.)

APPOINTMENTS VACANT

A VACANCY exists for a man who has extensive experience in the extrusion of light alloys and who will be required to carry out development work on improving the productivity of presses and accelerating finishing operations. A person with experience on production of extruded sections, rods, tubes, and wire, would be preferred. Salary to be discussed at interview, but will be in range of £800–£1,100 per annum. Applications in writing to Staff Officer, Box No. 349, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

MAGNESIUM ELEKTRON, LTD., have vacancy for experienced metallurgist for Research Department. Must have personality and ability to control Inspection Section, as well as technical qualifications of normal average calibre. Written applications should be addressed to Secretary, Magnesium Elektron, Ltd., Clifton Junction, Nr. Manchester.

PRESIDENTIAL ADDRESS*

By Professor F. C. THOMPSON,† D.Met., M.Sc.

SYNOPSIS

A survey is made of the state of metallographic knowledge when the Institute of Metals was formed in 1908.

Election to the Presidency of the Institute of Metals is an honour which any metallurgist must highly appreciate. The responsibilities of the office are, however, heavy, and it is in a spirit of some trepidation that I have taken them over.

The Institute came into existence at a time when the foundations of scientific metallography had been well and truly laid. The advances which have been made since then have in no inconsiderable degree been the result of the work of our own Members, and much of this work is recorded in our *Journal*. The Institute may take a legitimate pride in the part which it has played in fostering knowledge in its own field of science. It seems worth while, therefore, to look back and spend a few minutes in considering the state that metallography had reached when the Institute came into existence in 1908.

The metallographic techniques then available were the microscopic examination of metals and alloys, the construction and interpretation of thermal-equilibrium diagrams, and their correlation with a wide range of chemical, physical, and mechanical properties. These techniques are still employed, though now enriched by others which have since been developed.

Although modern microscopic techniques, such as phase-contrast or the use of polarized light, had not then been applied to metallic systems, the results obtained by the older metallographer were of a high standard, being little, if at all, inferior to those obtained to-day. In no instance is the swing of the pendulum of scientific fashion more evident than in the return to favour of the microscope, with its new accessories, after a period of semi-eclipse, during which some people believed that other, and more modern, techniques had rendered it obsolescent and perhaps even obsolete. After Sorby's disappointment at the earlier neglect of his work, it can readily be imagined with what pleasure he would have welcomed these further developments of the microscope for the examination of metals.

In 1908 most of the methods now employed for the development of microstructures had already been introduced. Heat-tinting had been used with success by Martens, and methods for etching both at high temperature and in vacuum had been employed. A technique then common, but which has since tended to fade out, was the simultaneous polishing and etching, known as "polish-attack", but, in some forms of electrolytic preparation, something not very different from this has again come to the front.

Not only was it *known* from evidence drawn from a wide variety of sources that the grains in a metal or alloy were crystalline, but the occurrence of sub-crystalline structures had been observed by Andrews as early as 1895. Howe, too, had recognized grains of the "first and second orders"; what, in the case of iron, we should now call α-veining, the smaller grains being regarded by him as "irregular fragments into which the larger grains had been broken". That this sub-granular structure consists of fragments of almost identical orientation was supported by the already well-known fact that slip bands passed from one cell to another without appreciable deviation, unlike the marked change of direction which normally occurs at the crystal boundaries.

Work on the thermal-equilibrium diagrams of metals will always be connected with the names of Roozeboom and especially of Heycock and Neville, and it is of interest that in his recent book on thermodynamics of alloys Lumsden [1] deliberately employed some of their data as being, for his purpose, the best even yet available. Nor was the early work by any means confined to alloy systems of low melting point, for Carpenter and Keeling's iron–carbon diagram was published in 1904. Although such diagrams were essentially regarded as being expressions of experimental fact, the surfaces depicting the heat contents of alloys had already received consideration, a point of view which a little later was discussed by Tammann, and which clearly bears some relationship to the more modern concept of "free energy".

The preparation of pure metals by electrodeposition was an already established technique, and work on the properties of iron, using electrolytic material, which even to-day would be regarded as of high purity, had been done by many people. It was also appreciated that the properties of such iron as deposited differed greatly from those of the same metal in the normal condition, and further that this difference might be due on the one hand to some unknown change of structure, or on the other to the occlusion

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* Delivered at the Annual General Meeting, London, 24 March 1953.

† Professor of Metallurgy, University of Manchester.

of hydrogen, the effect of which on iron, and particularly on palladium, had been extensively studied.

The sub-division of metallic alloys into eutectics, solid solutions, and intermetallic compounds had long been understood, and the work of Matthiessen had correlated such metallographic structures with physical properties such as electrical conductivity. Although the arbitrary separation of metallography into the ferrous and non-ferrous fields is of long standing, it is, in fact, one indivisible branch of knowledge. This is well exemplified by Arnold's paper of 1895 on " The Influence of Carbon on Iron ".[2] Shocked to the core as he himself would have been to be told so, this paper is of importance in that it represents one of the earliest attempts to examine in detail the breakdown of a solid solution into a eutectoid.

At the time which we are considering, the idea of the superlattice had not, so far as I am aware, been enunciated, but the evidence again was there. When Roberts-Austen published his liquidus curves for the copper–zinc system in 1897, a horizontal line at a temperature of about 470° C. was introduced over the range in which the β-phase exists. I believe that Le Chatelier had also observed an abnormality in the cooling curve at this point, and I may perhaps be permitted to digress somewhat to remind you of the fact that it was Tammann who, in 1918, first suggested, in the words of Desch : " That a kind of isomerism was possible in alloys consisting of a solid solution ; that is, there might be two alloys of the same composition, each consisting of a single phase and having the same crystalline form but differing in properties. The differences would arise from the different distribution of the several kinds of atoms on the points of the space lattice." This work of Tammann is not as generally known as it deserves to be, and it is a pleasure to pay tribute to a master of our science.

The first attempt to elucidate the relationship between atomic volume and the mechanical properties of alloys seems to have been due to Roberts-Austen. His work, coupled with that of Arnold and Jefferson on gold, and of Arnold himself on iron, had demonstrated that, so long as the added element passed into solid solution, the relationship of the atomic volumes of the solvent and solute atoms was a potent factor in controlling the strength (or in other words the structure) of the alloy. Looking back on this work in the light of present knowledge, it is clear, although the fact was then but dimly perceived, that the whole evidence required for an appreciation of the influence of the " size-factor " was already there. The Second Report to the Alloys Research Committee[3] in 1893 concludes that : " In all probability, therefore, the introduction of free molecules of an added element must create a disturbance, the nature and magnitude of which will bear some relation to the volume of the disturbing atom." The other factors determining solid solubility were still unknown until they were later revealed by the work of Hume-Rothery. It is of interest, too, that the possibility of interstitial as well as of substitutional solid solutions was thoroughly well appreciated, and the necessity of the solute atom in the former case being small compared with that of the solvent was fully realized.

With the aid of the recording pyrometer which he had designed, Roberts-Austen investigated, amongst other things, the surfusion of metals and alloys. He showed, for instance, that in those alloys of lead and tin which were slightly richer in lead than the eutectic, considerable surfusion could occur, a fact which was later to assume great significance in connection with the process of " modification ".

Lacking modern resources such as X-ray methods of investigation, these earlier workers were at times compelled to rely on specially devised techniques, some of which were of considerable ingenuity. As an example may be quoted the proof by Benedicks that the bars in the austenite–martensite structure were, in fact, magnetic. This he did by placing the polished and etched specimen between the poles of a powerful electromagnet in a bath of colloidal iron in acetone, the iron being attracted to the magnetic " needles " and thus producing a structure similar to that obtained by normal etching. It may be that, with the wider range of techniques at the disposal of the present-day worker, this spirit of ingenuity is now less evident, and, if this be so, metallography has suffered a serious loss.

The classical work of Benedicks on the marked effect which might be exerted by stress, both internal and external, on changes in metallurgical systems had just appeared when this Institute was founded. It was already well known that if a high-carbon steel is quenched from a high temperature, a partial retention of the austenite can be effected. This austenite, however, instead of occurring at the surface as would be expected, since the rate of cooling there is a maximum, is, in fact, found in the centre of the bar. Benedicks's demonstration that austenite could be retained, even in plain carbon steels, right up to the surface by the imposition of compressional stress is one of the most ingenious experimental results of early metallography. He did this by fusing the surface of the steel in a carbonaceous container, thus obtaining a shell of the white-iron eutectic, the rigidity of which during the subsequent heat-treatment of the composite bar inhibited the expansion which results when austenite transforms to martensite, and so prevented the change. This was not by any means, however, the first attempt which had been made to correlate the condition of strain with the transformations in steel, and Roberts-Austen, as far back as 1893, had shown that the change points in steel could be depressed by the application of pressure, and had studied the corresponding effects on Newton's alloy, the ternary eutectic of bismuth, lead, and tin.

Diffusion in solid metals had been the subject of much work both in the ferrous and non-ferrous fields, by Roberts-Austen amongst others, and that such diffusion followed the normal Fick law was known. The diffusion of carbon in iron at very high temperatures had been investigated in detail by Arnold, and his results have not, so far as I am aware, been examined from the theoretical point of view even to-day. It was realized by some, although this was probably never published, that such solid diffusion involved a distortion of the crystal structure of the metal, and I well remember as a student a most satisfying explanation of the changes of structure brought about by normalizing an overheated steel, based on such lattice imperfections. The day of the " dislocation " had not yet dawned, but that the metallic crystals were, in most cases at any rate, far from perfect was well understood.

Although Wilm had not yet published his pioneer work on age-hardening, there were already, if eyes had been available to see it, clear indications of some such effect. In particular may be mentioned the fact, already well known, that the electrical and magnetic characteristics of soft-iron transformer cores changed in the course of time.

This " ageing "—and the term was already in use—was known to bring about a slow deterioration, resulting in reduced permeability and increased hysteresis. It was further known that these effects were not due to mechanical fatigue, but were the result of some change in the iron resulting from long-continued low-temperature treatment. The effect of both time and temperature on the hysteresis was examined by Roget, whose death was but recently reported, in 1898, and results were obtained identical with those of the classical researches on age-hardening. At as low a temperature as 65° C. and in a period of only 27 days, an increase of the hysteresis loss of over 53% was observed. As the temperature was raised, the rate of increase became greater, then a maximum was shown, and later a falling off, which became more and more rapid, whilst the maximum in the curve fell progressively, until at 700° C. the effect was almost non-existent.

It was the same situation that we have come across repeatedly. The facts were there, but the explanation of the facts awaited more mature consideration, and further accumulation of knowledge drawn from a wide variety of sources.

The first experiments in powder metallurgy had been conducted long before the period with which we are here concerned, and with such experiments the name of Spring of the University of Liège must always be associated. Not only did he succeed in producing compacts even of brittle metals such as bismuth, but he showed that eutectics could be synthesized by pressure, producing the quaternary eutectic of bismuth, lead, tin, and cadmium with a melting point corresponding to that of the eutectic itself, although the most fusible of the constituents of which the powders were made did not commence to melt till a temperature some 130° C. higher had been reached. When such compacts were heated, Tammann, amongst others, obtained the clearest evidence of diffusion, a fact of which some recent workers seem to have been unaware.

The metallography with which we are concerned was in the main linked with physical chemistry; it might almost be defined as the physical chemistry of the metallic state; and the re-orientation with the formation of a metallurgical–physical bond was still in the future. The revolution which resulted from the introduction of X-ray techniques, for instance, can perhaps be best appreciated by reading a text-book of metallography written before 1914, when Bragg worked out the structure of copper. It may be argued, however, that metallography as a whole has lost substantially in this divorce from physical chemistry, and it is for this, amongst other reasons, that we welcome so whole-heartedly the magnificent text-book of metallography recently published by Professor Masing,[4] whom the Institute is honouring to-day.

Passing on now to the metallographic effects of cold work, the general outlines of the mechanism of plastic deformation, both by twinning and slipping, were common knowledge. So far as the former was concerned, and despite the fact that the day of X-ray investigation had not yet dawned, it was *known*, as a result, for instance, of the work of Osmond and Cartaud, that the twinning of iron took place on the {112} planes, and the crystallography of the Neumann lamella and of the Widmanstätten pattern in meteorites had been worked out by classical crystallographic methods. It was further known that the space lattice in which α-iron crystallizes was body-centred cubic. That the mechanical properties were a function of crystalline orientation was equally clearly appreciated. That metals and their alloys deform, in general, by a process of slip had been proved conclusively by Ewing and Rosenhain. This work was of fundamental importance not only in demonstrating the mode of deformation, but also in providing some of the most substantial evidence in support of the view that the grains in a metallic aggregate are truly crystalline. Although the process as they envisaged it is now known to be over-simplified, it was not long before the realization came that on the planes of slip fragmentation of the crystal must occur. Further, although the belief was unpublished, there were those who envisaged the slip not as a single, more or less catastrophic fault, but as a series of smaller movements extending only over atomic distances but on a correspondingly large number of planes.

No account of the state of metallographic knowledge in 1908 regarding the deformation of metallic crystals would be complete without some mention of the work of Beilby. As a result of simple, but ingenious experimental techniques, fortified by a keen and inquiring mind, he had shown the effect of polish on a metallic surface and demonstrated that the surface layer was non-crystalline or amorphous. Later electron-diffraction experiments probably did no more than confirm what Beilby had already demonstrated. His attempt to extend this idea to the slip planes of a metal when deformed was more hypothetical, but one aspect of this explanation, namely the temporary, so-called " mobile ", condition affords an explanation, at any rate no less convincing than some which have been subsequently proposed, of the effect of deformation on the elastic limit. That when plastic flow takes place the structure of the crystals is in a highly unstable state is beyond doubt, and it would be no surprise to me if some ingenious experimenter, sooner or later, demonstrates that Beilby was in fact right in his supposition, and that the *immediate* effect of cold work, so far from hardening the metal, is to soften it.

That most articles made from metals and alloys were a seat of internal stresses had long been known, and methods were already available for their measurement, in which field it would appear that Kalakoutsky may claim priority. The originator of the suggestion that the mechanical properties of metals could be improved by pre-stressing is uncertain, but the application of the process to gun tubes was suggested by Jacob in 1907, and probably to him the name "autofrettage" is due. Clearly, then, not only were the bad effects of internal stress only too well appreciated, but also the fact that these might, in suitable cases, be turned to good use, was not novel.

That such internal stresses were unstable even at ordinary temperatures was a matter of common knowledge. The "weathering" of castings before machining was frequently resorted to to minimize distortion during the machining operation, and I remember well a Sheffield firm that "weathered" their nickel-silver ingot for 12 months before cold-rolling, in the belief that the treatment diminished edge-cracking and similar difficulties.

The absorption of energy by a metal due to cold working had been inferred by Osmond and Werth from their calorimetric studies, in which they found that cold-forged steel when dissolved in copper ammonium chloride solution gives out more heat than does the same material in the annealed state. In the case of cold-worked copper, on the other hand, they were unable to find a corresponding difference; this is not surprising in view of the fact that the measured rise of temperature was only about 0·1° C., whilst their experimental error was roughly one-third of this. In discussing these results, Howe says: "Such minute differences in heat evolution are less naturally ascribed to allotropism (as Osmond and Werth had suggested) than to unnoted differences in condition." "The shattering, stretching, or crumpling of resilient crystals, the creation of stress, macro- or micro-, are possible causes." So wrote Howe in 1891, and these words are even more justified in the light of work which has been done on cold-worked structures since then. O'Neill had shown that cold hammering lowered the density of copper, as far back as 1861, and that a similar effect was to be observed with other materials was already known. "In short", says Howe, "the probabilities seem strongly against Osmond's theory [of allotropy], and in favour of the belief that cold work produces a special kind of change . . . roughly alike in the different malleable metals. What the nature of this change is I will not attempt to say, beyond surmising that it is essentially physical." Howe then dismisses the idea that the cold-worked material differs from the annealed only in the presence of stress, believing, as has already been mentioned, that some definite physical change has been produced in the crystals themselves. He was aware of the phenomenon which we should now call strain-ageing, and says that gentle heating, which should relieve the stress, may actually intensify the effect of cold working.

Although the term "preferred orientation" was not introduced until much later, the effect was already well known. The first published account of the phenomenon appears to have been due to Stead, who showed that after certain types of mechanical and thermal treatment, the crystals in soft iron assumed a specific orientation such that the cube faces lay in the surface and at 45° to the direction of rolling. He published, in diagrammatic form, a typical example of a uniform crystalline orientation associated with a definite form of brittleness, still known as "Stead's brittleness". In his own words "just as light produces a latent invisible picture on the photographic plate, so the act of rolling steel sets up a latent disposition to crystallize in definite directions". Surely as clear a definition of preferred orientation as, in the absence of that term, could be expected.

I am assured by Dr. Maurice Cook that the effect must also have been known in the non-ferrous alloys ever since cups were first made by pressing (by the ears which it causes), although it was not associated until many years later with a definite crystalline structure.

The excessive crystal growth which results from small amounts of strain followed by an appropriate annealing treatment had also been observed by Stead in the case of iron. It is almost certain that a similar effect must have been equally well known, by those in the industry, to occur with non-ferrous metals and alloys; and single metallic crystals had been cleaved from such coarsely crystalline samples and their properties examined. It was left to Carpenter and Elam, however, to devise, on the basis of such knowledge, their strain-anneal technique for the routine production of single crystals.

I may perhaps be pardoned for mentioning the property now known as damping capacity, a tool increasingly useful in metallographic research. Although that designation was not used until the nineteen-thirties, the ability of a solid to dissipate mechanical vibrational energy had been appreciated for a long time. Weber, in 1837, described work on the elasticity of silk fibres, in which he stated that the decay of amplitude of a vibrating system is not due entirely to air friction, but is partly the result of mechanisms operating within the elastic parts of the system. He further stated that a relationship must exist between the logarithmic decrement and the elastic after-effect which is observed, for example, in galvanometer suspension wires.

In 1865 Lord Kelvin published results of the measurement of the viscosity of solid metals, derived from the fall of amplitude of a system in torsional vibration; he noted the high rate of decay for zinc, compared, for example, with that for copper, silver, and aluminium. Previous analysis of the thermo-elastic effect allowed him to state from fundamental considerations that here was a mechanism which must of necessity, under certain conditions, endow even a perfect isotropic crystal with what we now call damping capacity. Another important conclusion drawn from his experiments was that the variation of damping with frequency differed from that to be expected if the damping forces obeyed the same law as that which holds for a classical viscous fluid.

There is one field in which factual knowledge was almost as complete fifty years ago as to-day. This is the inter-relation between magnetic characteristics and mechanical deformation. The remarkable, and often extremely complex effects of both elastic and plastic deformation on magnetic properties had been examined by a large number of workers. As an example may be cited the transitory circular magnetization produced when a longitudinally magnetized wire is suddenly twisted, or when a longitudinal magnetizing force is suddenly applied to a rod in a state of torsion, and it is of interest that it was in connection with these tests, in which the magnetization lagged behind the twist, that the word " hysteresis " seems first to have been used. This is connected with the Wiedemann effect, and has subsequently been re-examined and related to damping characteristics. There is here an almost inexhaustible mine of information which still awaits detailed examination in the light of modern magnetic and deformational theory. During the last forty years or so, explanations based on the " domain " theory of magnetism have clarified considerably our understanding of these phenomena, but much still waits to be done.

Such, then, was the broad state of metallographic knowledge when this Institute was formed. The foundations were built on rock and have withstood the storms of time. Is the superstructure equally sound?

Whole sections of the science, i.e. the chemical aspects, have perforce been left out, and many names of illustrious workers are omitted.

Well may it be said of such as these:

> " Let us now praise famous men, and our fathers
> that begat us. . . .
> " All these were honoured in their generations, and
> were the glory of their times."

Have I overpainted the picture? Well, read the first edition of Dr. Desch's " Metallography ", published early in 1910, and I think, like the Queen of Sheba, you will agree " Lo! The half was not told me! "

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THE KINETICS OF THE EUTECTOID TRANSFORMATION IN ZINC–ALUMINIUM ALLOYS *

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SYNOPSIS

Time/temperature/transformation diagrams have been constructed for a zinc alloy containing 22·5% aluminium from data obtained in dilatometric, metallographic, and hardness studies. The methods agree as to the times for the start of transformation, yielding a C-shaped curve, with the transformation taking place most rapidly at 150° C. Discrepancies occur, however, in the times for the completion of the transformation; these are discussed and related, on X-ray evidence, to the slow change in composition of the α constituent in the decomposition product.

Age-hardening studies on isothermally transformed specimens are included to explain peculiarities in the hardness results observed at the nose of the C curve. A four-stage mechanism is postulated for the transformation at low temperatures. The reaction rate becomes controlled by diffusion at a lower temperature than is expected from the theoretical "spinodal" line calculated by Borelius and Larsson (*Arkiv Mat., Astron. Fysik*, 1948, [A], 35, (13)).

I.—INTRODUCTION

Although eutectoid transformations occur in a number of binary alloy systems, only the decomposition of austenite in iron–carbon alloys has been extensively investigated. The most effective method of research has proved to be that used by Davenport and Bain,[1] of allowing austenite to transform isothermally at sub-critical temperatures. Besides yielding information on the morphology of the transformation product, this technique enables the course of the transformation to be recorded quantitatively in the form of a time/temperature/transformation (*TTT*) diagram which, for plain carbon steels, has a characteristic C shape. Until recently, structurally analogous transformations in other systems had been neglected, with the notable exception of Smith and Lindlief's research into the decomposition of the β phase in the copper–aluminium system.[2] However, since 1948 the method has been applied to eutectoid transformations in the copper–aluminium,[3, 4] iron–nitrogen,[5] copper–silicon,[6] and copper–beryllium [7] systems.

In the zinc–aluminium system the eutectoid transformation involves the decomposition of the α′ face-centred cubic phase containing 78% zinc into the face-centred cubic, aluminium-rich α solid solution and the close-packed hexagonal, zinc-rich β solid solution.[8] Although this offers the advantage that the precipitated phases should be primary solid solutions, no comprehensive study using the isothermal-transformation technique has been reported. This may be due to the experimental difficulties arising from the speed of decomposition of the quenched alloy at room temperature.

The rapid decomposition of quenched zinc–aluminium alloys containing between 40 and 80% zinc was first investigated by Hanson and Gayler.[9] Alloys so treated spontaneously evolved sufficient heat to raise the temperature of the specimen approximately 50° C. within a few minutes. The progress of transformation in a 77% zinc alloy at 0° C. was observed microscopically by quenching previously polished specimens in iced water and etching them after fixed intervals. Decomposition was reported to proceed by the appearance and growth of fine dark fragments within the grains of the α′ phase, the conversion being complete within 10 min. of quenching. Age-hardening studies showed that the hardness rose to a maximum within a few minutes and then decreased rapidly, reaching a low constant value after 3 hr.

Later investigations showed that the transformation is accompanied by a contraction in volume of 0·28%[10-12] and that the electrical resistivity varies in a similar way to the hardness.[13] However, Bugakov [14] reported that the alloy continues to soften slightly over a period of months after the change in resistivity is complete. The period of induction preceding rapid transformation at 0° C. may be prolonged by raising the quenching temperature;[12] alloying elements, notably copper and magnesium, also retard the transformation.[15, 16]

X-ray studies of the decomposition of the quenched alloy at room temperature have been possible only when the reaction was delayed by the addition of a small amount of magnesium. The face-centred cubic α′ lines in the spectra gradually fade on standing, as the hexagonal β lines become visible.[17] The lines of the aluminium-rich α phase appear only later, possibly owing to the lower reflecting power of the aluminium lattice.[18] There is no evidence of any intermediate structure being formed during the

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transformation,[16] the final products being the α and β primary solid solutions in equilibrium concentrations.[19]

Recently, Borelius and Larsson[20] have redetermined certain features of the zinc–aluminium equilibrium diagram, and from their revised data have estimated the points at which the second derivative of the free energy (F) with respect to the concentration (C) is zero ($d^2F/dC^2 = 0$), for the undercooled range of α–α' solid solutions. Kinetic studies on a series of alloys containing 9–82% zinc were carried out to confirm that decomposition is most rapid at these "spinodal" temperatures, the change in resistivity that occurs when the quenched alloy transforms isothermally in an oil bath being used to follow the course of decomposition. Agreement with the theoretical curve was good, but the complex shape of the resistivity/time curves at the higher zinc concentrations made the analysis uncertain.

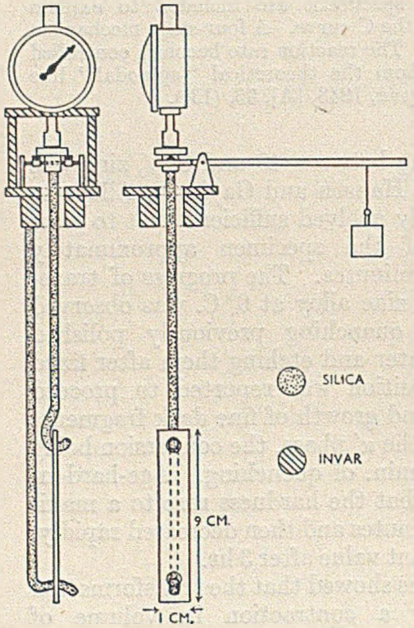

Fig. 1.—Dilatometer for Use with Thin Strip Specimens.

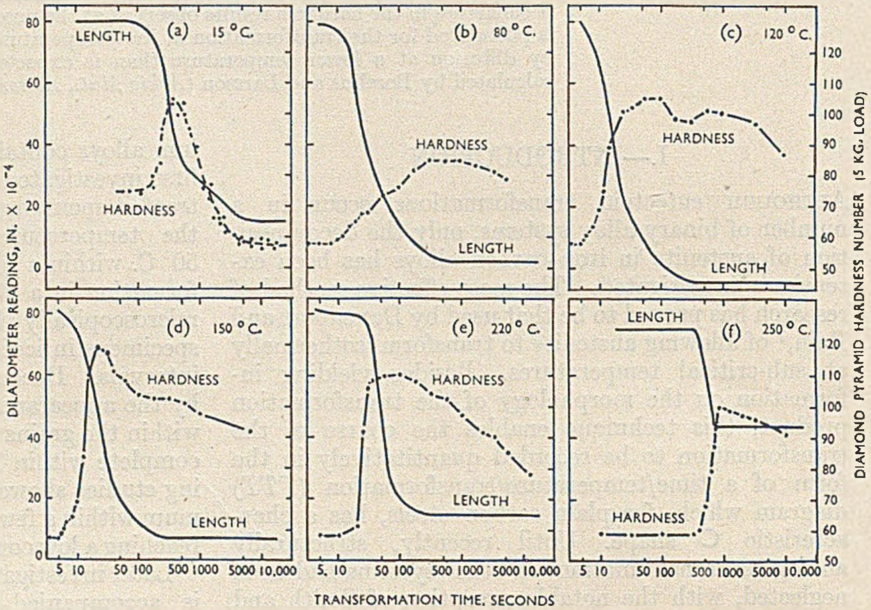

Fig. 2.—The Variation in Length and Hardness of a Zinc–22·5% Aluminium Alloy During Isothermal Transformation at Various Temperatures.

II.—EXPERIMENTAL WORK

In the experimental work to be described, the decomposition of a near-eutectoid zinc–aluminium alloy has been studied throughout the temperature range 0°–270° C., and changes in length, microstructure, hardness, and X-ray-diffraction patterns have been correlated at the various temperatures.

1. Preparation of the Alloy

The alloy was made from super-purity aluminium (99·99%) and "Crown Special" zinc (99·99%), melted in a Salamander crucible previously coated with an alumina cement wash. The 1-in.-dia. chill-cast bars were annealed for 48 hr. at 350° C. and then hot rolled into strip 0·045 in. thick after flattening by forging. Finally, the rolled strip was normalized by cooling in air after a 1-hr. treatment at 350° C. The alloy contained 77·5% zinc and 22·5% aluminium (by difference).

2. Dilatometer Experiments

A quenching-type dilatometer was designed to take a strip test-piece 9 cm. long × 1 cm. wide × 0·113 cm. (0·045 in.) thick. This was maintained in tension between hooked silica rods by means of a weighted lever, as shown in Fig. 1. The soaking furnace and the isothermal-transformation liquid bath stood side by side so that the dilatometer could be quickly transferred to the liquid. The specimens were heated for 30 min. at 375° C. before transformation, and these conditions were rigidly maintained in all tests, since a variation in either time or temperature has been shown to alter the speed of transformation in similar investigations.[3, 14] The transformation bath was thermostatically controlled to within ±1° C. and the liquid mechanically stirred.

The liquids used in the isothermal-transformation bath varied with the transformation temperature; they were low-vapour-pressure oil (270°–220° C.), glycerol–water mixtures (220°–100° C.), water (100°–10° C.), and iced water (0° C.). To test their quenching efficiency, trials were carried out with pure zinc specimens in the dilatometer. These showed that the specimen attained the bath temperature within 5 sec. in water, within 10 sec. in glycerol–water mixtures, and within 20 sec. in oil. The relatively inefficient quench at the higher temperatures was unavoidable because of the limited choice of suitable liquids.

Isothermal-transformation studies with the dilatometer were conducted at intervals of 20° C. throughout the temperature range 0°–270° C. Typical length/log time curves are plotted in Fig. 2. These and similar curves have been used to construct the time/temperature/transformation diagram shown in Fig. 3. In the range 80°–200° C. it proved impossible to fix accurately the time at which transformation begins; so that within these limits the lines for partial transformation are uncertain. To obtain the points shown, the length change within this range of temperature was estimated from a graph of percentage contraction against transformation temperature.

The *TTT* curve has the C shape characteristic of other eutectoid transformations. The rate of transformation is low just below the equilibrium eutectoid temperature, increases to a maximum at approximately 150° C.—the " nose " of the curve—and then decreases progressively as the temperature falls. Below 240° C., the later stages of contraction are prolonged, giving rise to a marked discontinuity in the line representing 100% volume change.

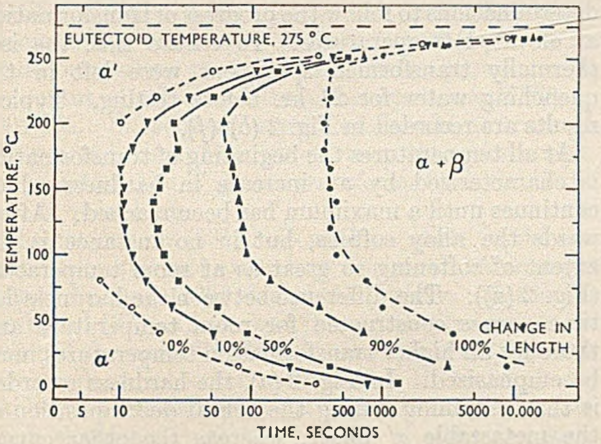

Fig. 3.—Time/Temperature/Transformation (*TTT*) Diagram for Zinc–22·5% Aluminium Alloy, Constructed from Dilatometer Curves.

3. Microstructure

For the metallographic studies small specimens of the strip, 0·5 in. long × 0·25 in. wide, were isothermally transformed in the liquid baths for appropriate time intervals. Transformation was arrested by water-quenching to room temperature, the specimens being left in the quenching water for 24 hr. before sectioning and preparation. This ensured the complete transformation of any residual α′ solid solution without the specimen being affected by the accompanying evolution of heat.

The microstructures of typical specimens are shown in Figs. 4–7 (Plate LXII), the high-temperature decomposition product being light in colour, whereas the room-temperature product, corresponding to the untransformed α′, is dark and granular. The mechanism of transformation at these higher temperatures closely resembles the formation of pearlite. At 265° C. transformation begins at a few scattered points which develop as typical pearlite-group nodules[21] encompassing parent α′ grains (Fig. 4). The newly formed product consists of regular, parallel lamellæ (Fig. 10, Plate LXII) but on prolonged treatment at this temperature the structure coarsens and takes on the appearance of irregular whorls. This process takes place slowly and is not complete (i.e. relatively straight, parallel lamellæ are still present in some areas) after 2 weeks at temperature. At lower temperatures the interlamellar spacing of the initial product decreases (Fig. 11, Plate LXII), the subsequent coarsening and whorling of the structure still being observed. It is not possible to resolve the lamellæ in decomposition products formed below 200° C.

As the transformation temperature falls, the number of centres of growth increases. Nucleation occurs chiefly at the parent α′ grain boundaries, but some centres of growth are distributed at random within the grains. The surface of the specimens also nucleates transformation. This can be seen in Fig. 6 (Plate LXII) at the junction between the specimens held, respectively, for 20 and 25 sec. at 220° C. In partially transformed specimens at 180° C. (Fig. 7, Plate LXII), it is difficult to identify the individual centres of growth along the grain boundaries, and the extent of transformation varies from one boundary to another. At 150° C. (Fig. 8, Plate LXII), the contrast between the etching characteristics of the high-temperature and room-temperature transformation products becomes so poor that it is impossible to judge the progress of the decomposition with accuracy, particularly when the reaction is nearing completion.

To overcome this difficulty, the progress of the transformation at lower temperatures was observed in specimens polished before heat-treatment and etched soon after water-quenching. A light repolish was found necessary before etching, but both operations could be completed within 30 sec. This is considerably shorter than the 80 seconds' induction period shown by the dilatation experiments to precede decomposition of α′ at room temperature. In this way, the transformation product is obtained in a matrix of the untransformed α′ grains. In the captions to the photomicrographs (Plate LXII) the two metallographic procedures are referred to as Method I and Method II, respectively. The clarity with which the course of the transformation can be followed by using the improved technique may be gauged by comparing Figs. 8 and 9 (Plate LXII), which are both of specimens held for 10 sec. at 150° C. The disadvantage of Method II is that the field examined is not truly representative, since the surface nucleates transformation. Stronger etching reagents were necessary to give short etching times, and the dark, mottled appearance of the α′ grains in Fig. 9 is due to the severe attack of the 5% caustic soda solution used for etching. An aqueous nitric acid solution gives a less severe attack and a colour reversal of the constituents similar to that reported by Greaves and Wrighton for zinc alloys.[22]

The morphology of the transformation product does not materially alter below the nose of the *TTT* curve. The α' grains are primarily consumed by the development of isolated spherulites in the grains, although transformation starts at the grain boundaries. This can be seen in Figs. 12–15 (Plate LXII), representing various stages of decomposition at 80° C. In Fig. 15, when transformation is just complete, the position of the α' grain boundaries can still be discerned. However, on repolishing and etching a day later, the boundaries were no longer visible.

For studies at room temperature and 0° C., the time when etching was complete was taken as the transformation time. The precipitates formed were similar to those shown for 80° C., although the centres of growth distributed at random within the α' grains were more numerous. The structure of the decomposition product formed at room temperature could not be resolved with an oil-immersion objective, but it appeared to be granular rather than lamellar. On re-examination some months later, a considerable coarsening of the structure had occurred (Fig. 16, Plate LXIII). The disappearance of all traces of the original α' grain boundaries from the newly formed transformation product, noted at 80° C., was also observed at room temperature.

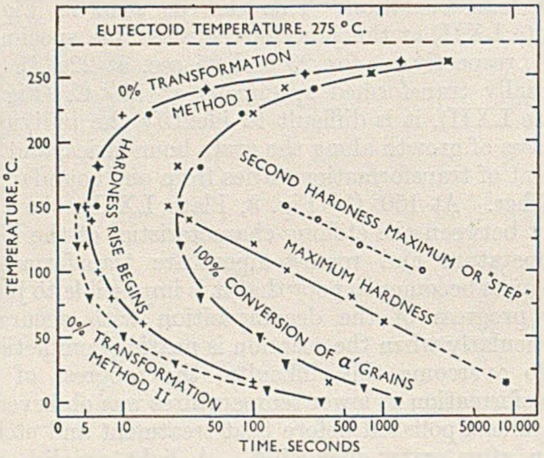

FIG. 18.—Time/Temperature/Transformation Diagram Constructed from Results of Metallographic and Hardness Tests.

KEY.

● Microstructure (Method I).
▼ " (" II).
+ Hardness change begins.
× Maximum hardness.
○ Second hardness maximum or "step".
✱ Hardness change complete at room temperature.

The times for the beginning and end of transformation obtained from these metallographic studies have been plotted in the form of a *TTT* curve in Fig. 18. At 150° C., where slightly different times for the beginning of transformation were obtained with the different experimental techniques, both points are shown. For this reason the curves corresponding to 0% transformation, as determined by the two experimental methods, have not been joined up.

4. HARDNESS

The hardness changes during room-temperature decomposition reported by previous workers were confirmed by a series of determinations on a specimen quenched from 375° C. The test-piece was left on the anvil of the hardness machine during transformation, the mass of steel preventing an excessive temperature rise. The variation in hardness is shown in Fig. 2 (*a*) (p. 408) for transformation at 15° C. The hardness rises rapidly to a maximum value 400 sec. after quenching, then falls at a decreasing rate, and finally becomes constant after approximately 90 min. This value was unchanged a week later, but over a period of months there was a further slight decrease of 4 hardness numbers. These results indicate that sufficient time must be allowed for the residual, untransformed α' to decompose, before making hardness determinations to follow the progress of transformation at elevated temperatures. To ensure this, the isothermally transformed specimens were left in the quenching water for 24 hr. before testing. Typical results are recorded in Fig. 2 (*b*)–(*f*).

At all temperatures the beginning of transformation is characterized by an increase in hardness which continues until a maximum has been reached. Afterwards the alloy softens, but in no instance is the extent of softening so great as at room temperature (Fig. 2 (*a*)). The difference between the hardness/log time curves constructed for room temperature and those for all higher transformation temperatures must be emphasized. In Fig. 2 (*a*), the hardness recorded is that obtaining during the actual decomposition of the metastable α' phase, whereas the other curves show the hardness of mixtures of the high- and room-temperature decomposition products. Hardness/time curves of the latter type could not be plotted below 60° C. because of the small difference in hardness between the two transformation products.

Two maxima appear on the curve for transformation at 120° C. (Fig. 2 (*c*)), and a similar effect was obtained at 100° C. Evidence of this secondary rise in hardness is also present at 150° C. (Fig. 2 (*d*)), where a retardation in the rate of softening occurs after 100 sec. A rather larger "step" was observed at 140° C., developing into the second maximum at 120° C. At this temperature, the hardness value obtained during the second rise is slightly lower than the first, but at 100° C., the reverse is true. No secondary rise could be detected at 60° or 80° C.

The times at which the hardness change begins and those corresponding to maximum hardness have been included on the metallographic *TTT* curve (Fig. 18). Below 150° C., where there is a marked discontinuity, the maximum hardness points fall away linearly on the log time scale. The point for room temperature (15° C.) transformation, lying on the dotted extrapolation of this line, represents the time at which softening ceased. The maximum hardness observed during room-temperature decomposition occurs much earlier and is shown as an isolated point.

5. X-Ray Examination

By using fast film, unfiltered cobalt radiation, and a specimen–film distance of 3 cm., it was possible to obtain satisfactory back-reflection photographs with exposure times of less than 15 min., operating at 50 kV. and 20 m.amp. Under these conditions a series of photographs was taken during the decomposition at room temperature (18° C.) of the water-quenched alloy. Cobalt radiation was chosen because its K_α components give the (331) α (aluminium) spectra at an approximate Bragg angle of 74°, well clear of the nearest β-phase (zinc) rings, the (203) and (105), which occur at approximate Bragg angles of 71° and 80°, respectively. The first exposure, made between 2 and 15 min. after quenching, is reproduced in Fig. 17 (*a*) (Plate LXIII). This shows a spotted (331) ring of the comparatively coarse-grained, parent α' solid solution and also weak, diffuse but continuous rings corresponding to the fine-grained α and β phases in the decomposition product. The lattice parameter of the parent α' calculated from this spectra is $3{\cdot}986 \pm 0{\cdot}001$ kX. Subsequent exposures showed no trace of this spotted ring, thus supporting the microscopic evidence that the α' grains transformed within 11–12 min. of quenching.

The continuous (331) halo of the α phase in the decomposition product extends over a much wider angle than the diffuse (203) ring of the β phase. Although this broad (331) α halo was present in the second exposure of the series (19–35 min. after quenching), it had intensified at its outer extremity. In later photographs this intensified region became a diffuse ring which sharpened with the passage of time, but at a lower rate than the corresponding β spectra. There was also a further slight increase in the diameter of this α ring. No equivalent shifts in the position of the β rings could be detected during ageing. The lattice-parameter changes corresponding to the increase in (331) α ring diameter are given in Table I, together with the equivalent percentages of zinc in solid solution, taken from Ellwood's lattice parameter/composition data.[23] The gradual sharpening of the spectra is also noted in the table, a process which is only finally complete some months after quenching (Fig. 17 (*b*), Plate LXIII).

The extent of the (331) α halo in the first exposure can be gauged by comparing Fig. 17 (*a*) and (*b*). The broad weak halo extends from a diameter just greater than that of the spotted (331) α' rings to the final position of the sharp (331) α rings, attained 3 months later. From this it is concluded that the face-centred cubic constituent in the newly formed decomposition product is of intermediate composition, the equilibrium composition being slowly approached over a period of not more than 4 hr.

Precision lattice-parameter measurements of the final decomposition products were obtained by means of a 19-cm. powder camera, the filings being taken from the specimen used in the back-reflection series. The values obtained were $a = 4{\cdot}0401(7)$ kX for the α phase, and $a = 2{\cdot}6584$, $c = 4{\cdot}9387$ kX for the β phase at 18° C. On the basis of Ellwood's data for the aluminium-rich solid solution and Burkhardt's[24] values for the zinc-rich phase, these correspond to 4·25 wt.-% zinc and 0·05 wt.-% aluminium in solid solution, temperature corrections being made where necessary.

No comprehensive X-ray study of specimens partially transformed at elevated temperatures was attempted, owing to the rapidity with which the residual α' transforms on water-quenching. Back-reflection photographs were taken of specimens transformed at selected temperatures either to the completion of the volume change or to the microscopically observed conversion of the α' grains, where these times differed. In all cases the exposures were made as soon as possible after heat-treatment.

Table I.—*Lattice-Parameter Measurements of α Phase in Zn–22·5% Al Alloy Observed During Decomposition at Room Temperature* (18° *C.*).

Ageing Time	Lattice Parameters			Remarks
	α', kX	α, kX	Zinc, in Solid Solution, at.-%	
2–15 min.	3·986	4·023	23	Corresponds to mean diffraction angle of broad (331) α halo.
19–35 min.	...	4·031	12	Calculated from intensified outer edge of (331) α halo.
45–60 min.	...	4·033	10	Broad halo replaced by diffuse ring.
2–2¼ hr.	...	4·035	8	...
4–4¼ hr.	...	4·039	3	$K\alpha_1$, $K\alpha_2$ just resolvable in Zn lines.
1 day	...	4·040	2	$K\alpha_1$, $K\alpha$ just resolvable in Al lines.
7 days	...			...
2 months	...			Separation of $K\alpha$ doublets complete.

At 260° and 240° C., specimens in which the volume change had ended gave sharp spectra, although the specimens were relatively hard when photographed (87 and 103 D.P.N., respectively). Both the α and β rings were shifted to correspond to the higher percentage of alloying elements in solid solution. At 260° C. the rings exhibited regions of locally enhanced intensity, indicating the existence of a definite crystallographic relationship with the lattice of the parent α' solid solution. The effect was not observed at 240° C. nor at lower temperatures.

At 150° C., a specimen held to the completion of the microstructural transformation (40 sec.) gave very diffuse zinc and aluminium rings. However, the broad (331) α halo observed during decomposition at room temperature could not be detected. A specimen transformed to the completion of the volume change (500 sec.) resulted in sharper spectra, particularly for the zinc-rich β phase. All the specimens were rephotographed after standing for several months at room temperature. In this interval both the α and β

phases had altered in composition to the equilibrium value for room temperature, and in the case of the 150° C. specimens there was a noticeable improvement in sharpness. In this interval the hardness of the 40-sec. specimen had decreased from 113 to 88 D.P.N., whereas the 500-sec. specimen was unchanged at 101 D.P.N.

III.—ANALYSIS OF RESULTS

1. Correlation of Kinetic Data

A comparison of the *TTT* diagrams obtained by the various experimental methods shows that the times at which transformation begins are substantially identical. The metallographic and hardness methods enabled the times for the beginning of transformation at the nose of the C curve to be determined, where the dilatometric data were inconclusive. The only significant discrepancy between results occurs when the start of transformation is judged by the microstructure of a surface polished before heat-treatment (Method II). Where this modified technique was necessary, the induction period appears shorter.

Above approximately 240° C. the times for the end of transformation, as determined by the dilatometer and metallographic methods, coincide, but at lower temperatures the contraction continues long after the conversion of the α' grains is complete. The onset of this deviation is marked by a discontinuity on the 100% length-change line of the *TTT* curve (Fig. 3). This continued contraction below 240° C. cannot be correlated with the observed coarsening of the lamellæ in the product. Both grain growth and softening continue after the volume change is complete.

Above the nose of the *TTT* curve, maximum hardness is attained slightly before the microstructural changes cease. This rise in hardness is due to the greater hardness of the newly formed high-temperature decomposition product compared with that formed at room temperature. Below 150° C., the times for maximum hardness recede from those corresponding to the conversion of the α' grains until, at 60° C. approximately, they coincide with the end of the volume change. Within this temperature range the " step " or second maximum is observed on the hardness/log time isotherms.

At room temperature, where the hardness/log time curve has a different physical significance, the contraction and softening end simultaneously. Maximum hardness at this temperature again occurs slightly before the α' grains have completely transformed. It is significant that there is a definite change of slope on the contraction/log time curve (Fig. 2 (*a*)) at this instant. A subsequent, prolonged contraction equivalent to 0·09% of the original gauge-length takes place during the softening stage. The X-ray results indicate that this contraction is due to the slow approach to the equilibrium concentration of the α phase in the newly formed transformation product. The simultaneous softening is associated with the sharpening of the X-ray spectra, but neither of these processes ceases entirely when the volume change is complete. Both continue over a period of months, accompanied by the gradual coarsening of the microstructure. However, the disappearance of all traces of the α' grain boundaries indicates that a definite structural alteration takes places within the first 24 hr.

The length/time curves for 0° and 40° C. also show abrupt changes in slope at points corresponding to the complete disappearance of the α' phase. The absence of similar discontinuities at 60° C. and higher temperatures is due to a more rapid reaction in the final stages of the contraction. This is also the reason for the absence of a broad (331) α halo in the back-reflection photograph of the specimen held at 150° C. until the conversion of the α' grains is complete. Certainly the contraction continues long after this and is accompanied by a sharpening of the X-ray spectra. At the higher temperatures, e.g. 260° C., where the end of the volume change coincides with the microstructural transformation, the X-ray rings are well defined.

2. Age-Hardening Studies on Isothermally Transformed Specimens

The close coincidence of maximum hardness with the 100% transformation line obtained from the metallographic studies, both above 150° C. and during decomposition at room temperature, suggests that a similar coincidence would be found at intermediate temperatures if it were possible to measure the hardness during isothermal transformation. The observed disappearance of the first hardness maximum between 150° and 60° C. may be due to the ability of the newly formed transformation product to soften at room temperature after quenching. On the other hand, the products formed at temperatures above the nose of the *TTT* curve are relatively stable on quenching, and the hardness maxima are retained on the hardness/time curves.

This possibility was investigated by determining the hardness at intervals after water-quenching of specimens held in the isothermal liquid bath for various times. Transformation temperatures of 220° and 80° C. were employed to represent behaviour above and below the nose of the *TTT* curve, respectively. The results of these studies, plotted in Fig. 19, show that :

(*a*) Quenching before appreciable decomposition of the α' has taken place, does not alter the course of subsequent decomposition at room temperature.

(*b*) For temperatures above and below the nose of the *TTT* curve, the newly formed transformation product is very hard on quenching. Whereas the 220° C. product undergoes a further slight increase of hardness on standing, the 80° C. specimen softens rapidly, but at a lower rate and to a higher final hardness than a specimen allowed to transform completely at room temperature.

(*c*) Prolonged holding, i.e. to the completion of the volume change, results in both cases in a reduction in hardness of the alloy. However, the softening is much greater at 80° than at 220° C. This treatment has also stabilized the 80° C. specimen, which undergoes little change on standing at room temperature. The 220° C. specimen again shows a small increase of hardness after quenching.

These studies have confirmed the hypothesis advanced concerning the relative stabilities of the initial transformation products formed at temperatures above and below the nose of the *TTT* curve. The rate and extent of softening of the product formed below the nose, after quenching to room temperature, are determined by: (i) the isothermal transformation temperature, and (ii) the time the specimen is held at temperature after the conversion of the α′ grains is complete. The lower the transformation temperature, the more marked is the softening, leading to the gradual elimination of the first hardness peak. The stabilizing effect that results from prolonged holding at the isothermal bath temperature gives rise to the "step" or second maximum on the hardness/time curves in this temperature range.

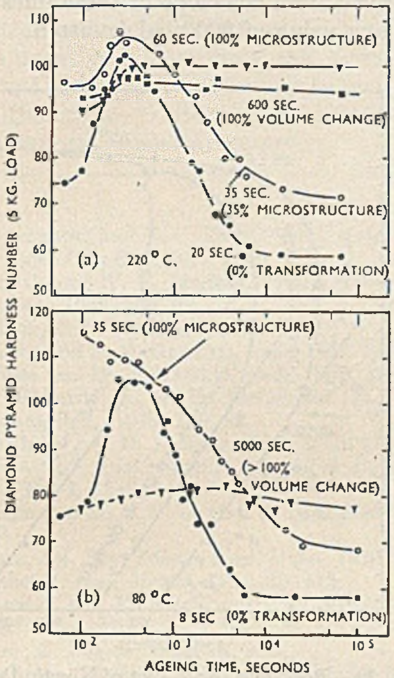

Fig. 19.—Age-Hardening at Room Temperature of Specimens Isothermally Treated at (*a*) 220° C., (*b*) 80° C. for the Times Indicated.

3. Calculation of the Theoretical Volume Change

The lattice-parameter values obtained during the X-ray investigation at room temperature (18° C.) permit the calculation of the theoretical volume change. The values used, together with the mean atomic volumes of each phase, are given in Table II. The theoretical decrease in volume obtained from these data is 0·91%, equivalent to a linear shrinkage of 0·31%. This is in good agreement with the experimental value of 0·28% reported by Fraenkel and Wachsmuth.[12]

Table II.—*Mean Atomic Volumes of Phases in the Zn–Al System at 18° C.*

Phase	Lattice Parameter, kX		Zinc Content		Mean Atomic Volume, kX^3
	a	*c*	Wt.-%	At.-%	
α (Al)	4·0402	...	4·25	1·8	16·485
α′	3·986	...	77·50	58·75	15·833
β (Zn)	2·6584	4·9387	99·95	99·85	15·113

Partial confirmation of the presence of a face-centred cubic constituent of intermediate composition in the newly formed room-temperature decomposition product, was obtained by an estimate of the volume change during transformation. This was based on the assumption that the composition of this material corresponded to a lattice parameter of 4·023 kX, the value derived from the mean diameter of the broad (331) α halo in Fig. 17 (*a*) (Plate LXIII). On this basis, the theoretical linear contraction during the conversion of the α′ grains should be 0·21%. This leaves a residual contraction of 0·10% in the final stage, compared with the observed value of 0·09%.

IV.—DISCUSSION

The eutectoid transformation in zinc–aluminium alloys has been found to be very similar to that of austenite in iron–carbon alloys, the time/temperature/transformation diagram having a simple C shape. The morphology of the transformation product is such that quantitative nucleation and growth studies similar to those of Hull, Colton, and Mehl on pearlite[25] could be readily carried out. However, it would be difficult to extend these below a transformation temperature of 180° C. without using specimens polished before heat-treatment. These would give an erroneously high value for *N*, the rate of appearance of nuclei, owing to the disturbed state of the structure at the surface.

There is little obvious change in morphology below the nose of the C curve. The low-temperature product is certainly not bainitic, but closely resembles the anomalous structures obtained by Greninger and Troiano[26] in hyper-eutectoid steels. Similar structures have been obtained in iron–nitrogen alloys by isothermal transformation below the nose of the C curve.[5]

A study of the quenched alloy at room temperature indicates that decomposition proceeds in four distinct stages:

(1) A well-defined incubation period during which no change in physical properties or microstructure is observed.

(2) The growth of the transformation product from the grain boundaries of the parent α' phase and as spherulites from random centres within the grains. This stage is accompanied by a marked rise in hardness and electrical resistivity [13] and a decrease in volume. The particle size of the phases in the product is sub-microscopic, and the diffuse nature of the X-ray spectra indicates a high degree of lattice strain. The β (zinc-rich) phase in the newly formed precipitate is of approximately equilibrium concentration, but the α (aluminium-rich) constituent is of intermediate composition.

(3) Relief of the highly localized stresses in the newly formed product by recrystallization, resulting in a marked reduction of hardness and electrical resistivity. This is accompanied by further precipitation of zinc from the non-equilibrium face-centred cubic regions. The two processes are related, for when the equilibrium concentration is reached and contraction ceases, the rapid softening also stops.

(4) A slow increase in particle size over a period of months in which the driving force for grain growth is provided by surface tension. During this time, the resolution of the doublets in the X-ray spectra is improved, and there is a further slight decrease in hardness.

The mechanism of decomposition at elevated temperatures is similar, except that stages (2) and (3) overlap to a greater extent as the transformation temperature is raised. Above 240° C. the end of the volume change coincides with the disappearance of the α' grains in the microstructure. It may be inferred that above this temperature the phases in the newly formed product are of equilibrium concentration. The greater hardness of specimens transformed to the completion of the volume change (i.e. to stage (3)), at elevated temperatures is unexpected, but may, in part, be due to further precipitation from the primary solid solutions in the eutectoid mixture after quenching. If this is the case, the hardening must occur very rapidly, for little change in hardness of either the 80° or 220° C. specimens was detected on standing at room temperature (Fig. 19). This rise of hardness does not continue indefinitely with higher transformation temperatures; the wider spacing of the lamellæ in the pearlitic product leads to reduced values above the nose of the curve.

The discontinuity on the maximum-hardness line on the *TTT* diagram (Fig. 18) arises because of the ability of the newly formed transformation product to soften on ageing at room temperature. That this softening occurs only when the transformation temperature is below 150° C., is a clear indication of some fundamental change in the nature of the product at this temperature. Similar views have been expressed for other eutectoid transformations.[5] The relative stability of the product formed at high temperatures may be attributed to a retarding effect which the higher percentages of alloying elements in solid solution exert on diffusion. However, this will not explain why the differences in behaviour set in so markedly at 150° C., the temperature of most rapid reaction. It is more probable that the size and shape of the sub-microscopic particles in the transformation product are the controlling factors.

In Fig. 20, results from the kinetic studies have been re-plotted in the form of a reciprocal-rate graph. Below 65° C. the curves representing the incubation period, the conversion of the α' grains, and the completion of the volume change, respectively, become linear and are approximately parallel. The activation energy derived from the slope of these lines gives the relatively low value of 9000 cal./g.-atom. This may occur because concentration gradients formed in the solid solution ahead of the advancing interface are largely dispersed by diffusion through the phases in the newly formed transformation product. In this respect there is some evidence [23] of the existence of vacant lattice sites in the metastable α solid solution.

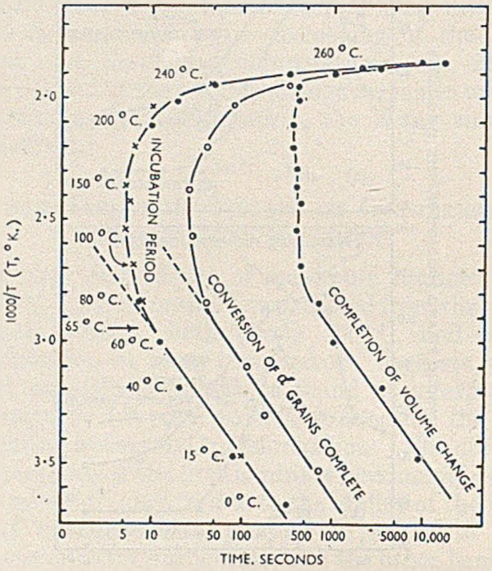

Fig. 20.—Reciprocal-Rate Plot of Kinetic Data.

Key to Experimental Methods.

● Dilation. ○ Microstructure. × Hardness.

The temperature at which the reaction appears to become diffusion-controlled (65° C.) is rather lower than the calculated "spinodal" of Borelius and Larsson,[20] where $d^2F/dC^2 = 0$ at 100° C. for a 77·5% zinc alloy. However, this is not unexpected in view of the microscopic and X-ray evidence that transformation proceeds by the growth of an $(\alpha + \beta)$ agglomerate. In the light of Fisher's analysis of the formation of pearlite from austenite,[27] the nucleation of the transformation product involves the nucleation of both the α and the β phases. Below the "spinodal", α segregates can form without an activation energy for nucleation, but this is not the case for the β phase.

Nucleation of the two-phase product is thus most likely at the grain boundaries, where the high degree of lattice strain aids the nucleation of the β. This is true for all temperatures investigated, although the number of centres of growth scattered at random within the grains increases as the transformation temperature is depressed.

The broad (331) α halo in the X-ray back-reflection photograph (Fig. 17 (*a*)), taken during decomposition at room temperature, has been attributed to the formation of α precipitates of intermediate composition. This is quite possible from free-energy considerations, where the first stage in decomposition is the formation of segregates. These can form without an activation energy for nucleation, provided that d^2F/dC^2 is negative. They may also form at higher temperatures and appear to do so up to approximately 240° C., although an addition of free energy, i.e. an activation energy, is now necessary.

V.—CONCLUSION

Although there are many similarities between the eutectoid transformation in zinc–aluminium alloys and those in other systems, there are also certain differences. The most notable is the formation of a non-equilibrium face-centred cubic constituent in the decomposition product at low temperatures. This is possible theoretically because of the continuous free-energy/composition curve for the undercooled (α–α′) range of solid solutions. The subsequent slow approach to equilibrium results in property/time isotherms of complex shape.

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The Liquid Immiscibility Region in the Aluminium–Lead–Tin System at 650°, 730°, and 800° C.* 1456

By MORGAN H. DAVIES,† B.Sc., JUNIOR MEMBER

SYNOPSIS

The analyses of successive pairs of congruent solutions, obtained by adding tin to liquid aluminium–lead or aluminium–lead–tin alloys, have been used to plot the miscibility gaps in the liquid alloys at 650°, 730°, and 800° C.

PREVIOUS work at the Fulmer Research Institute, under the sponsorship of the Tin Research Institute, has included a redetermination of the aluminium–tin phase diagram [1] and has shown that aluminium–tin alloys possess interesting potentialities as bearing materials.[2] The possibility that lead might replace part of the tin in aluminium–tin alloys has prompted a determination of the miscibility gap in the liquid alloys.

The materials used were Analar-grade lead

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(99·97+%), super-purity aluminium (99·99%), and Chempur tin (99·99%).

The alloys were melted in alumina crucibles about 10 cm. high with a total volume of 20–25 ml. For the 730° and 800° C. isotherms, aluminium and lead, or aluminium, lead, and tin, were melted in such proportions as to give approximately equal volumes of the two immiscible liquids. Tin was added and the two liquids sampled after holding times of up to 16 hr. In general, 4 hr. (with intermittent stirring with graphite or alumina rods) was sufficient to establish near-equilibrium conditions; 16-hr. holding times relate to overnight standing.

A monotectic reaction takes place in the aluminium–lead binary system at 658° C.;[3] the primary aluminium–lead–tin alloys for the 650° C. isotherm were therefore made by adding tin to the aluminium–lead liquids in equilibrium at about 675° C. The temperature was then allowed to fall to 650° C. and sampling proceeded in the usual way.

Samples were obtained from the less-dense liquid by dipping 1-mm.-bore silica quills into it, and creating a partial vacuum by means of a controlled leak to a vacuum reservoir. After the quill had been dipped into the denser liquid, argon under pressure was allowed to trickle for a short time through the tube to force out all the liquid. The pressure was then reduced, and a partial vacuum again created, so that only liquid appropriate to the depth of immersion was sampled.

A sample about 8 in. long was obtained. At the lower tin concentrations no appreciable errors were introduced by analysing only small sections of this sample, but for the higher tin concentrations the total sample was dissolved and aliquot parts were taken for the triple determination, thus avoiding errors due to segregation. Some silica contamination was encountered at 800° C., but this was reduced to a minimum by preheating the quills above the melt, thus reducing the time of immersion required to regain stable conditions. The method is similar to that used by Campbell *et al.*, in investigating the aluminium–indium–tin system.[4]

Tin was determined by titration of a chloride solution (to which $CaCO_3$ and KI were added) against chloramine-*T*. Lead was determined as PbO_2 deposited on the anode during electrolysis. The presence of copper ions in the solution prevented the deposition of lead on the cathode. Aluminium was precipitated as $Al(OH)_3$, and estimated as Al_2O_3. At high tin concentrations the error in individual determinations was estimated to be 0·5–1%. Analysis for lead and tin in aluminium-rich liquids (>75%), and for aluminium and tin in lead-rich liquids, was probably accurate to 0·1%.

Some 150 analyses were carried out, and the results were used to draw the three isothermal curves shown in Fig. 1. For the sake of clarity the tie-lines are included on the 800° C. isotherm only. The measured values of the solubilities in the aluminium–lead binary system at 730° C., and the extrapolated values at 800° C. are in fair agreement with the published work.[3]

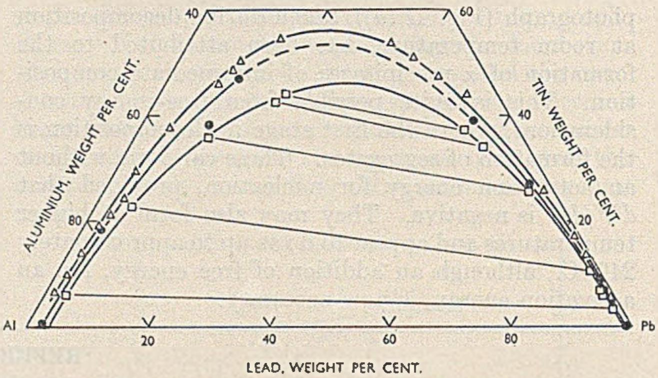

Fig. 1.—The Liquid Immiscibility Region in the Aluminium–Lead–Tin System at 650° (△), 730° (●), and 800° C. (□).

It is clear that only a small percentage of lead can be added to aluminium–tin alloys containing up to 30% tin, before the composition enters the miscibility gap. Additions of lead would therefore have to be closely limited and controlled.

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THE INFLUENCE OF THALLIUM ON THE CREEP OF LEAD*

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SYNOPSIS

Extruded specimens of high-purity alloys of lead with additions of thallium in the solid-solution range have been tested at room temperature, under constant load, after a standard annealing treatment. At a stress of 500 lb./in.^{2}, progressive additions of thallium produced first a decrease in minimum creep rate, then with more than 0·5% thallium a marked increase, and above about 8% a decrease again. Finally, at 40% thallium, the creep rate increased and corrosion and cracking occurred. Tests at lower stresses revealed similar but less marked changes. The extension on loading decreased regularly, but the complex variations in extension were present after one day's creep. The average creep rate and time to rupture showed variations similar to those of the minimum creep rate. The elongation at rupture was a maximum at any given stress for alloys containing about 8% thallium, and this elongation was greater the lower the stress; the largest extension recorded was 372%.

The variation in creep characteristics is explained in terms of the relative proportions of slip and grain-boundary movement, which depend on the grain-size, particularly in alloys containing less than 8% thallium. This has been confirmed by tests on specimens of one alloy prepared in a range of grain-sizes; the creep rate varied in a manner similar to that of a range of alloys of comparable grain-size. In alloys containing more than 8% thallium, although the grain-size continued to diminish, the reduced creep rate was apparently due to strengthening of the grain boundaries, since grain-boundary flow still appeared to be the major mechanism of deformation. It is suggested that this change in creep-resistance at about 8–10% thallium may be connected with the rate of change of lattice spacing, as Tang and Pauling (*Acta Cryst.*, 1952, 5, 39) have recently shown that only a slight decrease in lattice spacing takes place up to 10% thallium, but that it subsequently decreases more rapidly.

I.—INTRODUCTION

DURING a preliminary survey of the effect of solute atoms on the creep of lead, it was found that the progressive addition of thallium changed the creep characteristics in a complex manner, even though the maximum addition was well below the limit of solid solubility. A more detailed examination of the effect of thallium additions thus seemed desirable, both to make a general comparison of various additions of an element in solid solution and to make a systematic study of the creep of a series of alloys containing increasing amounts of an added element in solid solution. Although there are other elements which have a high solid solubility in lead, thallium was chosen because of the preliminary indications of complex behaviour, and also because it was known that the alloys could be polished and retained in the polished condition for sufficient time for microscopical observations to be made during creep. Such observations, with related X-ray back-reflection photographs for one of the lead–thallium alloys, have been published already.[1]

Previous workers have paid considerable attention to the effect on creep of specific additions of various elements to various metals, but they have been concerned mainly with comparing one addition with another and with the addition of amounts exceeding the solid solubility limit. Extensive reference to such investigations has been made by Sully [2] and by Smith.[3] Hanson and Sandford [4, 5] recognized the importance of grain-size as a factor governing creep; Greenwood and Worner [6] made standard small additions of a number of elements to lead and found some to produce marked changes in creep. These latter workers [7] and Russell [8] also showed that the influence of an element could be greatly affected by the purity of the basis lead. Phillips [9] found evidence that solid-solution alloying decreased the creep of lead, the effect of standard additions of elements being probably proportional to their maximum solid solubilities. Most of the above tests were made after standard treatments, so that variations in grain-size were present. Recently, Robinson, Tietz, and Dorn,[10] investigating the effect of solute atoms on the creep of aluminium, kept the grain-size constant, although this introduced the possibility of other changes dependent on the differing heat-treatments. Thus the present author, in uncompleted work, has found that the time of annealing at a given temperature greatly affects the creep of high-purity lead, although the grain-size is little altered.

The present paper describes experiments designed to examine the influence of a particular solute atom, thallium, on the creep of high-purity lead. The results given consist of creep data together with microscopical and X-ray observations which illustrate the variations in creep characteristics.

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II.—MATERIALS AND EXPERIMENTAL TECHNIQUE

The alloys used were made from lead containing 0·0008% total impurities and thallium containing traces of calcium and cadmium as determined spectroscopically. Full details of the experimental techniques of making the alloys and specimens, and of the testing procedure, have been given previously.[1] Briefly, the procedure consisted of vacuum casting, working and annealing to homogenize the alloys, and then extrusion at room temperature to form specimens $5 \times \frac{1}{2} \times \frac{1}{8}$ in. These were annealed for $2\frac{1}{2}$ hr. at 100° C. (except where otherwise stated) and chemically polished in a 70 : 30 mixture of acetic acid and hydrogen peroxide. Creep was induced by direct loading, axially arranged, at room temperature, namely 70° ± 2° F. (21° ± 1·5° C.). Although for short periods during some of the tests the temperature rose to 80° F. (26·7° C.) when the weather was very hot, it was not possible to detect any change in steady-state creep which could be attributed to such temperature changes. Extension was measured on a 2-in. gauge-length by a travelling microscope, reading to the nearest hundredth of a millimetre; the distance between the specimen grips was 3 in.

III.—EXPERIMENTAL RESULTS

1. CHEMICAL ANALYSIS

Chemical analyses were made by a straightforward gravimetric method for all except the two most dilute alloys, for which a colorimetric method was used.

TABLE I.—*Grain-Size and Hardness of Lead–Thallium Alloys, Extruded and Annealed for $2\frac{1}{2}$ hr. at 100° C.*

Thallium, wt.-%	Grains in Standard Area ($3\frac{1}{2} \times 2\frac{1}{2}$ in.) at × 160	Mean Grain Dia., d, mm.	Reciprocal of Grain Dia., $1/d$, cm.$^{-1}$	Diamond Pyramid Hardness No.
0	1·6	0·40	25	3·2
0·01	4·5	0·24	42	3·7
0·25	12	0·15	69	3·8
0·47	43	0·08	131	3·7
1·25	52	0·07	144	3·9
2·45	70	0·06	166	4·3
3·92	122	0·05	220	4·2
4·35	140	0·04	238	4·4
4·81	134	0·04	230	4·5
7·87	137	0·04	233	5·0
13·45	192	0·03	276	5·5
26·58	492	0·02	445	6·8
40·50	650	0·02	510	6·9

The results of these analyses are given in the first column of Table I. The alloys were well within the δ terminal solid-solution field, which extends to 87·5 at.-%* according to Tang and Pauling.[11]

2. GRAIN-SIZE AND HARDNESS

The grain-size of all the extruded and annealed alloys was measured by counting the grains on polished specimens in a number of random, standard areas. The mean grain diameters are given in Table I. The incidence of annealing twins became progressively greater with increasing thallium content; thus the number of twinned grains in the 2·45% alloy was about one in eight, whereas in the 13·45% alloy about half the grains were twinned. Table I also shows that the hardness increases fairly regularly with thallium content.

3. CONSTANT-LOAD CREEP TESTS

Annealed specimens of all the alloys and of the pure lead were loaded at initial stresses of 500 and 300 lb./in.^{2}; most of the alloys were also loaded at 200 lb./in.^{2}, and some at higher stresses. All the tests at 200 lb./in.^{2} and a few at other stresses are still in progress, but the rest were carried through to rupture. Certain numerical features of the tests are given in Tables II, III, and IV.

TABLE II.—*Creep of Lead–Thallium Alloys at 500 lb./in.^{2}.*

Thallium, wt.-%	Initial Extension, %	Extension after 1 Day, %	Extension at Rupture, %	Time to Rupture, days	Minimum Creep Rate, mm./mm./day $\times 10^{-4}$	Duration of Minimum Creep, days
0	0·93	1·55	34	26	32·5	2
0·01	0·69	1·51	50	46	27·5	10
0·25	0·45	1·35	45	151	14·7	11
0·47	0·65	1·17	60	390	7·0	85
1·25	0·47	1·30	100	220	17·1	42
2·45	0·48	1·60	208	99	48·0	25
3·92	0·28	1·60	159	48	93·0	7
4·35	0·32	1·56	180	55	86·0	9
4·81	0·27	1·57	122	68	66·0	7
7·87	0·10	1·38	228	50	100·0	5
13·45	0·08	1·15	101	89	53·0	23
26·58	0·08	0·30	67	299	7·6	50
40·50	0·03	0·30	(48) *	(88) *	(24·0)	(5)

* Specimens unloaded before fracture.

TABLE III.—*Creep of Lead–Thallium Alloys at 300 lb./in.^{2}.*

Thallium, wt.-%	Initial Extension, %	Extension at Rupture, %	Time to Rupture, days	Minimum Creep Rate, mm./mm./day $\times 10^{-5}$	Duration of Minimum Creep, days
0	0·39	(9·6) *	(717) *	14·5	190
0·01	0·25	(4·0) *	(400) *	8·0	(380) *
0·25	0·20	(4·4) *	(350) *	11·4	(330) *
0·47	0·20	(11·2) *	(724) *	15·0	(700) *
1·25	0·16	(3·1) *	(350) *	40·0	(330) *
2·45	0·12	191	538	127·0	186
3·92	0·22	273	267	290·0	38
4·35	0·02	175	188	316·0	20
4·81	0·08	205	241	230·0	30
7·87	0·10	372	221	310·0	19
13·45	0·02	325	321	120·0	48
26·58	0·00	(35·5) *	(740) *	27·6	148

* Test and minimum creep still in progress.

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* For lead–thallium alloys, wt.-% and at.-% may be interchanged without appreciable error.

Duplicate tests were made with the pure lead and several alloys; these showed, however, sufficient consistency for the properties to be assessed from single tests, except sometimes for the time and strain at rupture of specimens that recrystallized during creep. A test in which the "creep yield" curve[12] was plotted against strain on a logarithmic scale was devised. This served as a check on the self-consistency of tests at various stresses for any one alloy, because the logarithmic plot yielded a straight line, and unreliable tests were readily seen as those which did not lie on such straight lines.

Table IV.—*Creep of Lead–Thallium Alloys at* 200, 400, and 600 *lb./in.*^{2}.

Thallium, wt.-%	200 lb./in.^{2}		400 lb./in.^{2}		600 lb./in.^{2}	
	Min. Creep Rate, mm./mm./day × 10^{-6}	Duration of Min. Creep, days	Min. Creep Rate, mm./mm./day × 10^{-5}	Duration of Min. Creep, days	Min. Creep Rate, mm./mm./day × 10^{-4}	Duration of Min. Creep, days
0	46	(300) *	35·5	25	...	...
0·01	27	(336) *	34·5	29	60	2
0·25	43	(350) *	25·0	60	...	...
0·47	48	(600) *	25·0	286	14	15
1·25	150	(300) *	...	...	...	...
2·45	477	480	250	40	140	12
3·92	1100	235	...	...	...	...
4·81	850	200	400	15	...	...
7·87	895	175	...	...	...	...
13·45	682	175	...	...	...	...

* Test and minimum creep still in progress.

The addition of thallium produced marked changes in the creep of lead at 500 lb./in.^{2} stress; these can be summarized as follows:

(*a*) Up to 0·47% thallium there was a decrease in minimum creep rate and a marked increase in total life, while the total extension increased slightly.

(*b*) From 0·47 to 7·8% thallium there was a marked increase in minimum creep rate with a corresponding diminution in the duration of minimum creep and of the total life, but the total extension increased markedly.

(*c*) From 7·8 to 26·6% thallium there was a decrease in minimum creep rate, back to values comparable with those in group (*a*), and at the same time a marked increase in total life and a decrease in total extension.

(*d*) The 40·5% thallium alloy gave a curve similar to group (*b*), but no minimum creep rate, since the creep rate was found to increase continuously throughout the test. (The tests on this alloy were discontinued after 90 days because corrosion took place.)

These trends were repeated at lower stresses, although the alloys in group (*a*) showed progressively smaller changes with decreasing stress; at 200 lb./in.^{2} stress the change from group (*b*) to (*c*) occurred at a lower thallium concentration. Recrystallization during creep occurred at an increasingly high strain in tests with the alloys of group (*a*) only.

4. The Effect of Grain-Size on the Creep of 4·35% and 26·58% Thallium Alloys

Two specimens of the 26·58% thallium alloy were annealed at 250° C. instead of at 100° C., and this resulted in a larger grain-size; the creep rates of the coarse-grained specimens both at 500 and 300 lb./in.^{2} were found to be much lower than those for the fine-grained ones. Details of these tests are given in Table V.

Table V.—*Effect of Grain-Size on the Creep of the* 26·58% *Thallium Alloy.*

Annealing Temp., °C.	Mean Grain Dia., *d*, mm.	500 lb./in.^{2}			300 lb./in.^{2}		
		Extension, %	Duration of Creep, days	Min. Creep Rate, mm./mm./day × 10^{-4}	Extension, %	Duration of Creep, days	Min. Creep Rate, mm./mm./day × 10^{-5}
100	0·02	67 *	300	7·6	28	650	27·6
250	0·24	1·5	650	0·23	0·5	650	0·77

* Fractured; the other tests still in progress.

Table VI.—*Grain-Size, Hardness, and Creep at* 300 *lb./in.*^{2} *of the* 4·35% *Thallium Alloy.*

Annealing Temp., °C.	Grains in Standard Area at × 150	Reciprocal of Grain Dia., 1/*d*, cm.$^{-1}$	Diamond Pyramid Hardness No.	Initial Extension, %	Min. Creep Rate, mm./mm./day × 10^{-6}
100	140	240	4·3	0·02	316
114	96	190	4·1	0·10	141
126	63	155	3·9	0·20	56
136	43	130	3·9	0·22	30
146	29	110	3·8	0·22	18
175	12	69	3·7	0·28	6

All the tests still in progress (200 days).

Because of these striking differences, specimens of the 4·35% thallium alloy were annealed at various temperatures to provide a range of grain-sizes, and these specimens were tested at an initial stress of 300 lb./in.^{2}. Details of the tests are given in Table VI. They show a marked and regular decrease in creep rate with increasing grain-size, and the minimum creep period was well defined.

5. Metallographic and X-Ray-Diffraction Observations

Detailed examination of many specimens was carried out during creep, using both optical and X-ray-diffraction techniques. Those relevant to the changes in creep characteristics brought about by thallium additions are reported below. It is hoped to present more general observations elsewhere.

The polishing and X-ray techniques used have been described previously.[1] For optical or X-ray examination specimens were removed on a flat plate to the appropriate camera, but no systematic differences, such as those observed by Chaston,[13] were found between the creep curves for these specimens and others which were continuously under load. The alloys are considered in the groups used in the previous subsection.

(a) *Alloys Containing Up to* 0·5% *Thallium*

For pure lead and alloys containing less than 0·5% thallium, slip appeared progressively in all grains at the higher stresses, although there was evidence of grain-boundary flow, for boundaries broadened and migrated, as shown in Fig. 1 (Plate LXIV) on a specimen after 4·5% extension in 8 days. Fig. 12 (Plate LXV) shows, by the displacement of marker lines, that there was some rotation of grains at this stage. As the thallium content increased, slip became less complex and less marked. At low stresses the slip traces were fewer and more widely spaced; boundary migration was considerable, as Fig. 2 (Plate LXIV) shows, on a specimen of pure lead after 5% extension in 270 days. Recrystallization during creep occurred in the pure lead at 3–5% extension and at progressively higher extensions in the alloys; the new grains were large and usually very irregular in shape. Coarse slip appeared in the new grains, crossing the old grain boundaries; there was some migration at the new boundaries.

The X-ray back-reflection observations showed parallel changes; before recrystallization during creep, arcs occurred along the Debye circles of the initially sharp reflections at all stresses. Fig. 14 (Plate LXV), which is an X-ray back-reflection photograph of pure lead, illustrates this. After recrystallization, new large reflections were observed, and these subsequently became arcs and often tended to show a spotty sub-structure, particularly at the lower stresses. This type of reflection, which has been observed previously by Hirst [14] in large and single crystals of lead, will be considered in detail in a future paper.

(b) *Alloys Containing* 0·5–8% *Thallium*

With thallium additions between 0·5 and 8%, the tendency for the amount of slip to decrease continued. Thus, the only slip seen with normal illumination in the 7·87% alloy stressed at 200 lb./in.^{2} consisted of isolated groups of widely spaced traces. There was evidence, however, of a significant amount of very fine slip in this and similar specimens, but since this did not appear to vary markedly with thallium concentration, it will not be considered further in the present paper. The changes for this group of alloys are illustrated by Figs. 3–6 (Plate LXIV). In Fig. 3, which shows the microstructure of a specimen of the 1·25% thallium alloy after 10% extension in 50 days, the amount and type of slip is in contrast to that shown in Fig. 1. In Fig. 4, which is a multiple-beam interferogram corresponding to Fig. 3, the tilting of the grains and the nature of the slip traces are clearly shown by the fringe pattern. A similar extension was attained by the 7·87% thallium alloy in 11 days at the same stress, and the structure at this stage is shown in Fig. 5; multiple-beam fringes denoted a greater amount of tilting of the grains than that shown in Fig. 4. At the lowest stress of 200 lb./in.^{2}, marked slip traces were infrequent, especially with alloys containing 2·45% or more thallium. Fig. 6 shows an area typical of this, on the 2·45% alloy after 10% extension in 190 days.

Despite the comparative absence of slip, arcs occurred in the X-ray reflections to some extent with all specimens of this group, even during the first 10% extension; for example, Fig. 15 (Plate LXV) shows the X-ray pattern corresponding to Fig. 6. However, even at large extensions of over 100%, complete arcing to form a circle was not found at the lower stresses, as can be seen in Fig. 16 (Plate LXV), which shows the X-ray pattern from the 4·81% thallium alloy after 169% extension in 930 days. No subdivision of the reflections appeared to take place, and no new sharp reflections were observed at any stage of these experiments. The absence of recrystallization during creep for specimens of this group was indicated by the absence of new sharp reflections, of slip traces crossing the original boundaries, and of the characteristic acceleration of creep rate.

When specimens which, from these criteria, had not recrystallized during creep were repolished after fracture—some of them after considerable extensions—the grain-sizes were found to be essentially the same as before creep. There were detailed differences of structure,[15] however, which may be significant in the general elucidation of the mode of deformation in lead and its alloys.

It was found that the changes in microstructure obtained with alloys containing progressive additions of thallium up to about 8% were very similar to those associated with a decrease in strain rate for the 2·45% thallium alloy previously reported,[1] and to those of the specimens of the 4·35% thallium alloy which had been heat-treated to yield a similar range of grain-sizes.

(c) *Alloys Containing Over* 8% *Thallium*

Above 8% thallium the previous trends were continued, but a new tendency for marked broadening of the transverse boundaries was noted, particularly at lower stresses. Fig. 7 (Plate LXIV) shows the microstructure of the 13·45% thallium specimen after 10% extension in 24 days, stressed at 500 lb./in.^{2}; this should be compared with those shown in Figs. 1, 3, and 5. Fig. 8 (Plate LXIV) is the interferogram corresponding to Fig. 7; it illustrates the general absence of slip and the broadened, transverse boundaries which have formed with relatively little tilting of the grains (cf. Fig. 4). At lower stresses and higher extensions, the broadening of transverse boundaries was more evident, as Fig. 9 (Plate LXV) shows on a specimen which had extended 45% in 475 days. The absence of slip and surface rumpling at this stage was clearly shown by the phase-contrast microscope. Fig. 10 (Plate LXV) shows part of a large grain with negative phase-contrast illumination; it was necessary to select a large grain for this illustration, in order to adjust the illumination correctly for phase-contrast. However, since it was invariably found that large grains showed slip at an earlier stage than small grains, this served to emphasize the absence of surface changes within grains.

With the 26·58% thallium alloy, particularly at 300 lb./in.^{2}, grains or groups of grains tended to move apart or subside below the general level, with very few superficial markings. Fig. 11 (Plate LXV) shows this after 50% extension in 875 days, when the broadened transverse boundaries were very marked. This specimen had been very lightly repolished at 20% extension, when the very small amount of slip present was polished away. Despite the appearance of the transverse boundaries, no cracks were found on repolishing specimens of this group after fracture. The X-ray results confirmed that there was very little distortion within the grains below about 10% extension, for the reflections remained relatively sharp. An example is given in Fig. 17 (Plate LXV), which was obtained from the 13·45% thallium alloy at 4·5% extension in 65 days.

(d) *Alloy Containing* 40·5% *Thallium*

On specimens of the 40·5% thallium alloy superficial transverse cracks developed in association with scab-like areas of corrosion product; observation of detail was not possible because of rapid tarnishing. Since corrosion represented an obvious departure from the standard creep conditions, experiments with this alloy were discontinued, and no alloys containing higher proportions of thallium were made.

(e) *General Observations*

All the fractures were of the ductile knife-edge type, although those of the 26·58% thallium alloy showed a tendency towards the granular type of fracture. Specimens extended uniformly, necking often taking place only after considerable extension during tertiary creep, especially in alloys containing between 1·25 and 13·45% thallium. For example, a specimen of the 13·45% alloy, initially stressed at 300 lb./in.^{2}, showed a variation in width of 1 part in 25 along its length when the extension was 164% and therefore after about 100% extension during tertiary creep; it subsequently fractured after a further 160% extension, when the actual stress was many times that initially imposed upon it. The majority of fractures were located within the gauge-length, presumably because the high ductility of these alloys permitted the formation of shoulders at the grips. Fig. 13 (Plate LXV) shows the specimen of the 7·87% thallium alloy which has so far given the greatest extension before fracture; it is shown with a specimen before creep for comparison.

In general, although these observations followed trends which it was possible to correlate with the creep characteristics, they did not appear to be sufficient to predict or explain them completely.

IV.—ANALYSIS OF CREEP CURVES AND DISCUSSION

The information summarized in Tables I–IV has been used to construct a number of curves which illustrate the variation of creep characteristics with thallium concentration.

The effect of thallium concentration on the creep rate is brought out most clearly by Fig. 18, in which thallium content has been plotted against minimum creep rate at 500 lb./in.^{2}. It shows the grouping described in Section III, 3, there being turning points in the curve at about 0·5, 8, and 26% thallium, with fairly regular changes between these points.

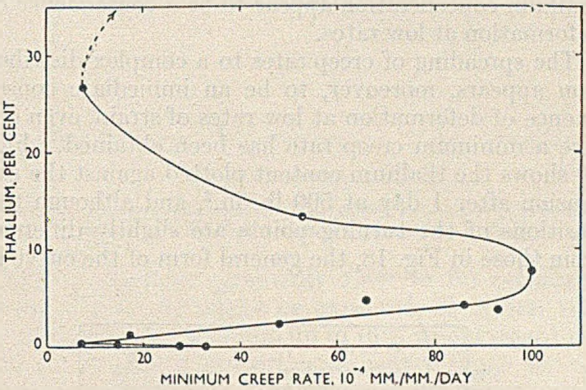

FIG. 18.—Variation of Minimum Creep Rate with Thallium Content at 500 lb./in.^{2}.

The changes at stresses of 300 and 200 lb./in.^{2} were similar to those at 500 lb./in.^{2}, but the inversion at the dilute alloy part of the curve became less pronounced as the stress decreased. It appeared from the scatter obtained that in the vicinity of 5% thallium the creep of the alloys may be very sensitive to some factor not controlled in these tests.

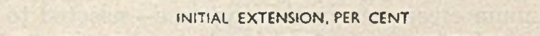

FIG. 19.—Variation of Initial Extension with Thallium Content at 500 lb./in.^{2}.

The thallium concentration which produced least resistance to creep diminished as the stress decreased.

At a stress of 400 lb./in.^{2}, tests were made with fewer of the alloys, but the concentration/minimum creep rate curves were intermediate between those obtained at 500 and 300 lb./in.^{2}

In Fig. 19 thallium content has been plotted against the extension obtained on loading at 500 lb./in.^{2}. The initial extension decreases as the thallium concentration is increased, and is the kind of variation which would be expected in the tensile extension of alloys having hardnesses as shown in Table I. At 300 lb./in.^{2} the initial extension decreased in a similar manner, but at 200 lb./in.^{2} the extensions were of the same order as the experimental error in reading them, and there was therefore considerable scatter. Thus the variations in creep rate obtained with increasing thallium concentration appear to be a consequence of deformation at low rates.

The spreading of creep rates to a complex distribution appears, moreover, to be an immediate consequence of deformation at low rates of strain, even before a minimum creep rate has been obtained. Fig. 20 shows the thallium content plotted against the extension after 1 day at 500 lb./in.^{2}, and although the positions of the turning points are slightly different from those in Fig. 18, the general form of the curve is remarkably similar. The curve is in fact, intermediate between those of Figs. 18 and 19. Similar curves have been obtained by plotting thallium content against the time to reach given small extensions.

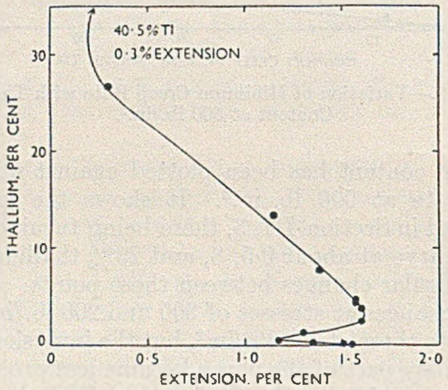

Fig. 20.—Variation of Extension After 1 Day with Thallium Content at 500 lb./in.^{2}.

In analysing creep results in the manner shown in Fig. 18, there is a danger of misrepresenting the fundamental variations in creep which have occurred because of the arbitrary choice of the characteristic—minimum creep rate in this instance—selected to represent the creep. In many of the tests in the present paper, the duration of minimum creep rate was short, especially when recrystallization during creep took place or when there was an extended tertiary stage. Thus, although it is common practice to select minimum creep rate as an index of creep behaviour, it is worth examining some other features.

Tests at 500 lb./in.^{2} have all been carried through to rupture, and Figs. 21 and 22 show the thallium content plotted against average creep rate and time to rupture, respectively. Even though some scatter may have been introduced by such factors as premature rupture, owing to the influence of the grips or other stress concentrators, these curves exhibit similar turning points and the same general character as Fig. 18.

It seems reasonable to conclude, therefore, that the minimum creep rate is a representative quantity to use in discussing the effects of additions of thallium on the creep of lead. This is supported by a further relation between minimum creep rate and time to rupture which has been plotted on a log-log scale in Fig. 23. The points fall fairly close to a pair of straight lines intersecting at about the point representing the 0·47% thallium alloy. Results for alloys containing 0–0·47% thallium are on one straight line, those for alloys containing 0·47–7·87% thallium are on another, and those for alloys containing 7·87–26·58% thallium move back along the latter. Thus there appears to be a close relationship between minimum creep rate and time to rupture, which also divides the alloys into the groups previously noted in Section III, 3. This linear relation between minimum creep rate and time to rupture has been noted by Servi and Grant [16] in tests with aluminium in which temperature and not composition was the variable; the creep curves for aluminium were such that a large proportion of the creep had been at the minimum rate, whereas in the present experiments with lead–thallium alloys this was only true for the tests represented by points near the intersection in Fig. 23.

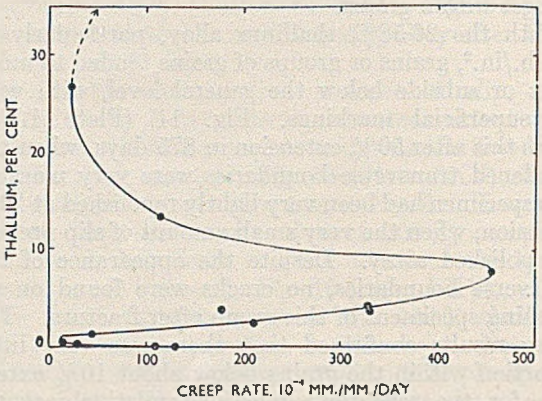

Fig. 21.—Variation of Average Creep Rate with Thallium Content at 500 lb./in.^{2}.

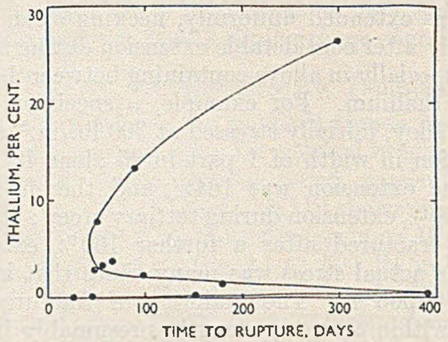

Fig. 22.—Variation of Time to Rupture with Thallium Content at 500 lb./in.^{2}.

In Fig. 24 the reciprocal grain diameter (1/*d*) has been plotted against minimum creep rate for the alloys at 500 lb./in.^{2}; such a curve will be referred to as a "1/*d*" curve in subsequent discussion. It will be seen that the general form of the curve is similar to that of Fig. 18, in which thallium content replaces 1/*d*, and there is a linear portion between the points representing the 0·47 and 7·87% thallium alloys. A similar linear portion has been found on the 1/*d* curves for stresses of 400, 300, and 200 lb./in.^{2}, between the point representing the 0·47% thallium alloy and a value of 1/*d* of about 240.

As suggested by Sully,[2] the linearity found in the 1/*d* curves, typified by Fig. 24, indicates that the variation in grain-size may be a major factor in determining creep rate, at least over a range of thallium concentration, provided that the deformation takes place by grain-boundary flow. This is demonstrated more clearly by Fig. 25, which shows part of the 1/*d* curve for the tests at 300 lb./in.^{2}, and the 1/*d* curve for the various grain-sizes of the 4·35% thallium alloy. The changes in creep rate brought about by altering the grain-size of one alloy are thus seen to be very similar indeed to those brought about by varying the grain-size by additions of thallium. It should also be noted that the 4·81% thallium alloy appears to behave anomalously, as in other diagrams, falling away from the most likely curve.

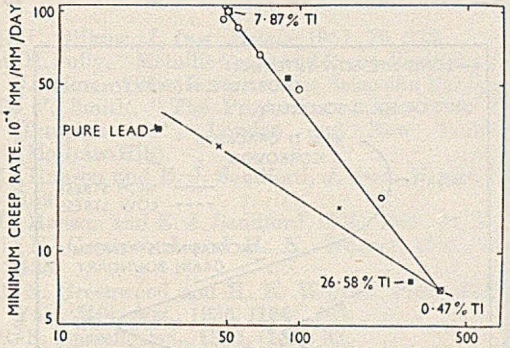

Fig. 23.—Relation Between Minimum Creep Rate and Time to Rupture at 500 lb./in.^{2}.

Key.

× 0–0·47% Tl. ○ 0·47–7·87% Tl. ■ 7·87–26·58% Tl.

Hanson[17] has suggested that the number of grains in the cross-sectional area is a fundamental way of denoting grain-size in creep relations, but using this quantity instead of 1/*d* in the present work did not give curves which differed greatly from those in Fig. 25.

There is thus evidence that the observed variations in alloys containing up to 7·8% at 500 lb./in.^{2} and alloys containing up to about 5% at 200 lb./in.^{2} are mainly the result of grain refinement. This is in agreement with the results of Tang and Pauling,[11] who have found that the lattice spacing of lead–thallium alloys decreases very slightly up to 10% thallium and then decreases at an increasing rate up to about 75%. From calculations of atomic radii, based on these results, they suggest that thallium replaces lead isomorphously and assumes the valency of lead (2·14) up to 10%; thereafter, the valency of dissolved thallium gradually rises to its calculated value of 2·5. It might be expected, therefore, that alloys containing thallium up to 10% would show little change in plastic properties, except through the influence of grain-size. The observed changes in creep due to the initial thallium additions up to 0·47% appear to be associated with the elimination of slip, because the changes are less marked at the lowest stress, when the amount of slip is small even in pure lead. It is possible that the proportion of slip in these coarse-grained alloys depends, in turn, on the relation of grain-size to the cross-sectional area of the specimen.

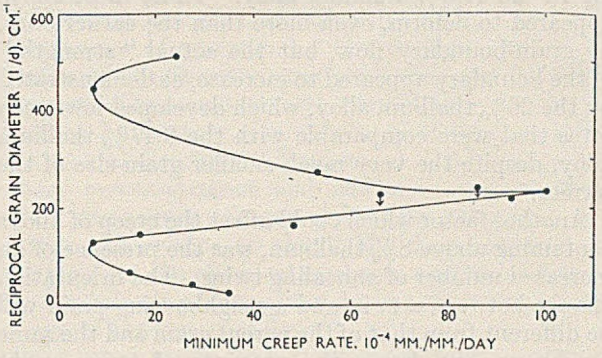

Fig. 24.—Relation Between the Reciprocal of the Grain Diameter (1/*d*) and the Minimum Creep Rate at 500 lb./in.^{2}.

The lead–thallium alloys containing up to 10% thallium thus differ from the solid-solution alloys investigated by Robinson, Tietz, and Dorn,[10] who attributed the increased resistance to creep to the "Cottrell effect" and the resistance to recovery induced by small solute additions. With the lead–thallium alloys, it might be expected that there would be very little difference in the creep rates of these alloys if tests were made on specimens of the same grain-size.

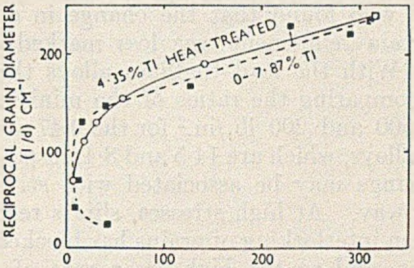

Fig. 25.—Effect of Variation of Grain-Size on Minimum Creep Rate at 300 lb./in.^{2}.

The change to a decreasing creep rate at about 8% thallium does not appear to be associated with any major property of the alloys, such as hardness or electrical resistivity,[18] and since the grain-size continues to diminish with thallium additions, some other factor must be sought. The X-ray and microscopical observations indicated a continuance of the earlier trends, although there was an increasing tendency for transverse boundaries to broaden, which led to cracking in the 40·5% thallium alloy. Thus, these alloys appeared to deform, even more than the earlier ones, by grain-boundary flow, but the actual "strength" of the boundary appeared to increase, as demonstrated by the 26% thallium alloy, which developed low creep rates that were comparable with the 0·47% thallium alloy, despite the very much smaller grain-size of the former.

Another factor which could affect the creep of alloys containing above 8% thallium, was the presence of an increased number of annealing twins. The orientation change between a twin and a neighbouring grain will be different from that of the parent grain and the same neighbour, and hence the presence of twins might inhibit boundary flow because of abrupt changes of energy in boundaries intersected by twins; alternatively, the presence of twins might prevent readjustment when interlocking at irregularities occurred in boundary flow.

The fact that recrystallization during creep occurs only in the alloys containing up to 0·47% thallium may be connected with the changes in the mechanism of deformation. It is probable that recrystallization during creep occurs after internal changes, such as slip, have increased the stored energy to a sufficiently high level, and when most of the movement is at the grain boundary the energy necessary to activate recrystallization is not built up.

It was found that there was a systematic variation with stress of the slope and position of the linear portions of the $1/d$ curves. The slope increased as the stress diminished in such a manner that a plot of the stress against the logarithm of the slope was a straight line. This was similar to an effect which has been noted by other observers, e.g. Gohn, Arnold, and Bouton,[19] who found that the change in creep rate due to grain refinement was less marked at lower stresses. With the lead–thallium alloys this can be seen by comparing the ratios of the minimum creep rates at 500 and 200 lb./in.^{2} for the 0·47 and 3·92% thallium alloys, which are 14·5 and 8·4, respectively.

The change may be associated with stress in the following way. At high stresses, slip is readily activated when interlocking of grains has blocked relative movement, and so the high creep rate of the fine-grained alloys results from the combination of increased boundary area and ability to slip. With low applied stresses, however, considerable time may be necessary to build up local stresses large enough to activate slip when boundary flow is blocked, and thus it is only the change in grain-boundary area which operates. It follows that the change in creep rate is then less than at high stresses, for the same change in grain-size. The logarithmic relation between stress and slope of the $1/d$ lines is probably an indication of the time factor involved in the above process.

Besides resulting in the relatively small change in creep rate at lower stresses, this effect may be the explanation of the high extensions at fracture of fine-grained specimens and of the increase in such extensions with lower stresses. Dollins[20] has recently found some alloys of lead to give first a decrease and then an increase in extension at rupture, as the stress is decreased, although the changes are much less marked than with the thallium alloys. He associates the original high ductility with slip and explains the other changes in a similar manner to that outlined above.

V.—SUMMARY

The investigations described have shown that the additions to lead of thallium in solid solution cause complex changes in the creep characteristics of extruded and annealed specimens of the alloys, and that these characteristics are modified as the testing stress is decreased. The diagram in Fig. 26, and points 1–6 below, attempt to summarize these; in Fig. 26, the full line is for tests at 500 lb./in.^{2}, and the dotted line for lower stresses.

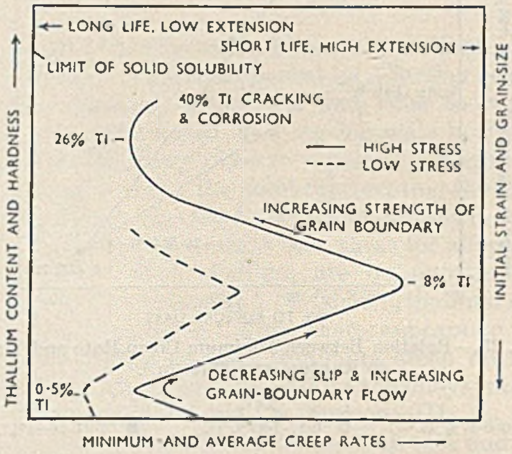

Fig. 26.—Schematic Summary, Illustrating the Effect of Thallium Concentration on the Creep of Lead.

1. The initial extension obtained on loading the alloys shows a regular decrease with additions of thallium. This confirms the observations of Robinson, Tietz, and Dorn[10] on aluminium.
2. After even the small amount of creep that takes place in one day, a complex relation between thallium content and creep extension is established.
3. A major effect of thallium additions is to decrease the grain-size obtained after a standard treatment. Consequently, the creep-resistance decreases at all stresses, and very markedly at high stresses, up to about 8% thallium; this is approximately the range in which there is very little change in lattice spacing. The creep rates and structural changes of alloys in this range at 300 lb./in.^{2} are very similar to those of specimens

of an alloy which have been heat-treated to yield a similar range of grain-sizes.

4. Above 8% thallium, the creep rate decreases markedly, despite continued grain refinement, and there is an increasing tendency to cracking, although grain-boundary flow is still predominant.

5. Thallium increases the extension at rupture of lead during creep and the maximum extension (nearly 400%) occurs at about 8% thallium; at compositions between 2 and 13% thallium there is a marked increase in the extension at rupture of any alloy with decreasing stress.

6. Grain movement and rotation have been found to replace visible slip progressively as the thallium content increases. With more than about 8% thallium there is, in addition, an increasing tendency to form superficial shallow fissures along boundaries transverse to the stress. Cracks, however, have only been observed in the 40·5% thallium alloy, which corroded rapidly.

ACKNOWLEDGEMENTS

The work described in this paper formed part of a programme of research by the Physical Metallurgy Section of The Commonwealth Scientific and Industrial Research Organization and was carried out at the Baillieu Laboratory, University of Melbourne, under the general direction of Professor J. Neill Greenwood, whose encouragement and advice in the course of many discussions are gratefully acknowledged. The author also wishes to thank Mr. J. A. Corbett for carrying out the chemical analyses, and his colleagues for the help they have given during discussions in the course of the investigations.

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Discussion

The Constitution of Tantalum–Titanium Alloys

By D. Summers-Smith

(*Journal*, this vol., p. 73.)

Mr. D. J. Maykuth * : We have recently concluded work on the constitution of titanium–tantalum alloys † and are pleased to report that our work is in substantial agreement with that of Mr. Summers-Smith on all but one of the principal features of the constitutional diagram.

We have confirmed the complete solid solubility of tantalum and β-titanium and have also shown that the solidus line for the system rises as a smooth curve, without a maximum or minimum, from the melting point of titanium to that of tantalum. The approximate melting points given by the author all fell within a 50° C. spread of the solidus curve which we established by optical observation on induction-melted alloys and by incipient-melting techniques.

We also observed that the solid solution in alloys containing 79·1 at.-% titanium or less was retained upon quenching from the β field; quenching of β alloys containing 85·0 at.-% titanium or more resulted in martensitic transformation structures. Comparable work by the author bracketed the limit of the stable β composition between 82·2 and 86·1 at.-% titanium.

Comparison of the $(\alpha + \beta)/\beta$ boundary curves from the two diagrams shows that both have essentially the same shape and slope. The curve obtained by Mr. Summers-Smith is, however, displaced to slightly higher temperatures in alloys containing above 8·1 at.-% tantalum.

The only point of major discrepancy between the two investigations is concerned with the range of solubility of tantalum in α-titanium. The author places the limit of tantalum solubility at less than 0·5 at.-% at 700° C., whereas the results of our investigation show that the solubility of tantalum increases with decreasing temperature to a maximum of about 3·64 at.-% at 550° C. Our data, based largely on metallographic work, show that, at 700° C., about 3·07 at.-% tantalum is soluble in the α-phase. In addition, some X-ray-diffraction data were obtained which show that, at 530° C., the α solubility extends beyond 1·49 at.-% tantalum. These data, given in Table A, show that increasing additions of tantalum in amounts up to 1·49 at.-% produce a linear expansion of the α-titanium lattice in the *c* direction, while having no significant effect on the *a* parameter.

Table A.—*Lattice Parameters of α-Titanium Solid-Solution Alloys Annealed at 530° C.*

Tantalum, at.-%	Lattice Parameter, Å.		c/a Ratio
	a	c	
0	2·9503 ± 0·0001	4·6834 ± 0·0001	1·587
0·26	2·9512 ± 0·0001	4·6845 ± 0·0001	1·587
1·01	2·9506 ± 0·0003	4·6858 ± 0·0002	1·588
1·49	2·9502 ± 0·0003	4·6873 ± 0·0002	1·589

We note that, in the metallographic procedures described by Mr. Summers-Smith, all the samples used to investigate the extent of the $(\alpha + \beta)$ field were initially treated by heating within the β-phase field. As he observed, such treatment of alloys containing above 86·1 at.-% titanium produces transformed β structures in which the degree of α acicularity is dependent upon the cooling rate from temperature. These transformation structures are very tenacious in that subsequent treatment below the $(\alpha + \beta)$ boundary does not modify the gross microstructure to any appreciable extent, i.e. the acicular or plate-like character of the initial α precipitate is not removed. As illustrated in Figs. 3 and 4 (Plate XI) of the paper, this offers no serious metallographic problems in phase differentiation when a relatively large volume of equilibrium β phase has been obtained in subsequent treatment. However, our experience with titanium-rich alloys shows that metallographic identification of trace quantities of β in an acicular α structure is exceedingly tenuous. For this reason, our metallographic studies on the α solubility were carried out by bringing cold-worked alloy samples into equilibrium. The combination of cold work and heat-treatment produces an equi-axed, α grain structure in which small quantities of β phase may be readily recognized.

The Author (*in reply*) : It is gratifying to see that the work at the Battelle Memorial Institute substantially confirms the diagram proposed in my paper. In an attempt to remove the discrepancy regarding the solubility of tantalum in α-titanium, four alloys containing 4·5, 2·4, 1·0, and 0·6 at.-% tantalum were cold worked to 50% reduction and separate samples of each alloy were annealed at 800° and 700° C. for 24 hr. and water-quenched. Metallographic examination showed the structures given in Table B.

Table B.—*Structure of Alloys Containing up to 4·5 At.-% Tantalum.*

Tantalum, at.-%	Structure	
	At 800° C.	At 700° C.
4·5	$\alpha + \beta$	$\alpha + \beta$
2·4	$\alpha + \beta$	$\alpha + \beta$
1·0	$\alpha + \beta$	α
0·6	α	α

As Mr. Maykuth states, the acicular decomposition product is very stable and, even after this treatment, traces of it could still be seen in the 4·5 and 2·4% tantalum alloys. To remove these and ensure that equilibrium had been obtained, the cold-working and heat-treatment cycle was repeated for all the samples. The alloys now had equi-axed grain structures, but no change in the proportions of the two constituents could be detected in the two-phase alloys. A considerable amount of β was present in the 2·4% alloy quenched from 700° C., suggesting that at this temperature the solubility limit lies closer to 1% than 2·4% tantalum.

While this shows that the solubility of tantalum in α-titanium is greater than that proposed in the paper, it seems most unlikely, as a result of the present experiments, that the solubility at 700° C. is as great as 3%, the value given by Mr. Maykuth and his co-workers; more probably it lies at about 1·5%. The present results no longer conflict with the X-ray data given by Mr. Maykuth.

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* Battelle Memorial Institute, Columbus, O., U.S.A.

† H. R. Ogden, R. I. Jaffee, and D. J. Maykuth, *J. Metals*, 1953, in the press.

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SOME METALLOGRAPHIC OBSERVATIONS ON AGED ALUMINIUM–COPPER ALLOYS *

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SYNOPSIS

The occurrence of Gayler's "light phenomenon" in aluminium–copper alloys has been studied in detail. It has been concluded that it is not formed as an integral part of the ageing process, but that it originates as a grain-boundary recrystallization process to relieve strains induced by cold-water quenching and subsequent precipitation on ageing.

The effect of very small additions of indium or tin on the ageing of aluminium–copper alloys has been investigated by optical and electron microscopy. The particles of θ' are smaller, more numerous, and occur at an earlier stage in the ageing of the ternary alloy. This supports the previous conclusion that the ternary additions facilitate the nucleation process in the ageing of aluminium–copper alloys.

I.—INTRODUCTION

SYSTEMS undergoing precipitation have frequently been studied by microscopic methods, and notable contributions have been made by Gayler [1, 2] to our knowledge of the structures of aged aluminium–copper alloys. During intensive metallographic studies of the binary aluminium–4% copper alloy, Gayler has been able to distinguish three precipitates under the microscope, viz. copper-rich regions, θ', and θ. The copper-rich regions occur before the peak on the ageing curve at 130° C. Gayler concludes that age-hardening is primarily due to the precipitation of these copper or copper-rich aggregates, together with the accompanying formation of crystallites of the aluminium solid solution stable at the temperature of ageing. These crystallites are better known as the "light phenomenon". The structure is clearly evident in the grain boundaries after ageing for 50 days at 130° C., and is said to be visible within the grains during the second rise to peak hardness (5–40 days at 130° C.).

More recent theories [3] also postulate that the ageing of binary aluminium–copper alloys involves a series of precipitates designated as G.P. (Guinier–Preston) zones [1] and [2], θ', and θ. The G.P. zones [2] are Gayler's copper-rich regions, but the "light phenomenon" does not form an essential part of the current theories of the ageing process. Thus it was thought that a wider metallographic investigation would help to determine the conditions under which the "light phenomenon" is formed.

The work described in the second part of this paper arose out of recent investigations by Hardy,[4] who has shown that small quantities—of the order of 0·05–0·1 wt.-%—of cadmium, indium, or tin, exert a pronounced influence on the response of aluminium–copper alloys to artificial ageing. The rate of artificial ageing is greatly enhanced, and the tensile properties, notably the proof stress, are considerably increased. It was thought that a metallographic comparison between the binary and ternary alloys might yield some information about the ageing mechanism involved in the latter series.

In carrying out the investigation it was found that the structures of the aged alloys were frequently finer than the limit of resolution of the optical microscope. Resort was therefore made to the technique of electron microscopy.

II.—THE "LIGHT PHENOMENON" IN AGED ALUMINIUM–COPPER ALLOYS

Most of the investigation was confined to the binary aluminium–copper alloys, though a limited comparative survey was made of the ternary series of aluminium–4% copper alloys with 0·05% indium or tin. In all cases the microspecimens were prepared from the hardness test-pieces used by Hardy [3, 4] in obtaining his hardness/ageing curves. In order that the work can be referred to, the hardness/time curves (Fig. 1) have been reproduced from previous papers; [3, 4] they show the ageing curves for aluminium–4% copper and aluminium–4% copper–0·05% indium alloys at 130° and 190° C.

1. EXPERIMENTAL METHOD

Binary aluminium–copper alloys with copper contents of 2, 3, 4, and 4·5% were examined after ageing for various times at 130°, 165°, and 190° C. Ternary alloys containing 4% copper were investigated after ageing at 130° and 190° C. Details of the

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preparation of the hardness specimens have been given earlier.[3] Briefly, it consisted of cold forging the chill-cast alloys, which were subsequently solution-treated and water-quenched before ageing.

Microsections were cut from the hardness specimens, which had been stored after ageing, and were rubbed down on a series of emery papers finishing with 000 grade. They were then electropolished for 4–8 min. in De Sy and Haemers's solution[5] consisting of 4 parts absolute alcohol and 1 part 60% perchloric acid, with a potential difference of 10–12 V. In general, etching was carried out for 10–15 sec. in a solution of 0·5 g. NaF, 1 c.c. HNO_3, 2 c.c. HCl in 97 c.c. H_2O, which is very similar to Keller's etch. This reagent gave excellent contrast between the "light phenomenon" and the matrix.

2. Metallographic Appearance

The distinctive characteristics of the "light phenomenon" made recognition simple, and, as shown in Figs. 2 and 3 (Plate LXVI), the structure possessed a straight boundary with one grain and a more clearly defined serrated boundary with the adjoining grain. The structure was in relief both in the unetched and etched condition and was thus easily distinguished from simple grain-boundary depletion of the solute. Etching caused some pitting in the region constituting the "light phenomenon", but otherwise it was unaffected. Contrary to Gayler's observation, the "light phenomenon" was seen only in the grain boundaries.

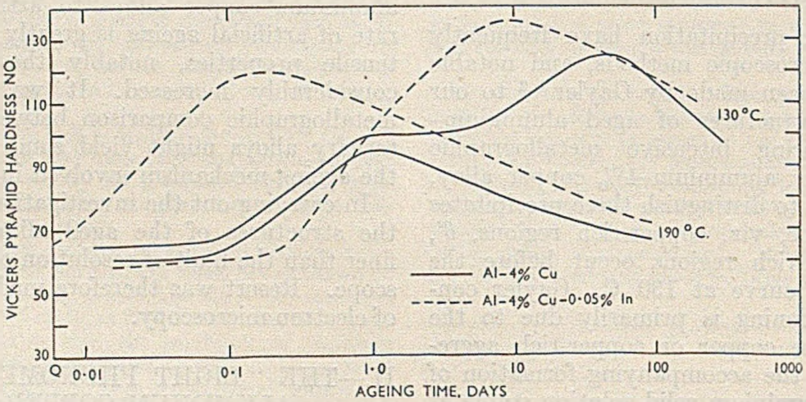

Fig. 1.—Hardness Curves for Aluminium–4% Copper and Aluminium–4% Copper–0·05% Indium Alloys Aged at 130° and 190° C. (*After Hardy.*[3, 4])

3. Occurrence of the "Light Phenomenon" in Binary Aluminium–Copper Alloys

(a) *Specimens Aged at* 130° *C.*

No traces of the structure were observed in the 2% copper alloy. However, with higher copper contents varying amounts of the "light phenomenon" were present in specimens which had been aged from several hours to 800 days or longer. Traces were found in the 3% copper alloy after 16 hours' ageing, whilst the effect was noted after 8 hr. in both the 4 and 4·5% copper alloys. This is much less than the time required to the completion of the initial rise of the hardness/time curve (see Fig. 1). Growth of the crystallites continued up to a time somewhat beyond that corresponding to the peak hardness (50 days), after which their size remained constant, while the hardness continued to decrease (examples are shown in Figs. 2 and 3 (Plate LXVI)). The number of sites did not change appreciably once the "light phenomenon" was clearly visible. The final size and, to a lesser extent, the frequency was greater in the alloys of higher copper content.

(b) *Specimens Aged at* 165° *C.*

The "light phenomenon" was again absent in the 2% copper alloy; traces or "threads" could be observed in the other alloys after short ageing times, e.g. 8 hr. in the 3% copper specimens and 4 hr. in the 4–4·5% copper alloys. The number and maximum size after peak hardness (3½ days for the 4% copper alloy) was less than that found at 130° C. As before, the distribution remained more or less constant from just beyond peak hardness up to at least one hundred days.

(c) *Specimens Aged at* 190° C.

The "light phenomenon" was not found in either the 2 or 3% copper alloys, and only minor traces were found in the alloys of higher copper content.

4. Occurrence of the "Light Phenomenon" in the Ternary Alloys

The restricted survey indicated that in general the "light phenomenon" was present to a similar extent in the ternary alloys.

5. Discussion of Results

As indicated previously, Gayler considered the "light phenomenon" to be an integral part of the ageing process of binary aluminium–copper alloys. She suggested that the flat on the ageing curve resulted from a balance between the formation of sub-microscopic copper-rich areas and the relief of strain by recrystallization to form sub-microscopic particles of the "light phenomenon". Gayler reported that the crystallites did not grow, but visible formation of the new particles continued during the rise to peak hardness and possibly for much longer ageing times.

The present observations gave no evidence of the "light phenomenon" within the grains, and the discussion will be limited to its occurrence in the grain boundaries. Perryman and Blade[6] have also observed the "light phenomenon" in the grain boundaries of aluminium–zinc alloys, and their results lend weight to the following remarks.

It has been revealed that the "light phenomenon" originates at a very early stage in the ageing process—certainly before the time corresponding to the flat on the two-stage ageing curve. Growth continues until the ageing has advanced just beyond the peak hardness, but the number of sites remains relatively constant. The "light phenomenon" also occurs at a very early stage in the ageing of the ternary alloys.

It would appear to originate as a recrystallization process, probably resulting from strains arising from the cold-water quench and precipitation on ageing. The observation that the amount of "light phenomenon" increases with increasing copper content is in accordance with this view, as the higher the solute concentration the greater is each of these two strain factors. The fact that growth ceases when over-ageing, or softening, occurs would indicate that the presence of the "light phenomenon" is closely associated with the precipitation process, since the continued precipitation of θ' and θ within the grains reduces lattice strain. The smaller amounts of the "light phenomenon" found after ageing at 165° and 190° C. may be due to relief of strain by grain-boundary flow, whereas recrystallization provides a more rapid means of relief at the lower temperature of 130° C. The smaller hardening on ageing at the higher temperature would also indicate less lattice strain than at 130° C. and account for a decreased tendency to form the "light phenomenon".

The intermediate precipitate, θ', is probably formed in the grain boundaries at an early stage in the ageing process, and, as suggested by Gayler, the "light phenomenon" may well have a composition close to that of the equilibrium solid solution at the ageing temperature. As described by Hardy,[3] the formation of a more stable phase must be accompanied by re-solution of the less-stable precipitate. Diffusion is more rapid in the grain boundaries than within the grains, and migration of the interface across a small volume of the metal will provide a path for easy diffusion and hence rapid precipitation. Particles of precipitate can also be formed at the new interface and may anchor the movement leading to the typical serrated edge.

Gayler has observed the "light phenomenon" in an air-cooled aluminium–4% copper specimen, in which the quenching strains would be considerably less. The build-up of strains here is to be associated with volume changes in the region of the grain boundaries brought about by precipitation in the boundaries during air cooling.

III.—COMPARISON OF THE STRUCTURES OF AGED ALUMINIUM–4% COPPER AND ALUMINIUM–4% COPPER–0·05% INDIUM (OR TIN) ALLOYS BY OPTICAL AND ELECTRON MICROSCOPY

The microspecimens were obtained from the hardness test-pieces of Hardy[3, 4] in a manner similar to that indicated in Section II, 1 (p. 428), but the work was limited to the binary and ternary alloys with 4% copper. Preliminary observation of the two ternary alloys showed negligible differences in structure, and so it was considered necessary to compare only one of these with the corresponding binary alloy.

1. Comparison by Optical Microscopy

A number of etching procedures were examined, but the most satisfactory one was to increase the HCl content of the etching reagent used in seeking the "light phenomenon". This was necessary because the more dilute mixture did not provide a satisfactory result on the ternary alloys. The maximum magnification attainable was 2000 diameters.

The alloys were compared after ageing at 30°, 130°, 190°, and 350° C. for varying times. With the exception of the specimens aged at 350° C., the structures were generally too fine to provide much useful information. Nevertheless, the following observations were made:

(i) Neither the binary nor ternary alloys showed evidence of visible precipitation within the grains or in the grain boundaries after ageing at 30° C. for times up to 800 days. However, the binary alloy exhibited greater preferential staining of the grains, which suggested that ageing, by segregation of copper atoms in preferred planes, was further advanced in this alloy. This would be in accordance with Hardy's findings that the rate of natural ageing was depressed by very small amounts of indium or tin.[4]

(ii) Specimens aged at 130° and 190° C. had structures which were incompletely resolved, and the only conclusion which could be drawn was that the precipitate in the ternary alloys appeared to be finer.

(iii) Any effects due to ternary additions that were shown up at lower ageing temperatures were not observed on ageing at 350° C., as the structures of both alloys were virtually identical.

2. Comparison by Electron Microscopy

Hardness measurements have shown that it is in alloys aged in the range 100°–200° C. that the ternary additions have their greatest effect (Fig. 1, p. 428). As the structures were too fine for optical comparison, it was decided to take some electron micrographs. It was hoped to compare the particle sizes determined by this means with previously published results on the binary alloy and with those obtained from X-ray analysis.

(a) *Previous Work*

Several authors have taken electron micrographs of age-hardened alloys: Castaing and Guinier [7–11] have investigated aluminium–copper and aluminium–magnesium–silicon alloys, and Keller and Geisler [12–14] have given results on aluminium alloys of commercial interest. Nishimura and Murakami,[15] Buinov and Lerinman,[16] and Saulnier and Syre,[17] have also taken electron micrographs of a series of age-hardened aluminium alloys. The relation between the structure of the alloy and the appearance of the replica has been discussed by Keller [12] and by Castaing.[10]

(b) *Experimental Technique*

Most workers on aluminium alloys have obtained replicas of the surface by depositing and stripping an oxide film. This technique has been followed, but a new method of stripping has been employed.

The microspecimens were first electropolished in De Sy and Haemers's solution,[5] and, after a quick wash in running water, they were then anodized in a bath consisting of 12% $Na_2HPO_4.12H_2O$ and 0·4% H_2SO_4 in distilled water, using a lead cathode. The thickness of the film depends primarily on the applied voltage and only to a small extent on the anodizing time. A potential difference of 20 V. for 2 min. proved satisfactory.

The conventional method of stripping involves immersion in a solution of mercuric chloride and results in total loss of the specimen. In addition, the films require careful washing before they are clean enough to be used as replicas. An alternative scheme was therefore adopted.[18] The specimen was removed from the anodizing bath and washed; cuts with a razor blade were then made $\frac{1}{8}$ in. apart, and the specimen was transferred to the Jacquet [19–21] electropolishing bath (65% acetic anhydride and 35% perchloric acid). The usual cooling jacket was discarded and a beam of light focused on the specimen to allow careful observation of the surface. An initial potential difference of 10 V. was slowly increased to 20 V. over a period of 2 min. and then maintained at this value until the films could be seen to be lifting from the surface. The specimen was then carefully placed in a beaker of distilled water and gently agitated until the squares of the anodic film became detached from the surface. A suitable replica was collected on a fine copper grid, dried, and placed in the electron microscope for examination.

In most cases the microspecimens were not etched before anodizing. Further replicas could readily be obtained once the film had been stripped completely, if the above procedure was repeated.

(c) *Experimental Results*

Examples of the structures observed are shown in Figs. 4–7 (Plate LXVI), and those shown in Figs. 8 and 9 (Plate LXVI) compare the results obtained by optical and electron microscopy on an aluminium–4% copper alloy aged for 96 days at 190° C. The platelets in Figs. 4–7 are both dark and light, but this is due chiefly to the particle size in relation to the replica thickness and not to a difference in alloy composition. After fairly short ageing times the platelets in the ternary alloys have a dotted appearance (Fig. 5), which is absent from the binary alloys (Fig. 4). No such difference was detected after longer ageing times when the particles were larger. Castaing and Guinier [8] have reported an appreciable variation in thickness along the diameter of individual θ' platelets in the aluminium–4% copper alloy aged at temperatures below 250° C. In the present work, the effect was much more noticeable in the ternary than in the binary alloy. The micrographs (Figs. 4–7) show that the particles were more numerous in the ternary than in the binary alloys.

The platelets visible in the micrographs can be identified as θ' from previous [22] and parallel [23, 24] X-ray work. A few small dots were noted in the aluminium–4% copper alloy which had been aged to peak hardness at 130° C. Here, X-ray analysis indicates the presence of G.P. zones [2] with a small quantity of θ'. No structures were observed in this alloy at times shorter than the time to reach peak hardness (50 days) at 130° C., although Castaing [10] has recorded a mottling effect, at magnifications of 33,000 in the 4% copper alloy aged 12 hr. at 150° C., due to G.P. zones [2]. Contrary to the results of Buinov and Lerinman [16] and of Saulnier and Syre,[17]

TABLE I.—*Sizes of θ' Particles.*

Ageing Time	Alloy	Diameter, Å.	Thickness, Å.
	Ageing Temperature 130° C.		
50 days (peak hardness)	Al–4% Cu	Dots approx. 100	...
6 days (peak hardness)	Al–4% Cu–0·05% Sn	<700	<100
200 days	Al–4% Cu	1000–5000	200
200 days	Al–4% Cu–0·05% Sn	<1500	100
	Ageing Temperature 190° C.		
1 day (peak hardness)	Al–4% Cu	900–5500	100
7 hr. (peak hardness)	Al–4% Cu–0·05% Sn	<1000	<100
3 days	Al–4% Cu	1000–10,000	<300
3 days	Al–4% Cu–0·05% In	500–3500	<300
96 days	Al–4% Cu	1000–9000	100–500
96 days	Al–4% Cu–0·05% In	900–7500	200–500
300 days	Al–4% Cu–0·05% In	1500–7500	500–600

small white dots were not observed in the micrographs of the aluminium–4% copper alloy aged for very short times.

The ternary alloy aged to peak hardness at 130° C. (6 days) showed the θ' precipitate already present to an appreciable size, and the same effect was noted in both the binary and ternary alloys aged to their peak hardnesses at 190° C. (1 day and 7 hr., respectively). The particle sizes are given in Table I, in which the values quoted for the thickness of the smaller plates are less reliable than those given for their diameters. This is because the resolving limit of the electron microscope employed was in the range 100–200 Å. It will be seen from Table I that, except in the specimens aged to peak hardness at 130° C., the θ' particles were larger in the binary than in the ternary alloy. This applied even though the θ' particles had been formed much earlier in the ternary alloys. For example, after 200 days at 130° C. the θ' particles have been present for at least 195 days in the ternary alloy, but only for about 150 days in the binary alloy, and yet the particles were appreciably larger in the latter. A similar effect occurred after 3 days at 190° C., although the sizes were more equal after 96 days at this temperature. The particle sizes reported previously are for rather scattered ageing conditions which do not allow a direct comparison with the present work. The available results indicate size ranges fairly similar to those of the binary alloy in Table I.

(d) *Discussion of Results*

It is not possible to deduce the ageing mechanism of the ternary alloys, because no structures were visible before the formation of θ'. However, the results clearly indicate the formation of θ' before the peak hardness is attained in the ternary alloy at 130° C., whereas θ' is formed only at ageing times close to the peak hardness in the binary alloy at the same temperature. For equal ageing times, and for similar relative positions on the ageing curves, the formation of the platelets of θ' is much more advanced in the ternary alloy, although the particles are smaller and much more numerous. The results support the general conclusion [4] that the ternary addition (cadmium, indium, or tin) accelerates the nucleation process in the ageing of aluminium–copper alloys, but a discussion of the mechanism must await the results of X-ray analysis.

ACKNOWLEDGEMENTS

The authors' thanks are due to their colleagues at the Fulmer Research Institute who contributed to the experimental work, and particularly to Mrs. E. M. Smith, B.A., for her assistance in obtaining the electron micrographs.

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1459 The Log–Log Plot of Solubility Data in Ternary Metallic Systems *

By H. K. HARDY,† M.Sc., Ph.D., A.R.S.M., A.I.M., MEMBER

SYNOPSIS

The slope of the solubility curve of an ideal ternary solution plotted on log–log co-ordinates has been analysed for the case when the phase in equilibrium is a ternary compound. A straight line is not predicted, but the slope will be equal to the ratio of the solute components in the compound when the same ratio exists between the solute components in the solution in equilibrium. The curvature will be slight for very dilute solutions, whose solubility data will give a straight-line plot as though in equilibrium with a binary compound.

HUME-ROTHERY [1] has deduced from mass-action principles that the isothermal solubility curve of a ternary solution should give a straight line when the atomic fractions of the solute elements are plotted on logarithmic co-ordinates, provided that (*a*) the solution is in equilibrium with an intermetallic compound composed of the two solute elements, and (*b*) the solution is thermodynamically ideal, so that the activities are equal to the atomic fractions. In addition, the straight line is expected to have a negative slope of the same value as the ratio of the constituents in the intermetallic compound. Hume-Rothery [1] showed that the principle applied to the liquidus isothermals of the aluminium–magnesium–silicon liquid in equilibrium with Mg_2Si, a case which has been further examined by Phillips.[2] Similar behaviour has been found experimentally when the solution is in equilibrium with a ternary compound, e.g. the $\alpha/(\alpha + Al_2CuMg)$ boundary of the aluminium–copper–magnesium system,[3] and the $\alpha/(\alpha + Al_2Mg_3Zn_3)$ boundary of the aluminium–zinc–magnesium system,[4] give straight-line log–log plots of the isothermal solubility curves.

The case in which the solution is in equilibrium with a ternary compound has not been examined theoretically. Let the equilibrium be represented by the equation :

$$n_X \,.\, X + n_Y \,.\, Y + n_Z \,.\, Z \rightleftharpoons (X_{n_X} \,.\, Y_{n_Y} \,.\, Z_{n_Z})$$

The sum of the chemical potentials is equal to zero at equilibrium and

$$n_X \,.\, \mu_X + n_Y \,.\, \mu_Y + n_Z \,.\, \mu_Z = \mu_{XYZ}$$

Putting $\mu_i = \mu^0{}_i + RT\ln a_i$, where a is the activity, leads to :

$$\ln a_{XYZ} - n_X \,.\, \ln a_X - n_Y \,.\, \ln a_Y - n_Z \,.\, \ln a_Z =$$
$$-\left(\frac{\mu^0{}_{XYZ} - n_X\mu^0{}_X - n_Y\mu^0{}_Y - n_Z\mu^0{}_Z}{RT}\right) = -\frac{\Delta F^0}{RT} = \ln K.$$

For homogeneous equilibrium all the reactants and products are gases or are in solution. For heterogeneous equilibrium it is conventional to give the activity of the condensed phase a constant value, which here applies along the equilibrium solubility curve, and we have :

$$\phi = n_X \,.\, \ln a_X + n_Y \,.\, \ln a_Y + n_Z \,.\, \ln a_Z - C = 0$$

where $C = \ln a_{XYZ} - \ln K$.

For an ideal solution the activities may be replaced by the atomic fractions and differentiation then gives :

$$\frac{d \ln y}{d \ln z} = -\frac{\partial\phi/\partial \ln z}{\partial\phi/\partial \ln y} = -\frac{n_Z}{n_Y}\left(\frac{1 - \frac{n_X}{n_Z} \cdot \frac{z}{x}}{1 - \frac{n_X}{n_Y} \cdot \frac{y}{x}}\right)$$

since $\partial x/\partial y = \partial x/\partial z = -1$ is obtained from the subsidiary condition $x + y + z = 1$.

We need to consider only the range of composition in which $x/z > n_X/n_Z$ and $x/y > n_X/n_Y$. For a binary compound $n_X = 0$ and the slope reduces to the value given by Hume-Rothery.[1] On the other hand, the term in brackets is unity for a ternary compound only when $z/y = n_Z/n_Y$ and the slope of $-n_Z/n_Y$ is limited to one point on the solubility curve. When $z/y < n_Z/n_Y$, the slope will become more negative (tending towards $-\infty$). When $z/y > n_Z/n_Y$ the slope will become less negative (tending towards zero), and this effect is to be seen in the results of Brommelle and Phillips for the liquidus curves of the Al_2CuMg phase.[5] If the solution is very dilute, the fractions z/x and y/x will be very small, and the slope will be close to $-n_Z/n_Y$. This probably accounts for the pseudo-binary behaviour of the $\alpha/(\alpha + Al_2CuMg)$ and $\alpha/(\alpha + Al_2Mg_3Zn_3)$ boundaries, although other compensating factors may also be present.

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* Manuscript received 29 December 1952.

† Senior Metallurgist, Fulmer Research Institute, Ltd., Stoke Poges, Bucks.

MICROSTRUCTURES OF ZINC-22·5% ALUMINIUM ALLOY ISOTHERMALLY TRANSFORMED.

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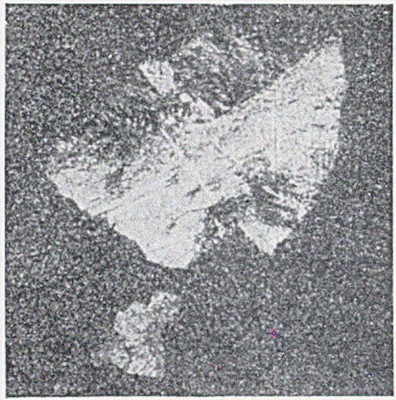

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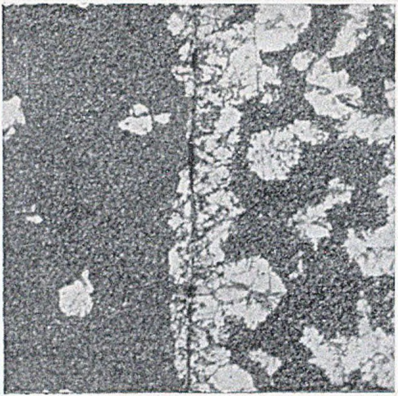

(a) (b)

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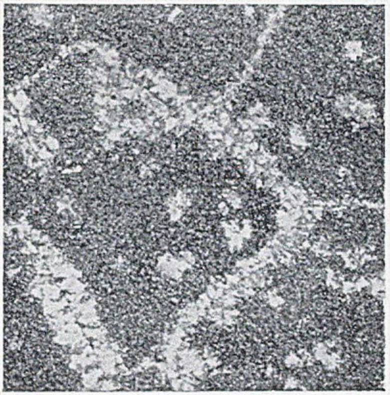

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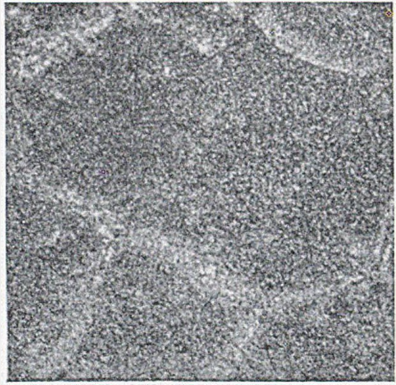

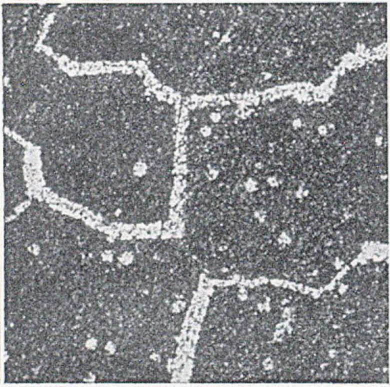

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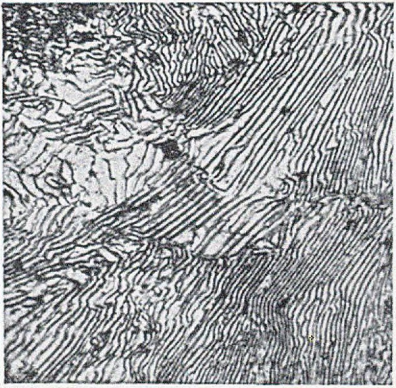

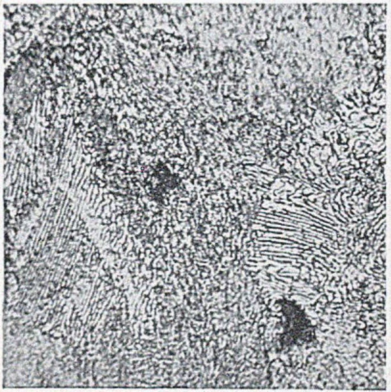

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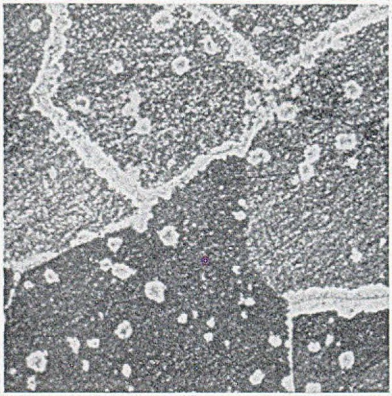

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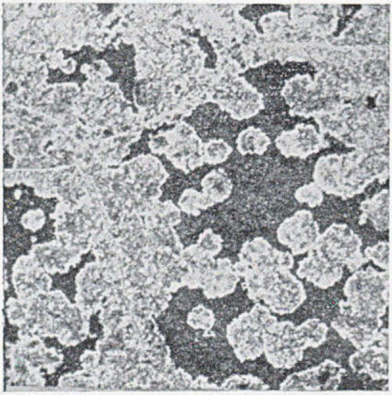

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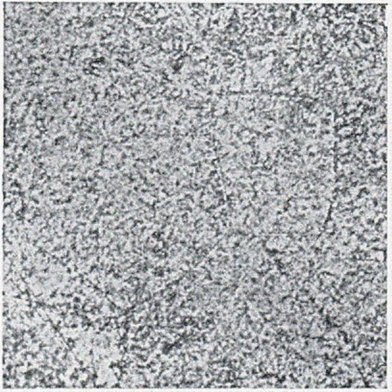


FIGS. 4–8.—Transformation Products at Various Temperatures. Prepared by Method I. × 200.

FIG. 4.—10,000 sec. at 265° C.; typical pearlite-group nodule.
FIG. 5.—70 sec. at 240° C.
FIG. 6.—20 (a) and 25 (b) sec. at 220° C.; troostitic nodules.
FIG. 7.—16 sec. at 180° C.
FIG. 8.—10 sec. at 150° C.

FIG. 9.—As Fig. 8, but Prepared by Method II. × 200.

FIG. 10.—Coarse Lamellar Product Obtained After 3200 sec. at 260° C. Whorling beginning in top left-hand corner. × 600.

FIG. 11.—Similar to Fig. 10, but Showing Decrease of Interlamellar Spacing at Lower Temperature. 400 sec. at 240° C. × 1000.

FIGS. 12–15.—Stages of Decomposition at 80° C. Prepared by Method II. × 200.

FIG. 12.—6 sec. FIG. 13.—10 sec. FIG. 14.—22 sec. FIG. 15.—100 sec.

[To face p. 432

PLATE LXII.

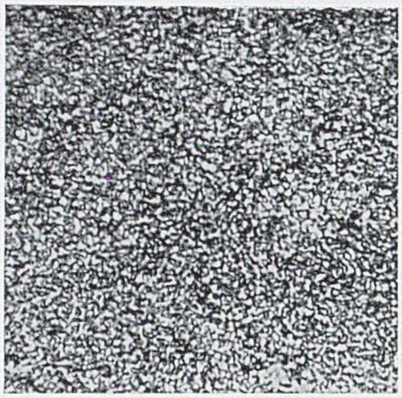
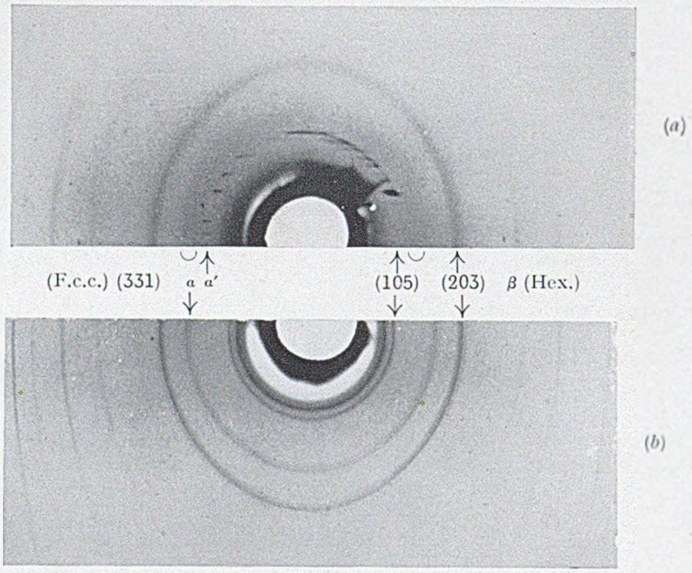


Fig. 16.—Granular Nature of Decomposition Product After 3 Months at Room Temperature. × 1300.



∪(331) α halo in newly formed transformation product

Fig. 17.—X-Ray Back-Reflection Photographs Taken During Decomposition of Zinc–22·5% Aluminium Alloy at Room Temperature (18° C.). (*a*) Between 2 and 15 min. after quenching; (*b*) 3 months later.

LEAD–THALLIUM ALLOYS. × 150.

← Direction of Applied Stress →

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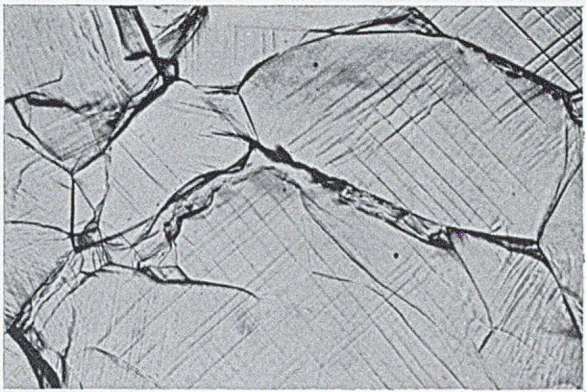

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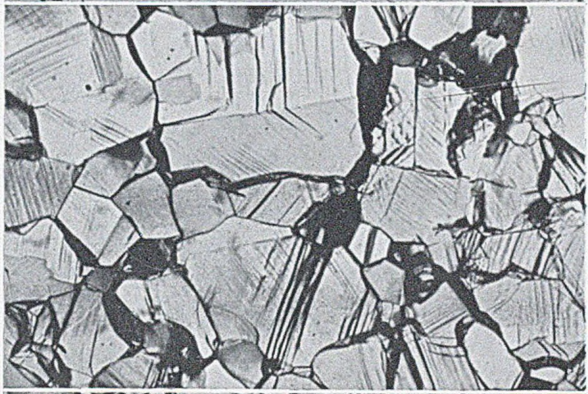

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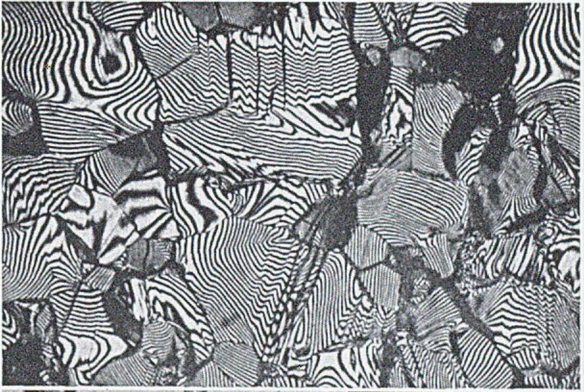

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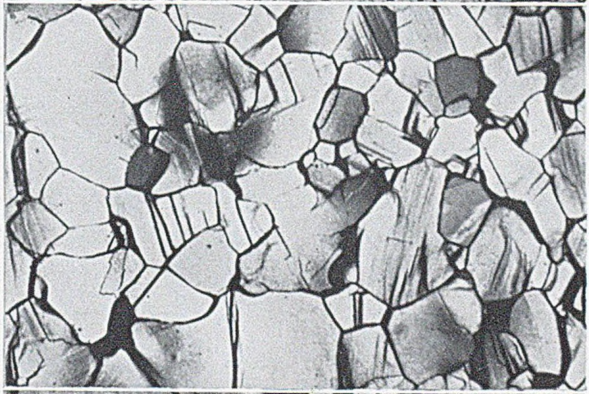

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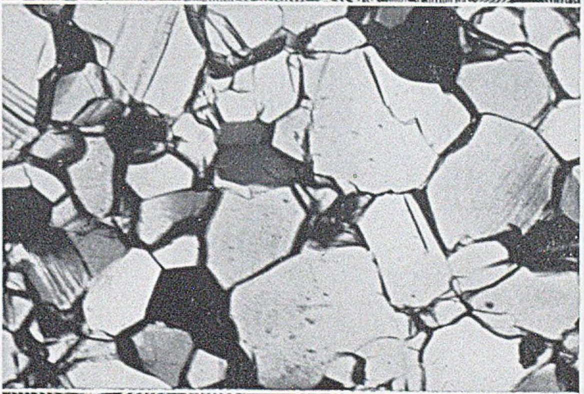

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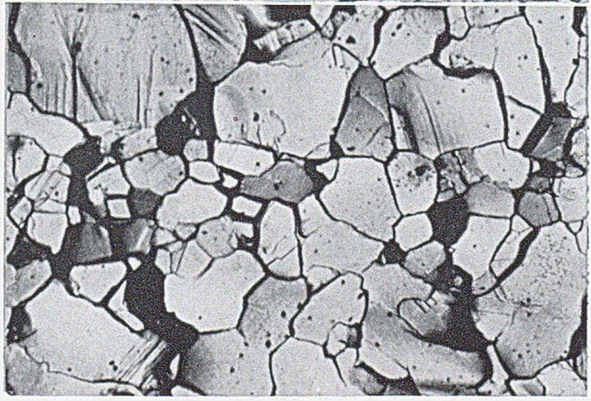

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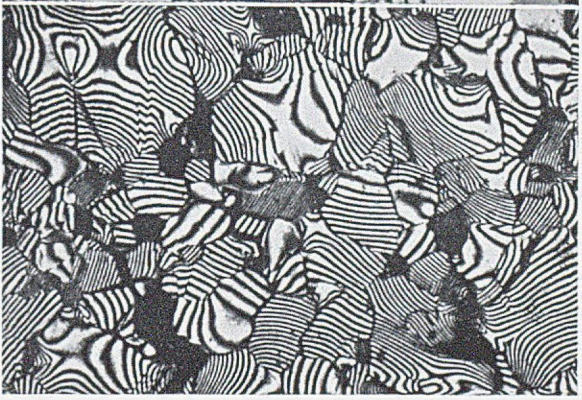


FIG. 1.—Pure Lead; 500 lb./in.^{2}; 4·5% Extension in 8 days.
FIG. 2.—Pure Lead; 300 lb./in.^{2}; 5% Extension in 270 days.
FIG. 3.—1·25% Thallium; 500 lb./in.^{2}; 10% Extension in 50 days.
FIG. 4.—Multiple-Beam Interferogram of same Area as Fig. 3.
FIG. 5.—7·87% Thallium; 500 lb./in.^{2}; 10% Extension in 11 days.
FIG. 6.—2·45% Thallium; 200 lb./in.^{2}; 10% Extension in 190 days.
FIG. 7.—13·45% Thallium; 500 lb./in.^{2}; 10% Extension in 24 days.
FIG. 8.—Multiple-Beam Interferogram of same Area as Fig. 7.

LEAD–THALLIUM ALLOYS.

← Direction of Applied Stress →

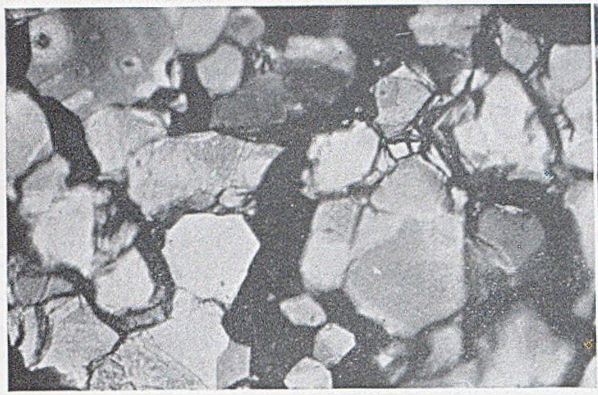


FIG. 9.—13·45% Thallium; 200 lb./in.^{2}; 45% Extension in 475 days. × 400.

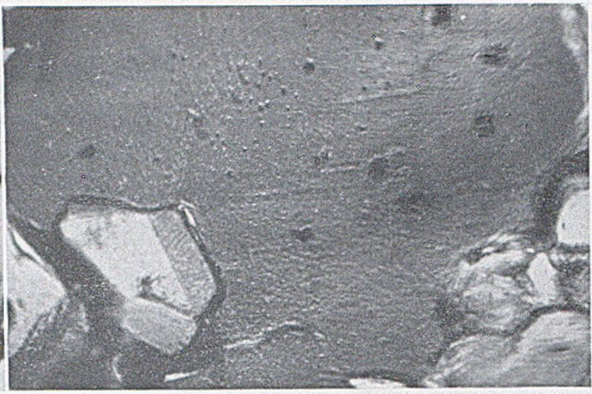


FIG. 10.—As Fig. 9. Another area, with negative phase-contrast illumination. × 600.

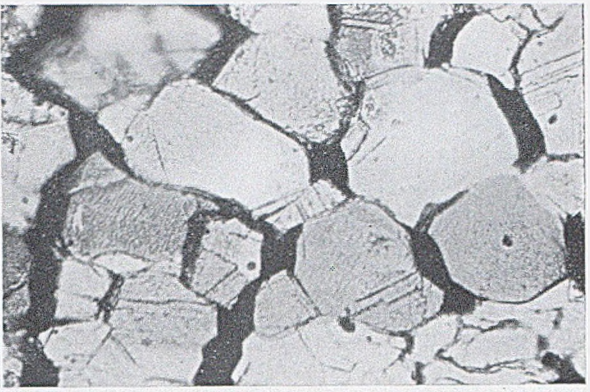
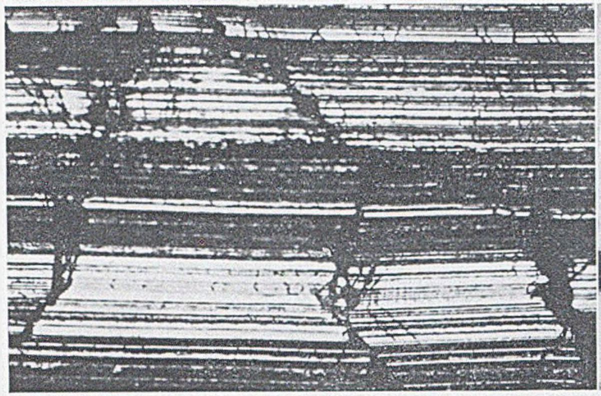


FIG. 11.—26·58% Thallium; 300 lb./in.^{2}; 50% Extension in 875 days. × 400.

FIG. 12.—Pure Lead; 500 lb./in.^{2}; 5% Extension in 9 days; Unpolished. × 150.



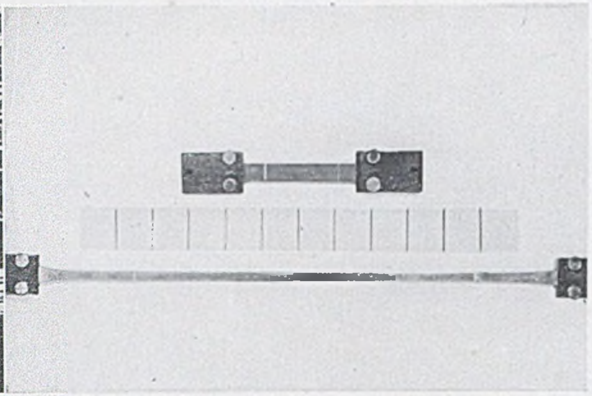
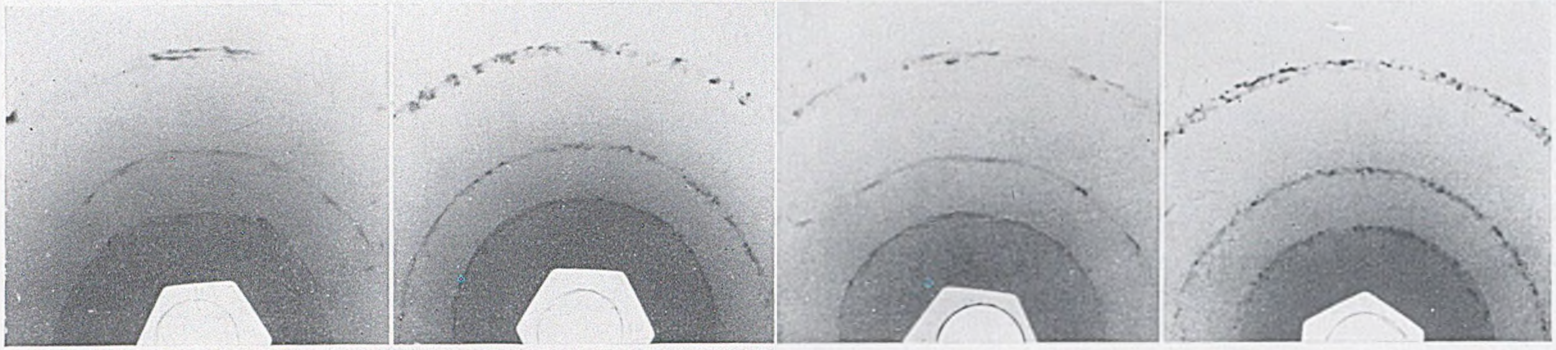


FIG. 13.—Fractured Test-Piece of 7·87% Thallium Alloy after Testing at 300 lb./in.^{2}, Showing 372% Extension in 221 days, Compared with an Untested Specimen. Approx. × $\frac{1}{3}$.



FIGS. 14–17.—X-Ray Back-Reflection Photographs.

FIG. 14.—Pure lead; 500 lb./in.^{2}; 4·5% extension in 8 days.

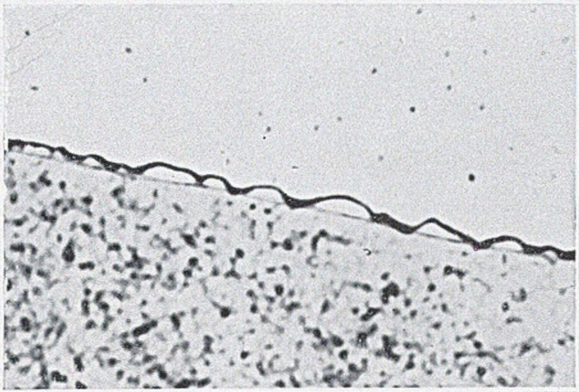
FIG. 15.—2·45% Thallium; 200 lb./in.^{2}; 10% extension in 190 days.

FIG. 16.—4·81% Thallium; 200 lb./in.^{2}; 169% extension in 930 days.

FIG. 17.—13·45% Thallium; 300 lb./in.^{2}; 4·5% extension in 65 days.

STRUCTURES OF AGE-HARDENED ALUMINIUM ALLOYS.

2



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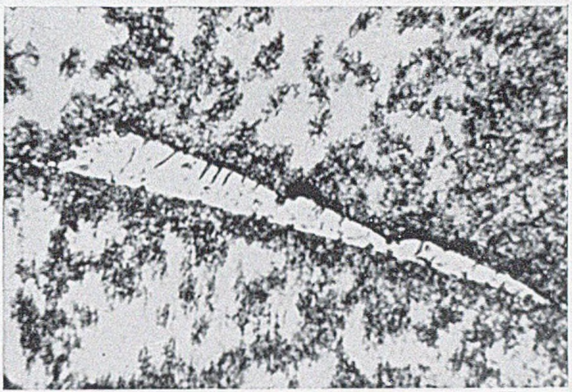

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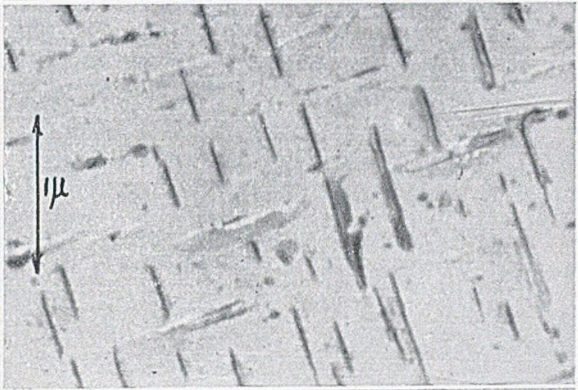

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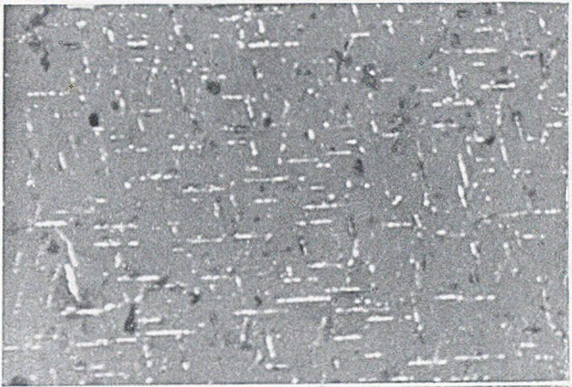

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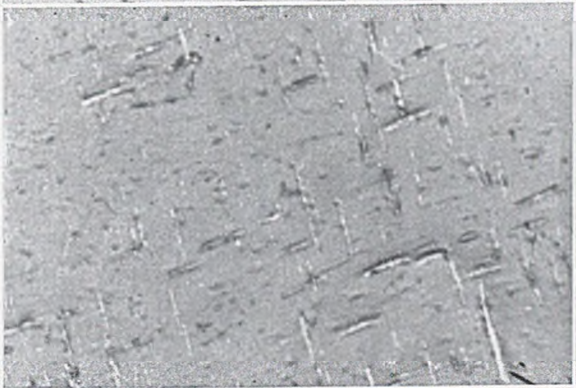

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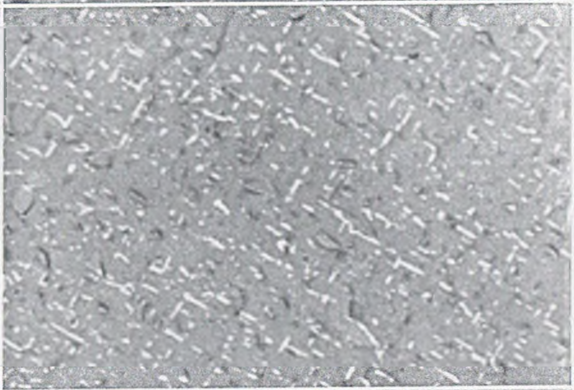

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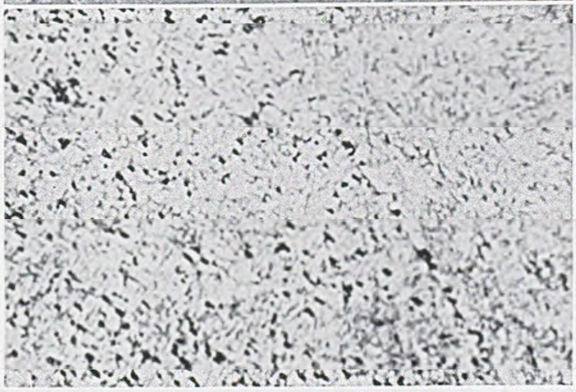

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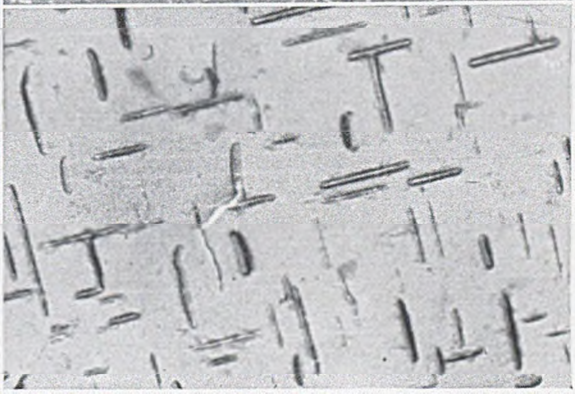

FIG. 2.—Al-4% Cu. Aged 16 hr. at 130° C. Optical microscope. × 2500.
FIG. 3.—Al-4% Cu. Aged 200 days at 130° C. Optical microscope. × 1600.
FIG. 4.—Al-4% Cu. Aged 3 days at 190° C. Electron microscope. × 20,000.
FIG. 5.—Al-4% Cu-0·05% In. Aged 3 days at 190° C. Electron microscope. × 20,000.
FIG. 6.—Al-4% Cu. Aged 200 days at 130° C. Electron microscope. × 20,000.
FIG. 7.—Al-4% Cu-0·05% Sn. Aged 200 days at 130° C. Electron microscope. × 20,000.
FIG. 8.—Al-4% Cu. Aged 96 days at 190° C. Optical microscope. Black plates in lower grain are approximately parallel to the plane of polish. × 1600.
FIG. 9.—Al-4% Cu. Aged 96 days at 190° C. Electron microscope. × 20,000.

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 20 APRIL 1953 PART 8

1 — PROPERTIES OF METALS

***Effects of Heating at 300° and 500° F. (150° and 260° C.) on the Properties of Aluminium and Copper Wires.** Allen S. Russell (*Wire and Wire Products*, 1952, 27, (3), 255–256, 307).—R. reports tests on Al and Cu wires of No. 18 and 20 A.W.G. Values for the elect. conductivity, U.T.S., Y.P., and elongation are given as functions of heating times up to 376 days at 300° F. (150° C.) and 286 days at 500° F. (260° C.). The hard-drawn Al softened slowly at 300° F., the conductivity increasing by 2% in 1 year. The U.T.S. decreased linearly with log time; Cu was not affected. In 1 day at 500° F. the conductivity of Al increased 2%; that of Cu diminished 8–12% in 286 days. Both deteriorated in U.T.S. Oxidation was not observed to be an important factor for either metal at 300° F., nor did the Al oxidize at 500° F. Cu shed an oxide film at the higher temp.—C. P. F.

***Anelastic Measurement on the Recovery of Cold-Worked Aluminium.** Nobuhiko Kunitomi and Teruko Ogura (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, 16, (7), 365–369).—[In Japanese, with diagrams in English]. The changes of internal friction accompanying the recovery from cold work in Al were measured and compared with the changes in other phys. properties, viz. hardness and thermoelect. power. Results show that: (1) the internal friction increases with increased work, (2) the values of internal friction and other properties decrease exponentially, the times and asymptotic values being functions of the annealing temp., and (3) the internal friction changes markedly on annealing at low temp., but other phys. properties do not change. K. and O. explain the results by considering that the internal friction is caused by the viscous behaviour of the grain boundaries, which contain many kinds of lattice defects.—AUTHORS.

***Diffusion of Metals in Samples of Sintered Aluminium [SAP].** W. Seith and G. Löpmann (*Z. Elektrochem.*, 1952, **56**, (4), 373–379; discussion, 379).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. Although the diffusion of metals in Al has been much studied, consistent results are difficult to obtain owing to the marked affinity of Al for O resulting in the Al becoming coated with an oxide film which cannot be reduced by H. v. Zeerleder, Rohner, and Irmann recently succeeded in producing a type of sintered Al, called " SAP ", which appears to possess some outstanding properties for tech. appn. (*Z. Metallkunde*, 1950, **41**, 228; *M.A.*, **19**, 293). SAP is made by sputtering degreased pure Al in an atmosphere contg. little O, whereby platelets of Al a few tenths μ thick and of length 3–5 μ are obtained and an O content of 5–20% is produced. The powder is compressed under a pressure of 20–50 kg./mm.^{2}, sintered at 550°–600° C., and hot pressed to a sp. gr. of 2·75 g./c.c. Phys. properties of the resulting product are given. In particular, the elect. conductivity of SAP is only ~25% below that of pure Al, although it is reasonable to assume that an oxide layer exists between individual grains of the sintered material; the thermal conductivity of SAP is only ~20% less than that of pure Al, while that of tech. Al alloys is only half that of pure Al. A comparison of the diffusion of Ag, Cu, Mg, Si, and Zn in SAP and in pure Al shows that the diffusion process is not hindered by the oxide layers presumably surrounding the grains of SAP; on the contrary, the alloy metals diffuse particularly quickly in SAP. Diffusion experiments using Al test-pieces in which contacting surfaces had been oxidized electrolytically showed no hindrance to diffusion produced by the Eloxal layer.—J. S. G. T.

***The Scattering of Slow Electrons by Solid Aluminium Surfaces.** O. Krenzien (*Z. Physik*, 1949, **126**, (6), 598–600).—Freshly evaporated Al surfaces were found to have a reflection coeff. of 0·15–0·22 for electrons with energies of 5–10 eV. After a few hr. at room temp. in dry O, the coeff. rose to 0·6. The high values of ~0·6 previously reported are thus attributed to an oxide layer.—J. W. C.

***A New Calculation of Some Properties of Metallic Beryllium.** B. Donovan (*Phil. Mag.*, 1952, [vii], **43**, (343), 868–874).—The quantum-mech., cellular-method, calculations of the lattice const., compressibility, cohesive energy, and work-function of Be metal, briefly reported previously (*Nature*, 1951, **168**, 836; *M.A.*, **19**, 689) are described in detail. Excellent agreement was found between the calculated and observed lattice const. and compressibilities. The cohesive energy is slightly overestimated, suggesting that the correlation energy has been underestimated by using Raimes's method (*Phil. Mag.*, 1950, [vii], **41**, 568; *M.A.*, **18**, 244). The disagreement between the calculated and observed work-functions is less than that of Herring and Hill (*Phys. Rev.*, 1940, [ii], **58**, 132; *M.A.*, **7**, 429), but still too large to be explained by a double surface-layer term.—P. C. L. P.

***Erosion and Heat Transfer with Liquid Metals [Bismuth and Lithium].** H. G. Elrod and R. R. Fouse (*U.S. Atomic Energy Commission Publ.*, 1950, (**NEPA-1270**), 10 pp.).—A statement of the problems and proposed investigations connected with the heat transfer of liq. Bi and erosion characteristics of Li.—B. W. M.

***A Surface Property of Copper Between 500° and 850° C.** F. Erdmann-Jesnitzer and F. Günther (*Z. Elektrochem.*, 1952, 56, (4), 386–389; discussion, 389–390).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. Lichtenecker (*ibid.*, 1942, **48**, 601; *M.A.*, **11**, 325) found that the surface of a Cu or Ag sphere, heated by a Bunsen burner to 700°–800° C., suddenly seemed to look wet, and it was easy to distinguish between an oxidation zone, which appeared dark, and a reduction zone on the sphere comprising the wet part of the surface. Interference colours were shown by both zones. The experiment has now been repeated with various refinements, using elect. heating in N or *in vacuo*, and a goniometer or X-rays to examine the state of the surface of the Cu sphere. The existence of a molten superficial layer of metal on the surface of the sphere at an abs. temp. equal to 0·7 × m.p. of Cu, suggested by

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

L., is confirmed; the thickness of the molten layer is of the order 10^{-3}–10^{-4} mm. Results obtained by comparing the expansion of the surface and the mass expansion of solid Cu in the form of either a block or wire are rather inconclusive in respect of deciding the suggestion made by Hume-Rothery and Andrews (*J. Inst. Metals*, 1942, **68**, 19; *M.A.*, **9**, 344) that the coeff. of expansion of Cu may be different for the surface of the metal and the metal *en masse*; the discrepancy found between results relating to the block of metal and wire are possibly attributable to superficial oxidation or the presence of slight impurity.—J. S. G. T.

***Diffusion Research [Diffusion of Silver into Copper]: Progress Report.** R. Smoluchowski, L. Couling, S. Hayes, C. W. Haynes, and R. W. Turner (*U.S. Atomic Energy Commission Publ.*, 1952, (**NYO-3184**), 3 pp.).—A study was made of the diffusion of Ag at 700° C. into a bicrystal of Cu cut so as to vary the angle between the direction of diffusion into the grain boundary and the direction of growth of the bicrystal. The results indicated a variation of penetration from 0·019 to 0·015 in. with a 90° variation of this angle for a bicrystal having an orientation difference of 28°. Owing to the loss of Ag from evaporation, the results are not considered to be conclusive. Attempts were made to compare the results obtained by etching with those given by an autoradiographic technique, but improvements in the latter method are required before satisfactory results can be obtained. The low-angle X-ray scattering in samples of Ag–Al alloys aged at 100° C. was found to decrease to a min. after 0·5 hr. ageing time and to increase on further ageing. The effect was related to an initial decrease of conc. fluctuations in the solid soln., followed by a growth of large clusters. —B. W. M.

***Heat-Exchange Between a Copper Surface and Liquid Hydrogen and Nitrogen.** Robert N. Mulford and Joseph P. Nigon (*U.S. Atomic Energy Commission Publ.*, 1952, (**LA-1416**), 7 pp.).—The rate of heat transfer from a Cu surface into liq. H at a pressure of 590 mm. Hg was determined for temp. differences between the surface and the liq. of 0·1°–300° C. The heat was supplied at a known rate calculated from the power input to a Cu-covered elect. heater immersed in the liq. gas. A max. rate of ~5·8 ± 0·4 W./cm.^{2} was found at a temp. difference between 1° and 2° C. for both rough and smooth surfaces when liq. H was used and ~6 W./cm.^{2} at a temp. difference of 4° C. for liq. N. —B. W. M.

***Radiation Damage and Recovery in Copper, Silver, Gold, Nickel, and Tantalum.** J. W. Marx, H. G. Cooper, and J. W. Henderson (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECU-2118**), 19 pp.).—Annealed specimens of Cu, Ag, Au, Ni, and Ta foils were bombarded in a cyclotron with 12 MeV. deuterons at ~ −140° and −150° C. The change of elect. resistivity was found to increase with the integrated flux and with the temp. of bombardment. The difference in the results at the two temp. was considered to be due to a thermal recovery process which was detected at temp. as low as −165° C. and was characterized by an activation energy of 0·2 ± 0·05 eV. for Cu, Ag, and Au and 0·3 ± 0·1 eV. for Ni and Ta. Various possible mechanisms are suggested to explain the results. It is considered that the primary effect of particle bombardment in pure metals is to produce interstitial vacancy pairs and that migration of the vacancies is responsible for the initial recovery process. 16 ref.—B. W. M.

***Thermal Conductivity of Germanium.** A. Grieco and H. C. Montgomery (*Phys. Rev.*, 1952, [ii], **86**, (4), 570).—A letter. The thermal conductivity along the (100) direction of a high-purity *n*-type Ge single crystal was found to be 0·14 cal./sec./cm./°C. at 25° C. and roughly 20% less at 100° C.—P. C. L. P.

***Measurement of the Absorption Coefficient of Gold in the Ultra-Violet and Visible Regions.** Wilhelm Elfers (*Z. Physik*, 1949, **125**, (4/6), 276–277).—Measurements of the decrease in intensity of light passed through suitable thin plates of Au were made, using a Pulfrich photometer in the visible region, and a counter apparatus in the ultra-violet. The absorption coeff. varies from 4·98 ($\lambda = 754$ mμ) to 1·73 ($\lambda = 254$ mμ).—J. W. C.

Hafnium Metal: Its Properties and Future. John L. Everhart (*Materials and Methods*, 1952, **36**, (5), 95–97).—The properties of Hf produced by the iodide process are compared with those of Zr and Ti. In addn., the corrosion-resistance of the three metals is compared, and the forming characteristics, heat-treatment, joining, cleaning, finishing, and future appn. of Hf are reviewed.—R. P. H. F.

***Some Mechanical and Adhesive Properties of Indium.** A. C. Moore and D. Tabor (*Brit. J. Appl. Physics*, 1952, **3**, (9), 299–301).—It is shown, by indirect methods, that the normal force required to break a cohesive bond between In and steel depends only on the area of bond and on the time that elapses before rupture. Creep and Meyer hardness measurements show that the rupture stress ≃ Meyer hardness, or ≃ 3 × yield stress, for the same time of stressing. This implies that the In always yields before rupture. Adhesion of a Cu hemisphere to flat In is increased by rotation of the former; this is due to the enhanced plastic flow in In under these conditions. The adhesive force of In to various metals and non-metals was measured. It is greater than the yield stress of In for diamond, glass, WC, several metals, and thick oxides on Cu and Ag. On plastics the adhesion is lower. For polytetrafluoroethylene plastic it is very low (though some transfer of plastic to In does occur), and the friction is also low. Here adhesion evidently plays the part, in friction, played by cold welding in the friction of metals.—R. W. C.

***Atomic Heat of Indium at Liquid-Helium Temperatures.** J. R. Clement and E. H. Quinnell ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 89–97).—Any satisfactory theory of supraconductivity must explain the difference in the sp. heats of the material in the normal and supraconducting states. Graphs are given of the atomic heat of: (1) Sn, (2) supraconducting In, (3) normal In, together with graphs of (4) C_N/T against T^2 for normal In, (5) C_S/T against T^2 for supraconducting In, and (6) $(C_S - C_N)/T$ against T^2 for liq.-He temp. up to 12° K., C_S denoting the sp. heat in the supraconducting state, and C_N the sp. heat in the normal state. The value of the Sommerfeld electronic sp. heat coeff. (γ) is $3{\cdot}23 \times 10^{-4}$ cal./mole/deg.^{2}. The Debye characteristic temp., θ_D, is ~106° K.—J. S. G. T.

***Magnetic Properties of a Hollow Supraconducting Lead Sphere.** Julius Babiskin ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 77–79).—The nature of the equatorial magnetic field distributions inside and outside a hollow supraconducting Pb sphere, of dia. 2 in. and thickness $\frac{3}{16}$ in., was studied at 4·2° K. Initial conditions for supraconductivity could be restored either inside or outside the sphere, but not both simultaneously.—J. S. G. T.

***Erosion and Heat Transfer with Liquid Metals [Lithium]: Progress Report.** —— (*U.S. Atomic Energy Commission Publ.*, 1950, (**NEPA-1423**), 9 pp.).—The design and calibration of an electro-magnetic flowmeter for use with Li are described, together with details of a double-loop system for testing the erosion characteristics of Li in the range 1100°–1600° F. (595°–870° C.).—B. W. M.

***The Electrical Resistance of Magnesium, Aluminium, Molybdenum, Cobalt, and Tungsten at Low Temperatures.** J. G. Thomas and E. Mendoza (*Phil. Mag.*, 1952, [vii], **43**, (343), 900–910).—The elect. resistance/temp. min. was found to occur at different temp. in specimens cut from a length of wire made from a Mg rod of purity >99·98%, contg. 0·013 wt.-% Fe. The normal behaviour of the magneto-resistance and the independence of temp., of the magnetic susceptibility suggest that the vacancies in the *d*-shells of the Fe and other transition-metal impurities have been eliminated. Similar measurements on an Al alloy contg. 0·1 at.-% Mn gave no indication that the *d*-shell vacancies of Mn remained. The resistance of several polycryst. Mo wires went through a min., then rose to a const. value below 0·1° K. High-temp. vacuum annealing caused a supraconducting impurity with a critical temp. of ~0·4° K. to form. Co and W wires showed no supraconductivity or resistance peculiarities down to 0·06° K. A brief discussion is given of the possibility that impurity atoms are responsible for the resistance/temp. min. shown by several materials.—P. C. L. P.

***The Influence of Temperature on the Mechanical Properties of Manganese.** E. M. Savitsky and V. E. Terekhova (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (1), 87–90).—[In Russian]. A systematic study of the mech. properties of Mn between $-195°$ and 1240° C. is described. The specimens were prepared from electrolytic Mn by melting in alumina crucibles in an induction furnace and casting into preheated cylindrical moulds. The properties studied were (*a*) hardness, which was measured in the furnace by pressing a cone of angle 90° into the specimen (height = dia. = 10 mm.) for 30 sec. with a load of 50–70 kg.; (*b*) plasticity, measured by suspending specimens of length 100 mm. and dia. 9·5 mm. horizontally and dropping a weight of 6 kg. on to them from a height of 1 m.; (*c*) relative shortening, which was measured by dropping 30 kg. from 1 m. on to the specimen (length 28 mm., dia. 18 mm.). In the last experiment the time for transferring the specimen to the stage was 2–4 sec., and to compensate for heat loss the furnace was heated 15°–20° C. above the required temp. For low-temp. work the coolants used were solid CO_2 and liq. N. α-Mn was hard and brittle at all temp. (400 kg./mm.^{2}). β-Mn was considerably softer (15 kg./mm.^{2} at transformation temp. and 1 kg./mm.^{2} at 900° C.), being mainly brittle, but at higher temp. becoming somewhat plastic. The $\alpha \rightarrow \beta$ transformation was well marked by a sharp drop in hardness, but other properties did not change appreciably. γ-Mn was soft and malleable and could withstand a 90% deformation. The $\beta \rightarrow \gamma$ transition was well marked by the type of deformation, brittle or plastic. Above 1180°–1200° C. the deformation again became brittle, owing to melting at grain boundaries.—Z. B.

***Magneto-Resistance of Supraconducting Mercury.** T. G. Berlincourt and C. T. Lane ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 273).—The changes of resistance, ΔR, at 3·5° and 1·2° K., in a magnetic field of strength 8 kOe. for Hg (average mass number 200·6) are found to be given by: at $T = 3{\cdot}5°$ K., $\Delta R/R_c \sim 2$; at $T = 1{\cdot}2°$ K., $\Delta R/R_c \sim 200$, R_c denoting the resistance of the metal in the critical magnetic field.—J. S. G. T.

***Supraconductivity of Isotopes of Mercury and Tin.** C. A. Reynolds, B. Serin, and L. B. Nesbitt ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 27–28).—Values of the critical magnetic field, H_0, at 0° K., and of the critical temp. of 4 Hg isotopes, of mass M, computed from experimental data extending down to 1·37° K. are: M, 199·5, H_0, 420 Oe., T_c, 4·185° K.; 200·7, 419, 4·175°; 202·0, 417, 4·160°; 203·2, 414, 4·146°. The corresponding results for three Sn isotopes are; M, 113·6, H_0, 312 Oe., T_c, 3·805°; 118·7, 304, 3·732°; 123·8, 298, 3·659°.—J. S. G. T.

[Discussion on a Paper by E. S. Byron and R. F. Baker:] Effects of Working on the Properties of Molybdenum. —— (*J. Electrochem. Soc.*, 1952, **99**, (12), 558).—See *M.A.*, **20**, 3.
—G. T. C.

***The Approach to Magnetic Saturation in Nickel Under Tension.** Otto Buhl (*Z. Physik*, 1949, **126**, (1/2), 84–97).—In sufficiently high fields, the intensity of magnetization, I, of a ferromagnetic material is related to the saturation intensity, I_s, by the formula $I = I_s\,(1 - b/H^2)$. Expressions for b are derived (i) for an externally applied stress $\parallel$ the field H, and (ii) for an isotropic internal-stress system. Measurements of induction were made on two specimens of carbonyl-Ni wire under various tensile loads. For small loads the variation of b with applied stress agreed with theory, but for larger loads it is necessary to postulate the existence of local internal stresses, of magnitude several times the external stress, due to plastic deformation. Measurements of the remanence and the coercive force as a function of applied stress were also made. With increasing tension the remanence fell off rapidly at first, and then became const., thus supporting the assumption of large internal stresses.—J. W. C.

The Law of Approach to Saturation in Polycrystalline Nickel. N. S. Akulov and N. Z. Miriazov (*Doklady Akad. Nauk S.S.S.R.*, 1949, **66**, (1), 29–32).—[In Russian]. The differential susceptibility of a cubic ferromagnetic material can be expressed:

$$\frac{dI}{dH} = \chi_p + \frac{A}{H^2} + \frac{B}{H^3} + \frac{C}{H^4} + \cdot \cdot \cdot \cdot$$

where χ_p is the susceptibility of the para-process, A differs from zero only when there is internal strain in the material, $B = 0{\cdot}152\ K_1{}^2/I_s$ and $C = 0{\cdot}115\ K_1{}^3/I_s{}^2$, and H is the field strength. The const. K, the first anisotropy const., can be calculated in magnitude but not in sign from B. The term CH^{-4} is very small and has hitherto not been determined experimentally. The object of the present work was to establish the existence of CH^{-4} and to determine its sign. This will give the sign of C and hence that of K_1. The material studied was electrolytic Ni melted *in vacuo*. The specimen was prepared in the form of an ellipsoid and had a demagnetization factor of $N = 0{\cdot}255$. It was annealed *in vacuo* at 1000° C. for 6 hr. and then furnace-cooled. dI/dH was measured at 17° C. and a field strength of 230–1500 Oe. At very high field strengths CH^{-4} is small; thus A and B can be calculated from the equation. C was obtained by plotting K against H. It was found to be negative. K_1 calculated from the equation taking into account CH^{-4} was at room temp. 6% higher than that calculated by neglecting CH^{-4}. This difference becomes greater at lower temp. Thus, for accurate work, either CH^{-4} should be taken into account or strong fields which would make its effect small, should be used.
—Z. S. B.

***Specific Heat of Niobium at Various Temperatures.** A. Brown, M. W. Zemansky, and H. A. Boorse ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 99–101).—Values have been determined of the sp. heat, C_n, of Nb at 64°–76° K. and are incorporated in a graph; a graph is also given of the corresponding Debye characteristic temp., θ_n, for the same temp. range. Value of C_n range from 2·52 cal./mole/°K. at 64° K. to 3·51 cal./mole/°K. at 76° K. The corresponding values of θ_n are 288° at 64° K. and 257° at 76° K. Preliminary detn. of θ_s, the Debye characteristic temp. of supraconducting Nb give $\theta_s = 160°$ K., compared with the value 69° K. found by Cook, Z., and B. (*Phys. Rev.*, 1950, [ii], **80**, 737; *M.A.*, **18**, 692).
—J. S. G. T.

***Nuclear Magnetic Resonance of Hydrogen Absorbed into Palladium Wires.** R. E. Norberg (*Phys. Rev.*, 1952, [ii], **86**, (5), 745–752).—Pd wires 0·03 in. in dia. of $>99{\cdot}5\%$ purity were loaded with H by a glow-discharge, to give samples with an atomic ratio H : Pd of 0·2–0·8. The measurements, using steady-state absorption and pulsed echo techniques, were made in the temp. range 218°–348° K. The results are interpreted to give information on the interaction of protons with the conduction electrons, and on the diffusion of protons within the Pd lattice. The small value of D_0 which is obtained, suggests that the chief mechanism of proton diffusion is via lattice imperfections. 27 ref.—P. C. L. P.

***The Extra Electrical Resistance Due to Cold Work and Neutron Irradiation of Platinum.** R. A. Dugdale (*Phil. Mag.*, 1952, [vii], **43**, (343), 912–914).—A letter. The decay of the extra resistance of two pile-irradiated and one cold-worked Pt resistor has been followed part of the way during annealing first for 640 hr. at 70° C., followed by 336 hr. at 90° C. Measurements of the extra resistance, r, at intervals, after cooling to the triple point of water, could be fitted to an empirical curve of the type $r = A - B \log (t + \gamma)$, where A, B, and γ are const. at a const. temp. The samples give the same activation energy suggesting that the same process is responsible. This result supports the suggestion of Seitz (*Advances in Physics*, 1952, **1**, 43; *M.A.*, **20**, 170) that part of the resistance caused by cold working metals is due to the creation of vacant lattice sites (and possibly interstitial atoms) by dislocations in motion.—P. C. L. P.

***Specific Heat of Silicon Below 100° K.** P. H. Keesom and N. Pearlman ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 279–283).—Values of the sp. heat of Si at 10°–100° K. have been determined. The measurements really give values of C_p, the sp. heat at const. pressure, but C_v differs from C_p by $<0{\cdot}4\%$ below 100° K. The sp. heat/temp. graph of the values obtained in the present work is slightly above those found by Nernst and Schwers (*Sitzber. Preuss. Akad. Wiss.*, **1914**, 355) and by Anderson (*J. Amer. Chem. Soc.*, 1930, **52**, 2301). The Debye characteristic temp., θ_D, is not const. over the temp. range, as predicted by the Debye theory, so that in particular, C_v is not $\propto T^3$ below ~40° K. 11 ref.—J. S. G. T.

***Self-Diffusion in Pure Polycrystalline Silver.** L. Slifkin, D. Lazarus, and T. Tomizuka (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECU-2076**), 13 pp.).—The self-diffusion rate in pure polycryst. Ag was measured, using Ag^{110} as a tracer, at six temp. in the range 450°–936° C. Above 576° C., diffusion was predominantly of the vol. type and the results showed good agreement with those of other investigators. Calculation by the method of least squares using all the available data on vol. diffusion gave a diffusion coeff. of $0{\cdot}724 \exp(-45{,}500/RT)$. Below 576° C., the activity decreased exponentially with the first power of the penetration depth, indicating grain-boundary diffusion. 8 ref. —B. W. M.

***Fundamental Research in Physical Metallurgy [Self-Diffusion of Silver]: Progress Report.** J. Hollomon and D. Turnbull (*U.S. Atomic Energy Commission Publ.*, 1952, (**SO-2022**), 5 pp.).—Preliminary measurements have been made on the rate of self-diffusion of Ag at a single grain boundary, using two bicrystals in which the angles between the 100 poles of the two grains were 27° and 45°, resp. Calculated grain-boundary diffusion coeff. were in essential agreement with each other and with results previously obtained on polycryst. samples having random orientation.—B. W. M.

***Investigation of the Exchange and Diffusion Process Occurring at Silver Electrodes by Means of Radioactive Indicators.** (Gerischer and Vielstich). See col. 587.

***Electron Emission from a Silver Surface Under the Influence of Iodine Vapour.** H. Paproth, W. Rathje, and J. N. Stranski (*Z. Elektrochem.*, 1952, **56**, (4), 409–413; discussion, 413–414).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. The elect. emission from a Ag cathode, in the presence of I vapour, under an applied voltage up to 7 V. and heated at 500°–750° C., is shown experimentally to consist entirely of electrons, and to be explainable in terms of the so-called chemico-electronic emission theory, developed by Denisoff and Richardson to explain the electronic emission of liquid alloys (*Proc. Roy. Soc.*, 1924, [A], **105**, 468; 1935, [A], **150**, 495). 12 ref.—J. S. G. T.

***Preparation of Grey Tin.** E. S. Hedges and J. Y. Higgs (*Nature*, 1952, **169**, (4302), 621–622).—Grey Sn has been prepared by cold working 99·99% pure Sn below the $\alpha \rightleftharpoons \beta$ transformation temp. (13·2° C.) and allowing it to recrystallize, the transformation being almost complete in 24 hr.; the rate increases with the purity. A few grains of this grey Sn sprinkled on commercially pure white Sn at low temp. will cause the latter to transform to grey Sn.—R. S. B.

***Properties of Metallic Solutions. IV.—Surface Tension of Tin and Tin–Sodium Alloys.** N. L. Pokrovsky and N. D. Galanina (*Zhur. Fiz. Khim.*, 1949, **23**, (3), 324–331).—[In Russian]. Cf. Semenchenko, Bering, and P., *ibid.*, 1936, **8**, 364; *M.A.*, **3**, 656. Measurements were made on Sn *in vacuo* at 250°–450° C. by measuring the max. pressure during the formation of a drop on the upper end of a capillary tube. The glass apparatus, evacuated to a pressure of 10^{-5} mm. Hg, was arranged in such a way that solid Sn, initially in a bulb outside the furnace which surrounded the lower end of the apparatus, could be melted with an external heater so that it ran down into a U-tube with a capillary tip in one limb. H was then admitted into the other limb, under a pressure measured with a special manometer, until drops formed at the capillary tip, where the surface was still *in vacuo*. The drops were collected in a separate bulb for chem. and metallographic analysis. The surface tension, σ, was calculated by the formula $\sigma = \frac{1}{2}rgP_m$, where r is the radius of the capillary, g is the acceleration due to gravity, and P_m is the max. pressure during drop formation, calculated from the applied H pressure and the difference in liq. Sn level between the two limbs of the U-tube. Hogness's values (*J. Amer. Chem. Soc.*, 1921, **43**, 1621; *J. Inst. Metals* (Abstracts), 1922, **28**, 537) for the *d* of liq. Sn were used. P. and G. find that the surface tension at temp. t (°C.) is given by $\sigma = \{537 - 0{\cdot}078\,(t - 232)\}$ dynes/cm.; commercial and spectrochem. pure Sn gave identical results. This line is slightly above H.'s values (*loc. cit.*) and appreciably below the values of Bircumshaw (*Phil. Mag.*, 1927, [vii], **3**, 1288; *J. Inst. Metals* (Abstracts), 1927, **38**, 371) and of Sauerwald (*Z. anorg. Chem.*, 1926, **154**, 79). P. and G.'s results give an Eötvös const. of 0·266 and a parachor which varies with temp., in agreement with previous results. The temp. coeff. of surface tension is twice as great as that calculated theoretically by Frenkel and Gubanov (*Zhur. Eksper. Teoret. Fiziki*, 1946, **16**, 435; *M.A.*, **15**, 258), probably because their calculation did not allow for the effect of the electron gas. Alloys of Sn with Na, contg. 0·006–0·96 wt.-% (0·03–4·6 at.-%) Na, were prepared by melting a master alloy, contg. 1·89 wt.-% Na, *in vacuo* with vigorous stirring, and then melting this alloy with Sn *in vacuo*. The surface tensions of 11 alloys were measured, and the results obey the relation $\sigma = 532 - 38 \ln (24c + 1) + 14c$, where c is the concentration of the alloy. This relation is of the form predicted theoretically by Semenchenko (*Zhur. Fiz. Khim.*, 1947, **21**, 613, 707, 1387). The adsorption, calculated from the experimental relation by means of the Gibbs equation, has a max. value consistent with monomolecular adsorption. Alloys contg. 0·01–0·48 wt.-% Na were cast and annealed *in vacuo* in contact with glass plates, and the cast surfaces were examined microscopically. The results show that Na is soluble in solid Sn up to 0·48 wt.-%, and that the grain-size of the solid soln. decreases with increasing Na content.—G. B. H.

***High-Frequency Resistance of Tin, Lead, and Indium.** C. J. Grebenkemper and John P. Hagen ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 103–108).—The surface resistances R, of Sn, Pb, and In have been measured, using a resonant cavity operating in the lowest mode, and the effect of surface finishes was studied. The normal surface conductivity, σ, of Sn just above the transition temp. is considerably lower than the value expected from classical theory; this result agrees with numerous other experiments and the Reuter–Sondheimer theory of the anomalous skin effect (*Proc. Roy. Soc.*, 1948, [A], **195**, 336; *M.A.*, **16**, 772). A pressed Sn specimen gave the lowest value of σ in the normal region, viz. 117 ohm^{-1}. Cast Sn, with inner surface lathe-turned, gave σ = 133 ohm^{-1}. The best mech.-polished specimen gave σ = 160 ohm^{-1}. Results for the different types of surface finishes of Sn at 1·6°–3·2° K. are given in the forms of graphs. All the graphs coincide at ~3·2° K. Both surface roughness and surface strain contribute to the resistance in the supraconducting region. The value of σ for Pb (cast, machine-finished) was 176 ohm^{-1}, somewhat above that of Sn. If the normalized skin resistance of Pb is extrapolated to 0° K., a residual resistance of ~1% is found. Pb is about as good a conductor as Sn in the supraconducting region. In is similar to Sn, except that the initial value of σ is quite low. The normalized surface resistance of In, extrapolated to 0° K., gives a residual resistance of ~0·50%. The work on Sn has been extended to frequencies as high as 24,000 Mc./s. The results show considerably better conductivity in both the normal and supraconducting regions than those found by Maxwell, Marcus, and Slater (*Phys. Rev.*, 1949, [ii], **76**, 1332; *M.A.*, **17**, 630). —J. S. G. T.

***Thermal Conductivity in the Intermediate State of Pure Supraconductors [Tin and Indium].** D. P. Detwiler and Henry A. Fairbank (*Phys. Rev.*, 1952, [ii], **86**, (4), 574).—A letter. The thermal conductivity of spectroscopically pure coarse-grained Sn and In rods at 2·5 and 2·1° K., resp., showed a min. in the intermediate state, together with some hysteresis, when a variable transverse magnetic field was applied. A brief discussion is given of the possible implications regarding the mechanism of heat transport in the supraconducting and intermediate states.—P. C. L. P.

***Supraconductivity of Tin Isotopes.** J. M. Lock, A. B. Pippard, and D. Shoenberg ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 31–32).—Transition temp., T_c, and critical magnetic fields, H_c, of each of two samples of three Sn isotopes of resp. masses, M, 116·2 ± 0·05, 119·9 ± 0·05, and 123·75 ± 0·1 were determined. The isotopes of resp. masses 116 and 124 showed particularly sharp transitions, so that the difference ΔT_c in their T_c values in zero magnetic field was very accurately found to be 0·108 ± 0·001° K. Abs. values of T_c on the 1949 thermometric scale are : M 116, 3·767° K.; 120, 3·712° K.; 124, 3·659° K. The isotopes that showed sharp transitions of T_c also showed very sharp magnetic transitions at const. temp. The graph of ΔH_c against T^2 is not quite linear. The value of H_0/T_c is the same for all the isotopes, within the limits of experimental error. There is good evidence for the geometrical similarity of the critical-field curves of the different isotopes. One consequence of this is that the electronic sp. heat of Sn is independent of M. A more important consequence of this constancy of H_0/T_c is that the theoretical prediction of Fröhlich (*Phys. Rev.*, 1950, [ii], 79, 845; *M.A.*, 18, 502) and of Bardeen (*ibid.*, 1950, [ii], **80**, 567; *M.A.*, 18, 757) that $H_0 \propto M^{-\frac{1}{2}}$ is very nearly substantiated, the value of the index being found to be 0·462 ± 0·014. This result lends strong support to the suggestion that supraconductivity is intimately dependent on interactions between electrons and the crystal lattice. The small numerical disagreement in calculated and experimental values of the index is probably to be explained by inexactitude of the theory in details.—J. S. G. T.

***Recent Work at the National Bureau of Standards on the Isotope Effect [Relating to the Supraconductivity of Tin and Thallium].** E. Maxwell ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 29–30).—Detn. of the critical fields, H_0, and transition temp., T_c, of Sn and Tl isotopes have been made. Quant. results for Tl are not yet available, but the isotope shift has been found to be normal for isotopes of resp. masses 203·3 and 205·0, i.e. the isotope of heavier mass has the lower T_c, so that a suggested simple correlation between at. vol. and isotope effects is ruled out. Owing to some secondary effects, not yet fully understood, the critical-field curves for the different samples of Tl used, including natural Tl, are slightly shifted from run to run, although the consistency and precision in any one experiment is good. Critical-field/temp. graphs for Sn isotopes of resp. masses 116·8, 118·1, 119·8, 123·1, and natural Sn (118·70) are given; resp. values of T_c are 3·7680°, 3·7438°, 3·7202°, 3·6658°, and 3·7384° K.—J. S. G. T.

***Penetration of a Magnetic Field into Thin Supraconducting Films [of Tin, Lead, and Indium].** J. M. Lock ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 73–75).—The penetration of a magnetic field into thin films of Sn, Pb, and In, deposited *in vacuo* upon thin mica sheets, was studied by measuring the magnetic susceptibility, (χ), from the initial gradients of magnetization (I/H) curves of the metals at various temp. χ decreases with rise of temp., owing to an increase of penetration depth as the transition temp. is approached. χ also decreases with decrease of film thickness, in accordance with the phenomenological theory of F. and H. London. Values for the penetration depth at 0° K. are $5{\cdot}0 \pm 0{\cdot}1 \times 10^{-6}$ cm. for Sn, $3{\cdot}9 \pm 0{\cdot}3 \times 10^{-6}$ cm. for Pb, and $6{\cdot}4 \pm 0{\cdot}3 \times 10^{-6}$ cm. for In. The magnetization curves become more rounded as the film thickness is reduced. The surface energy/unit area between the normal phase and an insulator probably differs only very slightly from that between the supraconducting phase and the insulator.—J. S. G. T.

***The Surface Migration of Tungsten on Its Own Crystal Lattice.** Erwin W. Müller (*Z. Physik*, 1949, **126**, (7/9), 642–665).—A fine W point was heated to various temp. in the range 1200°–2700° K. and its shape studied by observation of the field emission. The tip became rounded almost to hemispherical shape after heating above 1200° K. From the rate of change of the radius, it is deduced that the surface migration rate is $5.10^{24} \exp(-Q/RT)$ cm.$^{-2}$ sec.$^{-1}$, where the molar activation energy, Q, has the value 106,500 cal. In the temp. range studied, the rate of evaporation is negligible compared with the rate of surface diffusion. A tip formed by evaporating W from a filament had the usual faces and edges of a cubic crystal; these became rounded after annealing above 1200° K. The results are compared with a theory due to Stranski and Suhrmann.—J. W. C.

***The Reversible Dissociation of Hydrogen Molecules and the Para-Hydrogen Conversion [on Tungsten].** A. Couper and D. D. Eley (*Proc. Roy. Soc.*, 1952, [A], **211**, (1107), 536–543).—C. and E. discuss recent work by Rideal and Trapnell (*ibid.*, 1951, [A], **205**, 409; **206**, 39; *M.A.*, 19, 567, 568) on the adsorption of H by W. Calculations are made of the abs. velocity, activation energy, and pressure dependency of the reaction at temp. below 0° C., and pressures of <1 mm. Hg; the values obtained are found to be in disagreement with those observed experimentally. This leads to the conclusion that the reversible adsorption by dissociation of H mol. into the chemisorbed layer is too slow to account for the para-H conversion, but still leaves open the question as to how far this process will account for the reversible adsorptions experimentally observed by T. The interchange mechanism: $p\text{-H}_2 + \text{HW} \rightarrow \text{WH} + o\text{-H}_2$ is preferred to that advanced by R. and T. 16 ref.—E. N.

***The Activation Energy of the Para-Hydrogen Conversion on Tungsten.** A. Couper and D. D. Eley (*Proc. Roy. Soc.*, 1952, [A], **211**, (1107), 544–563).—Cf. *ibid.*, p. 536; preceding abstract. Using apparatus and techniques which are described in detail, a study has been made of the activation energy, E, and frequency factor, B^0, for the conversion of para-H on W wires, at temp. below 0° C., and pressures of ~1 mm. Hg. The results show that the E and B^0 values: (1) are independent (to a first approximation) of variations in wire length, and temp. and section size of the reaction vessels—factors which are likely to change the temp. distribution along the wire, (2) vary slightly from one wire to another, (3) are markedly modified by exposure to H atoms or Na vapour, and (4) for a "clean" surface, fit the equation: E (kg.cal.) $= -1 + 0{\cdot}6 \log_{10} B^0$, where $B^0 = 1{\cdot}82 \times 10^6$, and $2{\cdot}02 \times 10^6$ ml. cm.$^{-2}$ min.$^{-1}$ at 140° and 173° K., resp. The phys. reality of the equation relating E and B^0—the slope of which is probably determined by the resonance interaction in the adsorbed film, including the sites neighbouring the one on which the activated complex is formed—is supported by two facts: (i) a quite different slope—greater by ~1·2—is obtained if the W wire is "poisoned" with CO and other gases, and (ii) a relation of a different kind is found for Pd–Au alloys—an increase of E accompanied by a decrease in B^0. 26 ref.—E. N.

***Low-Temperature Thermal Expansion of Uranium.** Adam F. Schuch and Henry L. Laquer (*Phys. Rev.*, 1952, [ii], **86**, (5), 803).—A letter. The thermal expansion of U shows an inversion of sign between liq.-N and liq.-H temp. Since no variation of sp. heat occurs, the effect is not attributed to a phase transformation, and a d max. is indicated.
—P. C. L. P.

***Thermal Conductivity, Electrical Resistivity, and Thermoelectric Power of Uranium.** W. W. Tyler, A. C. Wilson, and G. J. Wolga (*U.S. Atomic Energy Commission Publ.*, 1952, (**KAPL-802**), 25 pp.).—The measurements were made on annealed β-U using Cu/Constantan thermocouples soldered to Cu pins which were pressed tightly in two 0·04-in.-dia. holes drilled 1·75 in. apart. The values of the observed Wiedemann–Franz ratio vary slightly from just below the theoretical value of $2{\cdot}45 \times 10^{-8}$ [V./°K.]^2 at 20° K. to more than this value at 300° K. This behaviour is similar to that of W, Mo, Pb, and indicates that the lattice does not contribute appreciably to thermal conductivity. Extrapolation of the data indicated a higher conductivity at 350° K. than given by Katz and Rabinowitch ("The Chemistry of Uranium", New York, 1951). In the range 20°–300° K., the thermal conductivity increased from 0·028 to 0·064 cal./cm./sec./°K. and the elect. resistivity from 4·0 to $29{\cdot}2 \times 10^{-6}$ Ω/cm. The abs. thermoelect. power of U was obtained from measurements of thermoelect. power of U/Cu and U/Constantan couples, using the known calibration of the Cu/Constantan couples and Borelius' data for the abs. power of Cu. The abs. thermoelect. power is positive for U and varies from ~2 μV./°K. at 20° K. to ~7·5 μV./°K. at 300° K., suggesting that conduction is predominantly due to holes. 7 ref.—B. W. M.

***The Thermal Electromotive Force of Uranium Against Platinum.** H. E. Cleaves and A. I. Dahl (*U.S. Atomic Energy Commission Publ.*, 1952, (**ANL-HDY-616**), 4 pp.).—Measurements of the thermal e.m.f. of U against standard Pt were made with various samples of U wire of ~0·03 in. dia. The thermal e.m.f. increased steadily from 0 mV. at 0° C. to 25·5 mV. at 900° C., no discontinuity being observed at either of the two transformations. Variations in the amount and nature of the impurities when the total impurity content was <0·1%, did not markedly affect the results.
—B. W. M.

***Nuclear Resonance and the Electronic Structure of Transition Metals [Vanadium and Niobium].** W. D. Knight and C. Kittel (*Phys. Rev.*, 1952, [ii], **86**, (4), 573).—A letter reviewing the various evidence concerning the possibility that the transition metals have an antiferromagnetic arrangement of *d* electron spins. Nuclear resonance experiments on V and Nb have disproved the existence of static antiferromagnetism.—P. C. L. P.

***On the Tensile Properties of Zinc Sheets Tested at Various Rates of Elongation and After Various Annealing Temperatures.** Rihei Kawachi and Kiyoshi Shinoda (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (7), 396–400).—[In Japanese, with tables and graphs in English]. It is known that the tensile properties of rolled Zn sheet vary greatly with the rate of elongation, but that those of annealed Zn sheet remain almost const., i.e. the former is superior in tensile strength but inferior in elongation to the latter when tested at low rates of elongation, and this is reversed at high rates. This abnormal phenomenon has been confirmed by detn. of the tensile properties of some Zn and Zn alloys, viz. (*a*) cold-rolled Zn–0·6% Pb sheet, (*b*) Zn–0–1·14% Pb alloy sheet cold rolled 0–70%, and (*c*) cold-rolled Zn alloy sheet contg. small amounts of Pb and Mn, at elongation rates of 1–140 mm./min. after various annealing treatments. The results are as follows: (i) the abnormality was clearly shown by alloy (*a*), not so clearly by (*b*), and not at all by (*c*); (ii) the variation of tensile properties decreased after treatments below the recrystn. temp.; and (iii) alloy (*c*) annealed at 200°–325° C. showed a hardening effect when tested at 1–100 mm./min., the max. strength being obtained at decreasing temp. with increasing rates of elongation. Specimens pulled at 140 mm./min. showed no max. It is suggested that further researches are necessary to clarify the abnormal phenomena.—AUTHORS.

***Yield Points in Bending Experiments on Zinc Crystals.** M. J. Dumbleton and B. W. Howlett (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 882–886).—It is shown by the use of a special bending machine, that undeformed Zn crystals exhibit the yield phenomenon, so that previous deformation and ageing are not necessary to produce this effect.—E. O. H.

***Some Observations on the Yield Point in Zinc.** H. L. Wain (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 886–896).—By careful tensile tests, it is shown that undeformed single crystals of Zn exhibit the yield phenomenon, and that prestraining is not necessary. It is also confirmed that dissolved N is necessary to produce the effect. On ageing a deformed crystal, it first recovered, and later the Y.P. reappeared, with a resulting increase in yield stress. The raising of the yield stress is proportional to log t (t=ageing time) and is temp.-sensitive. The results are discussed in terms of dislocation theory.—E. O. H.

Zirconium: A Bibliography of Unclassified Literature. Hugh E. Voress and Fred E. Croxton (*U.S. Atomic Energy Commission Publ.*, 1951, (**TID-3010**), 138 pp.).—A *bibliography* of 1134 abstracts of unclassified ref. to Zr, including data on its chemistry, metallurgy, physics, biology, and ceramics. Author, subject, and numerical indexes are given.—B. W. M.

***The Dissociation of Zirconium Iodide on Hot Metal Surfaces.** H. Döring and K. Molière (*Z. Elektrochem.*, 1952, **56**, (4), 403–407; discussion, 407–408).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. The decompn. of ZrI_4 and the deposition of Zr upon a heated Zr filament at various temp. of the filament (1070°–1506° C.) and of the surrounding vessel (150°–550° C.) were studied. After a time, equilibrium is reached between Zr, I, I_2, and various iodides. The results are interpreted in terms of the formation of iodides of lower valencies in accordance with the equations, $ZrI_3 + I = ZrI_4$; $4ZrI_3 = 3ZrI_4 + Zr$; $2ZrI_3 = ZrI_4 + ZrI_2$.—J. S. G. T.

Lanthanide Series Rare-Earth Elements: An Annotated List of Unclassified References. Robert L. Morgan (*U.S. Atomic Energy Commission Publ.*, 1952, (**TID-3029**), 6 pp.).—A *bibliography* of Atomic Energy Commission unclassified reports on the rare-earth metals, with emphasis on the properties and methods of prepn. of the elements.—B. W. M.

Fracturing and Fracture Dynamics. G. R. Irwin and J. A. Kies (*Weld. J.*, 1952, **31**, (2), 95S–103S).—A descriptive review is given of fracturing and fracture dynamics. A simplified direct appn. of energy balance principles to calculating the conditions for development of rapid fracturing is discussed. Photographs show cleavage in Mo. 9 ref.—K. B.

Rheotropic Embrittlement. E. J. Ripling (*Amer. Soc. Test. Mat. Bull.*, 1952, (**186**), 37–42; discussion, 42).—Metals not crystallizing in the f.c.c. system exhibit a ductility deficiency when strained at low temp., high strain rates, or in the presence of hydrostatic tension. A portion of this embrittlement can be overcome by prestraining under certain conditions, and this portion is called "rheotropic embrittlement". A review is given of several papers which describe the influence of a number of variables on rheotropic properties. Most work to date has been on steels, but the behaviour of Zn has also been investigated. 9 ref.—P. T. G.

The Elasticity of Crystals Near the Melting-Point. G. M. Bartenev (*Zhur. Fiz. Khim.*, 1950, **24**, (12), 1437–1441).—[In Russian]. On the basis of his theory of phase fluctuations (*ibid.*, 1948, **22**, 587; 1949, **23**, 1075; *M.A.*, **20**, 75), according to which a solid near the m.p. contains small regions of mass μ in which liquid is produced by thermal fluctuations, B. examines the rigidity modulus of a solid near the m.p., showing that, if the regions of liquid are assumed to be spherical, randomly distributed, and without resistance to shear, their existence reduces the actual rigidity modulus G to a value $G_N (1 - \alpha)^2$, where G_N is the value to be expected by extrapolation of the modulus/temp. curve from lower temp. and α is the mass fraction of liquid present, given approx. at temp. T by the relation (B., *ibid.*, 1950, **24**, 1016; *M.A.*, **20**, 392) $\alpha = kT_sT/\mu\gamma(T_s - T)$, where k is Boltzmann's const., γ is the latent heat of fusion, and T_s is the m.p. Comparison with Kornfel'd and Shestikhin's values (*Doklady Akad. Nauk S.S.S.R.*, 1942, **36**, 52; *M.A.*, **11**, 107) for the rigidity moduli of Sn and ice near the m.p. confirms this relation; for Sn the "quantum of fusion" $\mu = 1020$ atoms, compared with a value of 1400 atoms deduced from sp. heat anomalies near the m.p. (B., *loc. cit.*), and for ice $\mu = 4500$ atoms. K. and Sh.'s measurements were made dynamically at a frequency of 3000 c./s., so that thermodynamic equilibrium should have been rapidly established. Their results were the same for single crystals and polycrystals of Sn, and for pure and impure ice, so that the anomalies cannot be attributed to impurities.
—G. B. H.

Study Metal Behaviour with Large Metal Crystals. Robert Maddin and N. K. Chen (*Iron Age*, 1952, **170**, (17), 108–111).—The study of the plastic behaviour of polycryst. indust. metals is assisted by observations on the behaviour of large single crystals, though data for single crystals cannot yet be applied to polycryst. aggregates. Methods of growing single metal crystals are briefly outlined. Cf. C. and M., *M.A.*, **19**, 189, 535.—J. H. W.

Anisotropic Plastic Flow. J. C. Fisher (*Trans. Amer. Soc. Mech. Eng.*, 1949, **71**, (4), 349–356).—The math. theory of plasticity has been generalized to apply to plastically anisotropic materials. The anisotropic flow theory has been shown to reduce to the resolved shear stress/shear strain relationship for single crystals, and to the distortion energy theory for isotropic polycrystals. Experimental data concerning the plastic flow of anisotropic polycryst. Al indicate that the predictions of the anisotropic flow theory are in good agreement with experiment, and are significantly better than the predictions of the distortion-energy theory. As plastic anisotropy is the rule rather than the exception, caution is indicated in interpreting the results of combined-stress tests by means of the distortion-energy theory.—H. Pl.

Work-Hardening and Recovery as Basic Phenomena in Plastic Deformation. M. A. Bol'shanina (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (2), 223–231).—[In Russian]. B. puts forward the view that the course of plastic deformation is determined by work-hardening, which increases with increasing strain, and by recovery, or relaxation at a rate depending only on temp. Assuming (i) that work-hardening produces a relative increase of stress $(d\sigma/\sigma) = b(dl/l)$, where b is a const. and (dl/l) is the strain, and (ii) that recovery proceeds at a const. rate, causing a decrease of stress $a\,.\,dt$ in time dt, the stress/strain relation at const. strain-rate v is given by $\sigma = \sigma_0(l/l_0)^n$, where $n = b - (a/v)$. This equation is usually valid only at strains $>20\%$. At lower strains the experimental curve can be expressed, on the assumption that the recovery "const." $a \propto$ strain while b remains const., by the equation $\sigma/(\sigma-\mu) = c(l/l_0)^m$, where c, μ, and m are const. B. expresses the effect of strain-rate on plastic deformation by the "dynamic coeff." $z = \sigma_2/\sigma_1$, where σ_1 and σ_2 are the stresses required to produce the same strain at strain-rates v_1 and v_2, resp., $(v_2 > v_1)$. At low temp., recovery is negligible even at low strain-rates and z approaches 1; e.g. for some compression experiments on Pb, in which the low strain-rate was produced in an Amsler machine and the high strain-rate by a falling weight, z fell from 6·04 at 280° C. to 1·19 at −50° C. and 1·0 at −100° C. At high temp., close to the m.p., z also falls, because recovery is complete even at high strain-rates. With different metals at the same temp., z is lowest for those with the highest m.p. For every metal there is a temp. at which z reaches a max. Because of slower recovery, the fraction of the deformation energy absorbed in lattice strains is greater in metals of higher m.p., as shown by experiment. Similarly, any metal absorbs more energy in lattice distortion at high strain-rates than at low strain-rates, though X-ray line-broadening appears to indicate the contrary result, as it does not measure the energy absorbed. Lashko's measurements (*Zhur. Tekhn. Fiziki*, 1941, **9**, 1008) of the hardness of cold-worked metals after annealing indicate that a rapidly deformed metal contains highly strained regions, which recover rapidly, but that subsequent softening proceeds more slowly than in slowly deformed specimens. Many other experiments have shown that if rapid deformation is followed by slow deformation, the energy absorbed in the lattice is initially small, then rises to the normal value for deformation of an annealed metal. The same phenomenon is observed when deformation in one manner (e.g. wire-drawing) is followed by deformation in another manner (e.g. tension). —G. B. H.

The Theory of Glide-Hardening (with a Description of a Hardening Model). Albert Kochendörfer (*Z. Physik*, 1949, **126**, (6), 548–568).—A theory of work-hardening is developed in terms of the assumption that new dislocations can be formed by thermal agitation. The formation of a dislocation is accompanied by a distortion of the lattice in the immediate neighbourhood of the dislocation. The interaction between this distortion and bound dislocations already present in the lattice is responsible for the hardening, since the interaction raises the height and steepness of the energy barrier for a new dislocation. A mech. model is described which illustrates this process. The theory is compared with that of Mott and Nabarro.—J. W. C.

***Boiling and Condensing of Liquid Metals: Progress Report.** C. F. Bonilla, J. S. Bush, H. T. Chu, and B. Misra (*U.S. Atomic Energy Commission Publ*, 1952, (**NYO-3148**), 9 pp.).—The boiling at atmospheric pressure of a shallow layer of Hg on a Fe-plated surface was investigated. The boiling film coeff. of heat transfer obtained was in the range 1600–8400 B.Th.U./hr./ft.^{2}/°F., but no film boiling was observed up to a heat-flow rate of 260,000 B.Th.U./hr./ft.^{2}. Dropwise condensation of Hg vapour occurred at atmospheric pressure on a short water-cooled vertical Fe tube and appeared to impair the heat transfer slightly, owing to the adherence of the droplets to the surface. The condensing film heat-transfer coeff. were lowered by an oxide film which developed and by traces of non-condensable gas, values of 220–530 B.Th.U./hr./ft.^{2}/°F. being obtained. Data on the thermal e.m.f. of a Fe/Hg couple were ~10% higher than those given by Oosterhuis (*Arch. Néerland. Sci.*; 1912, [IIIA], **2**, 27) and were expressed up to 400° C. by $E = 0{\cdot}01939t - 8{\cdot}834 \times 10^{-6}\,t^2 - 9{\cdot}675 \times 10^{-9}\,t^3$ mV., where the temp. of the hot and cold junctions are t and 0° C., resp. 8 ref.—B. W. M.

***Fundamentals of Glass-to-Metal Bonding: Wettability of Some Group I and VIII Metals by Several Sodium Silicate Glasses.** Victor F. Zackay, David W. Mitchell, Stephan P. Mitoff, and Joseph A. Pask (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECU-2088**), 25 pp.).—No apparent correlation of the contact angles of three Na silicate glasses with Cu, Ag, Au, Ni, Pd, and Pt at 900° C. in He, H, air, and O was found with the position of the metals in the Periodic System. It appeared that within the range of glass compn. 30·9–36·9% Na_2O, there was no difference in contact angle for identical conditions. Contact angles on single-crystal and polycryst. specimens of Au and Ag were nearly equal, and the angle on Pt in a He atmosphere varied linearly with temp. between 900° and 1200° C. Young's equation indicates that any phenomenon which decreases interfacial tension between the molten glass and metal will tend to increase spreading. The observed variations in contact angle from metal to metal were apparently related to the polarizing power of the metal, lower angles being obtained for glasses on metals of greater polarizing power, indicating that the interfacial tension is lowered in such cases. In O-contg. and H atmospheres, chem. reactions are possible and result in the formation of surface oxides or hydrides, stable or transient, which appear to favour spreading. Such surface reactions may be expected to lower the interfacial tension between glass and metal. Although gases in solid soln. in metals might be expected to affect the interfacial tension between glass and metal, a limited study of the Ag–Pd alloy system which has a max. solubility for H at an intermediate compn., did not show a corresponding discontinuity in the contact angle/compn. curve. 7 ref.—B. W. M.

***†Sorption of Gases on Metal Films.** F. C. Tompkins (*Z. Elektrochem.*, 1952, **56**, (4), 360–363; discussion, 363).—[In English]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. Discrepancies between experimental results relating to the adsorption of gases by metal films are critically reviewed. The necessity for obtaining and using clean metal surfaces in adsorption work is stressed; it is considered that reduction of the oxide layer on metal films by H at 900° C. for 16 hr. followed by outgassing for 1 day is ineffective for cleaning the surface. Modern vacuum techniques combined with adequate protection against contamination, by the use of Hg cut-off and liq.-air traps, followed by conditioning the film by alternately raising and lowering the film temp. within a temp. range outside that used in the adsorption experiments, can produce a clean metal surface. T. does not consider that activated adsorption occurs on a metal film covered with O. 9 ref.—J. S. G. T.

The Adsorption State of Foreign Molecules in Conducting [Catalytic] Surfaces. R. Suhrmann (*Z. Elektrochem.*, 1952, **56**, (4), 351–359; discussion, 359–360).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. After brief ref. to the 12 experimental methods available for the study of the elementary processes occurring at, and the manner of activation of participants reacting at, a catalytic surface, an account is given of experimental studies of the electronic properties of catalytic surfaces. The following matters are discussed: (1) effect of electron affinity of a catalytic surface upon the adsorption process; (2) photoelect. emission and (3) elect. conductivity of a catalyst as methods of studying catalytic action; (4) experimental results relating to (*a*) the adsorption of H and O at Ni and Pt surfaces, and (*b*) N and NH_3 at Pt and Fe surfaces and (*c*) H_2O and N_2O at Pt surfaces. Mechanisms of the catalytic reactions occurring are briefly discussed. 33 ref.—J. S. G. T.

***Infra-Red Absorption by Metals at Low Temperatures.** K. G. Ramanathan ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, **(519)**, 257–259).—Values of the infra-red adsorption coeff. of Cu, Al, Pb, Sn, Sn + 1% In, Sn + 5·4% In, brass, Staybrite, Bi–Sn eutectic, and mech. polished brass, for $\lambda = 14$ μ at liq.-He temp. are in good agreement with theoretical values calculated in accordance with the anomalous skin-effect theory. Experimental values for alloys show good agreement with values calculated in accordance with classical theory, except Sn + 1% In, in which the resistivity is still low enough to produce the anomalous skin effect, and the Bi–Sn eutectic, in which the internal photoelect. effect may be responsible for the high absorption. The importance of using electrolytically polished surfaces for determining absorption coeff. is stressed. —J. S. G. T.

***Thermal-Resistance Measurements of Joints Between Stationary Metal Surfaces.** N. D. Weills and E. A. Ryder (*Trans. Amer. Soc. Mech. Eng.*, 1949, **71**, (3), 259–266; discussion, 266–267).—Results are presented of the measurements of thermal resistance of dry and oil-filled joints formed between two flat surfaces of various metals. Two test blocks 3 in. in dia. × 3 in. long were stacked axially under pressure on one another in contact with a heated Cu block at one end and a water-cooled Cu block at the other. The temp. at the joint ranged from 300° to 500° F. (150°–260° C.), the thermal current across the joint from 10,000 to 130,000 B.Th.U./hr./ft.^{2} and the pressure on the joint from 2 to 8000 lb./in.^{2}. The results indicate that: (1) the thermal conductance of a dry joint increases with pressure, linearly for steel, and generally exponentially for Al and bronze; (2) the thermal resistance of both dry and oil-filled joints decreases with a decrease in the roughness of the surface; (3) at a given temp., pressure, and roughness, the thermal resistance of both dry and oil-filled joints decreases in the order, steel → bronze → Al; (4) the ability to measure thermal conductivities of the metals with a standard error of 4% confirms other estimates of the precision attained in the experimental data generally.—H. Pl.

***The Effect of Temperature of Deformation on the Electrical Resistivity of Cold-Worked Metals and Alloys.** T. Broom (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 871–881).—A study has been made of the change in resistivity of Al, Cu, Ni, Fe, 50 : 50 Ag–Au, 75 : 25 brass, and Ag_3Mg, deformed by drawing in a special apparatus at −183°, −78·5°, 0°, and +100° C. The results show in every case that the max. increase occurs at the lowest drawing temp., but the magnitude of the change bears no relation to crystal structure or other property. A review is included of other work on this subject, and it is concluded that no correlations will be possible until the effects of the three types of lattice disturbance—dislocations, stacking faults, and vacancies—are separated.—E. O. H.

***Recovery of the Resistivity of Metals After Cold Working.** J. A. Manintveld (*Nature*, 1952, **169**, (4302), 623).—More accurate measurements than those already described (Molenaar and Aarts, *ibid.*, 1950, **166**, 690; *M.A.*, **18**, 500; Druyvesteyn and M., *ibid.*, 1951, **168**, 868; *M.A.*, **19**, 691) have been carried out on Au, Ag, and Cu wires. Activation energies for the two low-temp. resistance-annealing processes are calculated. The results suggest that the annealing is indeed due to the diffusion of single vacant lattice sites and aggregates of vacant lattice sites.—R. S. B.

***General Theory of Deviations from Matthiessen's Rule.** Max Kohler (*Z. Physik*, 1949, **126**, (6), 495–506).—A general formula is derived, according to which the elect. resistance of a metal can always be expressed in the form $\rho = \rho_i + \rho_R + \Delta$, where ρ_i is the temp.-dependent resistance of the ideally pure metal, ρ_R is the temp.-independent residual resistance at 0° K., and Δ is a term produced by the simultaneous interaction of thermal vibrations and static imperfections of the lattice. This formula should be applicable to all metals, and $\Delta \geqslant 0$ if the residual resistance is small, i.e. the metal is reasonably pure, so that the electronic band structure is not appreciably disturbed by the impurities. An approx. formula for Δ is derived in terms of two empirical const. In the temp. range where $\rho_i \leqslant \rho_R$, Δ is small compared with ρ_R, but in the region where $\rho_i \geqslant \rho_R$, Δ may be greater than ρ_R. The experimental results are reviewed. Most metals show some deviations from Matthiessen's rule ($\Delta = 0$) and for Pb, Pt, and Cu there are large deviations which can be explained qual. by the present theory. The min. in the resistance of Au at low temp. also implies a deviation from Matthiessen's rule.—J. W. C.

2 — PROPERTIES OF ALLOYS

Industrial Aluminium-Base Alloys. Jean Hérenguel (*Rev. Mét.*, 1952, **49**, (11), 765–776).—H. classifies the wrought Al alloys according to their principal properties: (1) *Conductivity*: 99·5% Al has an elect. conductivity 62% that of pure Cu and is used for overhead cables; Al–Mg_2Si alloy (Aldrey) (52%) is also used for cables; Al-0·8% Mg alloys (50%) for fine wires; and Al-3% Mg alloy (30%) for bars and tubes at transformer stations. (2) *Corrosion-Resistance*: 99·99–99·95% Al has the highest resistance to corrosion, though 99·5–99·7% Al is usually satisfactory for many purposes; Al–Mn (0·8–1·5%) alloys have been largely superseded by the Al–Mg alloys (0·8–5% Mg), which are now used extensively under a variety of conditions. (3) *Cold Working*: Al 99% and over is used for drawing and pressing; the Al–Mg (0·8–3·5% Mg) alloys are nearly as good as Al and are gradually replacing it; the Al–Zn–Mg alloys (Zn 3, Mg 1·75%) age at room temp.; the Al–Mg–Si (Mg 1, Si 1%) alloy has good corrosion-resistance but requires full heat-treatment; Al–Cu–Mg (Cu 2, Mg 1%) is used for cold driven rivets. (4) *Hot Working*: the Al–Mg–Si and Al–Zn–Mg alloys, as above, are used. (5) *Welding*: the Al–Mg alloys (up to 5% Mg) are useful under certain conditions; the Al–Mg–Si (Mg 0·7–0·9, Si 0·9–1·3%) and Al–Zn–Mg (Zn 3·5, Mg 1·75%) alloys have good welding properties, the former requiring heat-treatment after welding. (6) *High Strength*: Duralumin and the Al–Zn–Mg alloys are extensively used. (7) *High-Temp.*: These have addn. of Ni or Ni + Fe and are preferably made with 99·99% Al. (8) *Free-Cutting*: These are analogous to the free-cutting brasses (e.g. Pb-contg.). H. describes the probable developments of light alloys for various purposes. 14 ref.—J. H. W.

***The Properties of Some Binary Aluminium Alloys at Elevated Temperatures. [—II.]** J. V. Lyons and W. I. Pumphrey (*Metallurgia*, 1952, **46**, (278), 299–304).—The results previously reported (*ibid.*, (277), 219; *M.A.*, **20**, 461) are discussed and the implications considered. An attempt has been made to correlate high-temp. tensile properties of Al–Cu, Al–Mn, Al–Zn, and Al–Fe alloys with their tendency to crack during casting and welding. Good correlation has been obtained between the curve relating the variation with alloy content of the range of temp. above the solidus in which a brittle fracture is obtained in the tensile tests, and the curve relating alloy content with the tendency to crack during casting and welding.—F. M. L.

***Creep Correlations in Alpha Solid Solutions of Aluminium.** Oleg D. Sherby and John E. Dorn (*J. Metals*, 1952, **4**, (9), 959–964).—A report of the results of stress-rupture and creep tests carried out at various temp. on pure Al (99·987%), and its binary solid-soln. alloys with Mg 1·617, Cu 0·101, and Zn 1·616%. The specimens were (1) prepared from homogenized 0·100-in. sheet which was rolled to 0·07 in., and recrystallized to a mean grain dia. of ~0·25 mm., and (2) machined so that their tensile axes were in the rolling direction of the final sheet. All tests were performed under const.-load conditions, the testing temp. being 305°, 373°, 422°, 477°, and 528° K. The results show that at temp. above 400° K. (when recovery becomes active), the creep and tensile data can be simply related by the Zener–Hollomon relation (*J. Appl. Physics*, 1944, 15, 22), and that the activation energy (ΔH) for creep—in both its primary and secondary stages—is 35,800 cal./g.-mol. irrespective of grain-size, alloy or compn., and structure. Below ~400° K., however, the more complex phenomena associated with strain-hardening enter and prohibit the unmodified appn. of the Zener–Hollomon concept. Tensile and creep data, obtained by other workers, for 2S-O (commercial-purity Al) can also be correlated by the Zener–Hollomon parameter, but the ΔH values obtained differ from one series of data to another, dependent upon the annealing conditions preceding creep. When creep strains for a stated creep stress are plotted in terms of a temp.-compensated time, $\theta = t \exp(-\Delta H/RT)$, where t is the time under stress, a single creep curve is obtained, independent of the creep temp. T. This suggests a simple method for extrapolation of short-time (10^3 to 10×10^3 hr.) creep data to service lives of as much as 250×10^3 hr. All compn. given are in at.-%. 11 ref.—E. N.

***Creep Tests at 26°, 100°, and 130° C. on Wrought Anticorodal B, Age-Hardened Avional-23, and Annealed Peraluman-50.** E. von Burg (*Schweiz. Archiv angew. Wiss. Techn.*, 1952, 18, (3), 73–86).—[In German]. In an attempt to improve on the information obtained by the standard German short-time creep test (D.V.M.), three Al alloys were tested up to 15,000 hr. and creep limits are quoted as the stresses (in % of the 0·2% proof stress at room temp.) which result in a continuously decreasing creep rate and such a low total elongation that it may be presumed that fracture will not occur. The smaller the product of the elongation and the rate of elongation, the lower the probability of fracture. It is shown that a circumferential notch increases the strength and creep-resistance and lowers the ductility of test-bars. As the number of notches is increased, the strength and creep-resistance fall, while the ductility rises; the original values for the smooth bar are not reached with 11 notches, but the trends recorded apparently continue.—K. S.

***New Aluminium Alloy[s] of the Aluminium–Copper–Silicon Type.** W. Thury and H. Landerl (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1951, 96, (10), 201–205).—A report of a study of the properties of alloys, based on the compn. Cu 2, Si 2, Mg 0·5, Mn 0·4%, balance Al (except for the usual 0·4% Fe present in such commercial alloys), suitable for the prodn. of simple castings, and wrought forms such as sheet, strip, and forgings. *Cast Alloy*.—This should contain more Si than Cu, the optimum contents being 2·4 and 1·5%, resp. It has a U.T.S. of 13 kg./mm.^{2}, an elongation of 1%, and a B.H. of 70 kg./mm.^{2}, which after soln.-treatment at 500° C., quenching in water, and artificial ageing at 160° C. for 12 hr. become 24 kg./mm.^{2}, 0·3%, and 115 kg./mm.^{2}, resp. *Wrought Form*.—The most suitable compn. for working contains Si 1·5 and Cu 2·2%, with or without addn. of ~0·1% Ti or Cr; this alloy has liquidus and solidus temp. of 641·5° and 508° C., resp. Hot rolling can be carried out at 450°–500° C., soft annealing at 350° C. The U.T.S. (in kg./mm.^{2}), P.S. (in kg./mm.^{2}), elongation (1%), B.H. (in kg./mm.^{2}), and elect. conductivity (m./Ω.mm.^{2}), of (*a*) soft annealed material are 16·5, 11·0, 20, 41, and 28, resp., and of (*b*) pptn.-hardened material, 38·5, 35·0, 13, 95, and 23, resp. The metallography of the alloys is described, and illustrated by photomicrographs; the following phases (which vary in number and amount present, according to the condition of the material) have been identified: Al solid soln., Al_2Cu, Si, *P* (Al–Fe–Si–Cu–Mn–Mg), *Q* ($Al_4CuMg_5Si_4$ or $Al_4CuMg_4Si_4$), and *U* (Al–Mn–Fe), of which *Q* is the hardest and exerts the greatest influence on the workability of the material. 14 ref.—E. N.

***Solid Solutions and Grain Boundaries [in Aluminium–Zinc Alloys]: Progress Report.** B. L. Averbach *et al.* (*U.S. Atomic Energy Commission Publ.*, 1952, (NYO-583), 3 pp.).—It was found that e.m.f. data on Al–Zn alloys of varying compn. obtained from measurements in a molten salt electrolyte contg. $AlCl_3$ and NaCl, could be represented satisfactorily by linear equations. Preliminary calculations indicate that: (1) the thermodynamic activity of Al exhibits a large positive deviation from Raoult's law, (2) the relative partial molar entropy of Al in the α-solid soln. is close to the ideal, and (3) the relative integral enthalpy has a max. of ~700 cal./mole at 40 at.-% Al. The large positive deviation in Al activity is to be expected from the miscibility gap in the solid soln., and the approach to an ideal soln. can be partly accounted for by the similarity in at. vol. of Al and Zn. Extrapolation of the data to the solidus temp. gives a value of 440 cal./mole for the integral excess free energy of the liq. alloys relative to liq. Al. A marked low-angle scattering of X-rays has been found in Al–Zn alloys at 400° C., indicating the presence of short-range ordering.—B. W. M.

***Microconstituents in Chromium-Base Chromium–Iron–Molybdenum Alloys, and Their Behaviour with Heat-Treatment.** Joseph P. Hammond, Arthur B. Westerman, and Howard C. Cross (*J. Metals*, 1952, 4, (12), 1329–1342).—Metallography, three X-ray methods, and hardness measurements have been used to study the phases occurring in Cr-base, heat-resistant alloys (contg. <0·005–0·09% C) of the compn. Cr 60–Fe 15 and 25–Mo 25 and 15, Cr 50–Fe 50, Cr 40–Fe 40–Mo 20, Cr 55–Fe 25–Mo 20, and Cr 58–Fe 15 and 25–Mo 25 and 15–Ti 2 wt.-%. The alloys were prepared from commercial-purity metals, by melting and casting *in vacuo*. In addn. to the b.c.c. α matrix, the following phases have been found (which are readily distinguishable metallographically): (1) f.c.c. $M_{23}C_6$ type—indicated to be $(Cr, Mo)_{23}C_6$, by X-ray data; (2) σ—isomorphic with the FeCr and FeMo intermetallic compounds; and (3) *Z*—of unidentified crystal structure. The behaviour of these phases is discussed for alloys in the as-cast state and after ageing for up to 144 hr. at 1400°–2000° F. (760°–1095° C.). When Ti and N were present in the alloys, the following additional phases occurred: (*a*) f.c.c. TiC, (*b*) constituents consisting of a continuous series of solid soln. between the TiC and TiN phases, and (*c*) a hexagonal Cr_2N-type phase. Homogenization before ageing increased the hardness of the alloys; the higher hardness of homogenized and aged alloys, as compared with as-cast and aged alloys, is due to the more uniform distribution of the finely dispersed constituent which, for homogenized Cr 60–Fe 15–25–Mo 25–15%-type alloys contg. 0·015–0·025% C, is the $M_{23}C_6$-type carbide. Twinning was found adjacent to diamond hardness indentations made at room temp. in alloys free from finely dispersed phases. The X-ray and metallographic techniques which were used are described in detail; photomicrographs illustrate the typical alloy structures encountered. 10 ref.—E. N.

***Copper–Zinc Constitution Diagram, Redetermined in the Vicinity of the Beta Phase by Means of Quantitative Metallography.** Lilian Heikkinen Beck and Cyril Stanley Smith (*J. Metals*, 1952, 4, (10), 1079–1083).—A careful re-examination has been made of the phase boundaries in the vicinity of the β part of the Cu–Zn diagram, at temp. between 700° and ~250° C., in order to determine whether a two-phase region exists between the ordered and disordered forms of the β phase. The alloys were prepared from re-electrolysed cathode Cu and Horsehead 99·99% Zn, and the impurities in the samples before examination were found to average: Ag 0·0002, Fe 0·004, Si 0·0008, Cd 0·003, Mg 0·0002, and Pb 0·0002%. The methods of heat-treatment used in the prepn. of the specimens before examination are described (temp. were accurate to ± 2° C.); evaluation of the nature and the amounts of the various phases present was carried out by quant. metallography for temp. >400° C., and by precision lattice-parameter measurements for lower temp. The results are plotted in the form of a constitutional diagram for the range 40–65 wt.-% Cu. No new features are observed, and the lines as redetermined are in close agreement with those chosen by Raynor (*Inst. Metals*: *Annotated Equilib. Diagr. Series* No. 3, 1944) as best representing previous data, except for the Zn-rich boundary of the β phase in the vicinity of the disorder point. This lies at 48·85 wt.-% Zn, compared with R.'s point *I*, which is at 49·7% Zn. The width of the β phase at 454° C. becomes only 3·5% instead of 4·9%. The other fixed points are 38·9, 45·4, and 57·7% Zn for R.'s points *G*, *H*, and *J*, which he gives as 39·0, 45·0, and 57·7%, Zn resp. There is no evidence for a two-phase field between the β phase in the ordered and disordered conditions for, although changes of slope are observed in the α, γ, and Zn-rich β boundaries at the dis-

ordering temp., there is no offset, and there is neither change of slope nor discontinuity in the Cu-rich boundary of the β phase. The max. extent of the α-phase decreases from 49% Zn at 450° C. to ~35·5% at 250° C. The β-phase field lies entirely on the Cu-rich side of CuZn below 600° C., and though it tends in this direction at temp. below 480° C., it never attains the stoichiometric compn.; it also retains approx. the same width at all temp. below the transition. 9 ref.
—E. N.

***On the Reversal of the Strain-Induced Martensitic Transformation in the Copper–Zinc System.** J. E. Reynolds, Jr., and M. B. Bever (*J. Metals*, 1952, **4**, (10), 1065–1066).—Specimens of Muntz metal (Cu 60·12, Pb 0·15, Fe 0·02, and Zn (by difference) 39·71%) were soln.-treated at 870° C. for 45 min. and quenched in a brine–ice mixture. After mech. polishing the specimens were polished and etched electrolytically in dil. H_3PO_4. The structure consisted almost entirely of β with slight pptn. of α in the grain boundaries. When the specimens were deformed (within the plastic region) by compression in a vice in a direction ‖ the polished surface, long narrow plates of martensite—the f.c. tetragonal transition lattice—were formed, which grew smoothly or in steps according as to whether the stress was increased slowly and continuously, or in steps, resp.; the plates which had a $(155)_\beta$ habit plane tended to disappear when the applied force ceased to act, although in the areas where the plates disappeared, the surface showed a slight residual disturbance. It is concluded that martensite, strain-induced in metastable β brass at room temp., is mech. reversible with considerable hysteresis and that elastic stresses are operative in this reversal. 3 ref.
—E. N.

***On the Anomaly of the Specific Heat at High Temperatures in α-Phase Alloys of Copper and Zinc.** Hakaru Masumoto, Hideo Saito, and Makoto Sugihara (*Nippon Kinzoku Gakkai-Si* (*J. Japan Inst. Metals*), 1952, **16**, (7), 359–361).—[In Japanese, with tables and diagrams in English]. The sp. heat of 7 alloys in the α solid-soln. range of the Cu–Zn system (4·56–36·87% Zn) have been measured in the annealed and quenched conditions by an inverse-rate curve method. The sp. heat/temp. curves show a slight anomaly at ~200°–260° C., the anomaly being most marked in the alloy contg. 30-35% Zn. Curves of alloys contg. <20·75% Zn show a kink at a const. temp. (~260° C.), and with increasing Zn content a kink occurs at decreasing temp. This anomaly can be prevented by quenching the alloys from 500° C. The authors conclude that it may be caused by the existence of the superstructure Cu_3Zn in a short range of Zn content and not by a change in solubility of Zn in the α solid soln.
—AUTHORS.

***Relative Diffusion Rates of Zinc and Copper in Alpha Brass.** R. W. Balluffi and B. H. Alexander (*J. Metals*, 1952, **4**, (12), 1315–1316).—The relative diffusion rates of Cu and Zn in α brass have been measured in a vapour/solid type of couple, with which the amounts of Cu and Zn diffusing past inert Kirkendall markers can be determined by means of a combination of dilatometric measurements, metallographic examination, and chem. analysis. The ratio of diffusivities, D_{Zn}/D_{Cu}, characteristic of the average compn. (27 at.-% Zn) at the markers during the run, was found to be: (1) single-crystal brass—5·2 and 5·8 at 850° and 890° C., resp., and (2) polycryst. brass—5·7 at 890° C. 2 ref.—E. N.

***The Effect of Four Transition Metals on the α/β Brass Type of Equilibrium.** J. B. Haworth and W. Hume-Rothery (*Phil. Mag.*, 1952, [vii], **43**, (341), 613–631).—The effects of Mn, Fe, Co, and Ni on the α/(α + β) and (α + β)/β phase boundaries of Cu–Al and Cu–Zn alloys at 672° C. have been investigated by microscopical examination. On the assumption that the boundaries follow a const. electron concentration, the results would indicate effective valencies of 0·6, 0·8, 1·0, and 1·9 for Ni, Co, Fe, and Mn, resp. The phase boundaries in the Cu–Al–Ni system are not parallel and neither indicates the valency of 0·6 for Ni deduced from Cu–Zn–Ni alloys. The apparent valencies approx. equal the numbers of *s* electrons/atom deduced from other considerations, though it is not understood why this should be the case. A comparison is made between the general behaviour of the transition metals in brasses with that in other alloy phases.
—P. C. L. P.

***Effect of Certain Impurities on Free-Cutting Brass.** Grayson B. Wood, Jr. (*Wire and Wire Products*, 1952, **27**, (10), 1027–1029, 1128–1130).—Economic conditions having necessitated the use of low-grade Cu, the effects of such impurities as Fe, Ni, Sn, and Al have been studied. Nominal free-cutting brass (Cu 62, Pb 3, Zn 35%) was examined for machinability and ductility. Fe up to 0·35% was found to have little effect, save on cutting tools, which required more frequent sharpening, and on the annealing cycle. Al up to 0·20% had no effect on machinability, but grain growth was retarded with increasing Al content. During high-temp. salt-bath anneals, some blistering occurred, an effect produced also by the presence of Sb.—C. P. F.

[Discussion on a Paper by N. Grossman:] The Effect of Shot-Peening on Damage Caused [to Steel and Brass] by Cavitation. —— (*Amer. Soc. Test. Mat. Bull.*, 1952, (**186**), 46).—See *M.A.*, **20**, 240.

***An X-Ray Diffraction Study of the Hafnium–Hydrogen System.** S. S. Sidhu and J. C. McGuire (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECU-2042**), 17 pp.; also *J. Applied Physics*, 1952, **23**, (11), 1257–1261).—The lattice parameters of the Hf metal are $a = 3{\cdot}200$ and $c = 5{\cdot}061$ Å., giving a calculated *d* of 13·1, compared with a measured *d* of 12·9 g./c.c. Three Hf–H phases were identified at room temp.: (*a*) deformed f.c.c. phase with $a = 4{\cdot}702 \pm 0{\cdot}012$ and $c = 4{\cdot}678 \pm 0{\cdot}012$ Å., which co-exists with Hf up to a compn. of $HfH_{1{\cdot}53}$; (*b*) f.c.c. phase, with $a = 4{\cdot}708 \pm 0{\cdot}002$ Å., which is stable between $HfH_{1{\cdot}53}$ and $HfH_{1{\cdot}70}$; and (*c*) f.c. tetragonal, with $a = 4{\cdot}882 \pm 0{\cdot}002$ and $c = 4{\cdot}384 \pm 0{\cdot}002$ Å., which forms between $HfH_{1{\cdot}80}$ and $HfH_{1{\cdot}87}$ and is stable up to $HfH_{1{\cdot}98}$. The *d* of the third phase is 11·48 g./c.c. and it has $2HfH_2$ mol./unit cell with the Hf atoms at 0 0 0 and $\frac{1}{2}\,\frac{1}{2}\,\frac{1}{2}$ and H atoms at $0\,\frac{1}{2}\,\frac{1}{4}$, $0\,\frac{1}{2}\,\frac{3}{4}$, $\frac{1}{2}\,0\,\frac{1}{4}$, and $\frac{1}{2}\,0\,\frac{3}{4}$. The transformation of the tetragonal cell into the f.c.c. form at $HfH_{1{\cdot}8}$ produces only a slight contraction, indicating that it may be a second-order transformation.—B. W. M.

***Supraconducting Properties of Indium–Thallium Alloys.** J. W. Stout and Lester Guttman ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 51–60).—By an X-ray examination of the supposedly continuous solid-soln. range of In–Tl alloys contg. up to 40 at.-% Tl, it has now been shown that a transformation from a f.c. tetragonal to a f.c.c. structure occurs. The phase diagram found by G. for the whole range of alloys is given; the alloys used in the present study, contg. 0–20% Tl, are all f.c. tetragonal at low temp. Results show that in In–Tl alloys consisting of solid soln. contg. up to 20 at.-% Tl, the Meissner effect is large, and the resistance transition in zero field is sharp. In these respects the alloys behave like pure metals. For the In–5% Tl alloy the penetration of flux, on the appn. of a magnetic field, occurs at a sharply defined field, and restoration of resistance occurs at about the same field strength. As the Tl content increases to 20%, penetration of the magnetic field occurs over a gradually increasing range, and the ratio of the field corresponding to the appearance of resistance to that for the beginning of flux penetration continually increases.
—J. S. G. T.

***Increase of High Magnetostriction [of Iron–Aluminium and Cobalt–Iron Alloys] by Magnetic Anneal.** H. E. Stauss and G. Sandoz (*J. Metals*, 1952, **4**, (12), 1342–1343).—A 87 : 13 Fe–Al alloy and a 69 : 31 Co–Fe alloy—Curie points 560° and 890° C., resp.—were heated to 700° and 900° C., resp., and then cooled in the presence and absence of a magnetic field, cooling taking place, in the former case, at ~100° C./hr. down to 100° C., after which the specimens were quenched, and, in the latter case, cooled in the furnace so that ordering was induced. The effect of the " magnetic anneal " was to cause an appreciable increase in linear magnetostriction over that obtained with normally annealed specimens; the increase was, however, only by a factor of 1·2 instead of the 1·5 anticipated from theoretical considerations. 8 ref.
—E. N.

***Investigation of Supraconductivity in Lead Compounds, Gold Alloys, and Molybdenum Carbide.** R. P. Hudson and K. Lark-Horovitz ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 61–63).—Darby, Hatton, and Rollin (*Proc. Phys. Soc.*, 1950, [A], **63**, 1181; *M.A.*, **18**, 598) found PbS, PbSe, and PbTe all supraconducting at $\sim$5° K.; PbS and PbTe are shown not to be supraconducting, and D., H., and R.'s finding is attributed to the presence of excess Pb in their specimens. Au–Pb alloys contg. resp. 60, 70, or 80% Au and a Au–Pt alloy contg. 85% Au are found not to become supraconducting down to 1·3° K., and so contradict Fröhlich's suggestion that Au–Pd and Au–Pt alloys contg. >60 at.-% Au might become supraconducting (*Phys. Rev.*, 1950, [ii], **79**, 845; *M.A.*, **18**, 502). Contrary to Meissner's finding (*Z. Physik*, 1930, **65**, 30; *Met. Abs.* (*J. Inst. Metals*), 1931, **47**, 15) pure Mo_2C is not a supraconductor; Meissner's result is attributed to contamination of Mo_2C with MoC.—J. S. G. T.

Rare-Earth Metals Improve Elevated-Temperature Properties of Magnesium Castings. J. C. McDonald (*Materials and Methods*, 1952, **36**, (1), 162, 164–166).—The compn., heat-treatment, and mech. properties of Mg alloys contg. 3% rare-earth metals as Mischmetall, together with 0·25–0·75% Zr, are discussed. Zr is added to refine the grain-size and to prevent cracking during freezing. The addn. of 3% Th and 2% didymium instead of Mischmetall, gave alloys with increased high-temp. properties. It is claimed that creep strengths of 8000 lb./in.^{2} at 400° F. (205° C.) can be developed in the new alloys, compared with 1500 lb./in.^{2} for the conventional Mg–Al–Zn type. The 3% Th–0·7% Zr alloy has a creep strength of 3100 lb./in.^{2} at 600° F. (315° C.), as compared with 1200 lb./in.^{2} for Mg alloy contg. 3% Mischmetall and 0·55% Zr, both alloys being in the heat-treated condition.—R. P. H. F.

***Thermodynamics of Crystalline Solutions [Magnesium–Cadmium Alloys]: Progress Report.** W. E. Wallace, R. S. Craig, L. W. Coffer, D. A. Edwards, and R. A. Flinn (*U.S. Atomic Energy Commission Publ.*, 1952, (**NYO-949**), 9 pp.).—The adiabatic compressibility of a 49·4 : 50·6 Mg–Cd alloy was found by the velocity-of-sound method to be $2{\cdot}11 \times 10^{-6}$ atm.$^{-1}$, compared with Bridgman's values of $2{\cdot}07 \times 10^{-6}$ for Cd (*Proc. Amer. Acad. Arts Sci.*, 1945, **76**, 9; *M.A.*, **13**, 204) and $2{\cdot}96 \times 10^{-6}$ for Mg (*ibid.*, 1923, **58**, 166; *J. Inst. Metals* (Abstracts), 1923, **30**, 442). An interpretation of the asymmetry in the curve showing heat of formation/compn. for Mg–Cd alloys is given in terms of the Brillouin-zone structures of the pure metals, and this is extended to account for the markedly different crystallographic changes in the development of the ordered $MgCd_3$ and Mg_3Cd structures.
—B. W. M.

***Supraconductivity in MgTl and LiBi.** Lester Guttman and J. W. Stout ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 65–67).—The intermetallic compounds MgTl and α-LiBi are shown to be supraconductors. For the former, the critical field, H_c, is linearly related to T^2 (T being measured in °K.) and is given by $H_c = 220{\cdot}0[1 - (T/2{\cdot}745)^2]$. The zero-field transition for α-LiBi is estimated to be at 2·47° K.
—J. S. G. T.

***A Study of the System Manganese–Nickel.** N. N. Kurnakov and M. Ya. Troneva (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (1), 73–76).—[In Russian]. A study of the system Mn–Ni was made by microstructural analysis, measurement of elect. resistance, and dilatometry. The alloys were prepared from electrolytic Mn and Ni in alumina crucibles in an induction furnace, the crucible and couple shields being protected from contamination by Mn with a coating of ZrO_2. A Kurnakov pyrometer was used in the thermal analysis. The curve for pure Mn gave three arrests at 1180°, 1080°, and 750° C., corresponding to the $\delta \rightarrow \gamma$, $\gamma \rightarrow \beta$, and $\beta \rightarrow \alpha$ transformations. Alloys with 1·12, 3·03, and 3·29% Ni gave two arrests, corresponding to $\delta \rightarrow \gamma$ and $\gamma \rightarrow \beta$ transformations. Those with 3·62–11·8% Ni gave only one arrest, corresponding to the $\gamma \rightarrow \beta$ transformation. The $\delta \rightarrow \gamma$ transition temp. increased with increasing Ni content of the alloy. At 14·9–31·24% Ni there is no transformation in the solid state. Alloys ranging from 45·95 to 52·24% Ni had two arrests, corresponding to $\gamma \rightarrow \beta$, $\beta \rightarrow \gamma + \varepsilon$ and $\gamma \rightarrow \beta$, $\beta \rightarrow \varepsilon$. Alloys with 53·35 and 57% Ni gave one arrest. No arrests were detected in the alloy with 62·24% Ni. The elect. resistance of annealed specimens was measured, using a Thomson bridge. This revealed the compound NiMn at 50·74% Ni and the compound Ni_3Mn at 75% Ni. The coeff. of elect. resistance, dilatometric and microstructural results, confirmed the existence and boundaries of these phases.—Z. S. B.

***Intermediate Phases in the Mo–Fe–Co, Mo–Fe–Ni, and Mo–Ni–Co Ternary Systems.** D. K. Das, S. P. Rideout, and Paul A. Beck (*J. Metals*, 1952, **4**, (10), 1071–1075).—A report of a study of the 1200° C. isothermal sections of the phase diagrams of these systems, the experimental procedures being identical with those used in previous work (cf. R., Manly, Kamen, Lement, and B., *Trans. Amer. Inst. Min. Met. Eng.*, 1951, **191**, 872; see *M.A.*, **19**, 371). The experimental results show that the μ phase: (1) forms narrow elongated fields in all three diagrams, (*a*) completely traversing that for Mo–Fe–Co from (Fe, Mo) μ to (Co, Mo) μ, and (*b*) extending approx. half-way across those for Mo–Fe–Ni and Mo–Ni–Co; (2) coexists directly with δ in the Mo–Ni–Co system; and (3) and does not coexist with the δ phase in the Mo–Fe–Ni system, for the *P* phase is interposed between them. The *P* phase, of unknown cryst. structure, is isomorphous with, and behaves in a similar manner to, the Mo–Ni–Cr ternary phase of the same name; even though in both systems the compn. range of the *P* phase field is relatively small, it is possible that in the quaternary Mo–Cr–Ni–Fe system an elongated *P* phase field extends from (Mo, Cr, Ni)*P* to (Mo, Ni, Fe)*P*. If the electron vacancy numbers assigned to Mo, Fe, Co, and Ni are 4·66, 2·2, 1·71, and 1·6, resp., the elongated μ phase fields in the ternary systems correspond quite well to a const. electron vacancy range of $\sim$2·95 to 3·25. This type of phase, like σ, whose boundaries are largely determined by electron vacancy values, might be termed "electron vacancy phase", by analogy with the term "electron compound". It is probable that there are still other phases of this type among the intermediate phases in alloys of the transition elements. 23 ref.—E. N.

***Creep of [Nickel-Base] Alloys: Progress Report.** E. R. Parker and T. H. Hazlett (*U.S. Atomic Energy Commission Publ.*, 1952, (**COO-61**), 4 pp.).—The U.T.S. and lattice parameter of Ni are relatively unaffected by addn. of Co up to $\sim$9 at.-%, but both are increased regularly by Fe addn. Results to date indicate that the *d*-level electron interactions have very slight influence on the plastic properties. The shape of the creep curve for solid-soln. alloys of Ni is given by $\varepsilon = \varepsilon_0 + At^b$, where ε = total true strain, ε_0 = instantaneous true strain at zero time, and t = time. The values of b increase linearly with at.-% alloy addn. and log A decreases linearly with alloy content. The creep curve for an alloy with 0·5 at.-% W is given by $\varepsilon - \varepsilon_0 = Ce^{-Q/RT}t^b\sigma^n$ in the range 650°–800° C.—B. W. M.

***Thermodynamic Properties of Solid Nickel–Gold Alloys.** L. L. Seigle, Morris Cohen, and B. L. Averbach (*J. Metals*, **4**, (12), 1320–1327).—The free energies, enthalpies, and entropies of solid Ni–Au alloys contg. Ni 5–95 at.-%, have been determined by potential measurements at temp. of 700°–900° C. An electrolytic cell was used—similar to that of Weibke and Matthes (*Z. Elektrochem.*, 1941, **47**, 421; *M.A.*, **9**, 325)—in which the electrodes were solid Ni and Ni–Au alloys, and the electrolyte was a fused equimolar KCl–NaCl mixture (m.p. 660° C.) contg. $NiCl_2$ 0·2 and NH_4Cl 0·6 mole-%. In conformity with the phase diagram, which contains a miscibility gap, the thermodynamic activities indicate a large positive deviation from Raoult's law. The enthalpies of mixing are positive (heat is absorbed) and are attributable to the lattice distortion resulting from the size difference between the Ni and Au atoms; this factor, rather than the relative bond energies between like and unlike pairs of atoms, appears to be responsible for the unmixing at lower temp. The entropies of mixing of Ni–Au solid soln. are almost twice those of ideal soln.; this large entropy is probably due to an increase in heat capacity on mixing. 25 ref.—E. N.

***Detection of Microwave Signals by NbN Supraconductors in the Transition Region.** J. V. Lebacqz and M. G. Bodmer ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, **(519)**, 109–113).—The detection of microwaves of frequencies between 9000 and 10,000 Mc./s. by NbN bolometers was studied; equipment used, results obtained, and theoretical deductions are briefly reviewed.—J. S. G. T.

***The Mechanism of Energy Transfer and Reaction Rate at Metallic [Platinum–Copper Alloy] Surfaces.** K. Schäfer (*Z. Elektrochem.*, 1952, **56**, (4), 398–403).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. The thermal accommodation coeff. of ethane at 0°–105° C. and the catalytic thermal decompn. of the gas, in accordance with the equation $C_2H_6 = C_2H_4 + H_2$, at temp., up to ~600° C., at pressures from 10^{-3}–10^{-2} Torr (1 mm. Hg/cm.^{2}) on filaments of the Pt–Cu series of alloys were studied. The accommodation results are interpreted in terms of degrees of freedom of translatory–rotatory molecular reactions and indicate considerable parallelism between the accommodation coeff. and the catalytic decompn. of ethane in ethylene and H.—J. S. G. T.

***Phase Changes in Silver–Tin Amalgams.** J. C. Moffett, G. Ryge, and A. G. Barkow (*J. Appl. Physics*, 1952, **23**, (10), 1188–1189).—A letter. X-ray-diffraction spectra were recorded at intervals on a Geiger counter spectrometer from amalgams between Ag_3Sn and Hg, the first spectrum being obtained 10 min. only after trituration. γ_1 and γ_2 phases (Ag_2Hg_3 and Sn_7Hg, resp.) were already present at this stage, and study of successive spectra suggested that they were formed by independent reactions. No evidence of β (Ag–Hg) phase was found. The observations settle a disagreement about the sequence of reactions. Well-formed crystals of the various phases can be grown by amalgamating polished specimens of Ag–Sn alloys, and some have been used for single-crystal analysis.—R. W. C.

***Properties of Metallic Solutions. IV.—Surface Tension of Tin and Tin–Sodium Alloys.** (Pokrovsky and Galanina). See col. 551.

***Informal Progress Report: [Dissociation Pressure of UH_3 and the Effect of Nitrogen on Ti–H Equilibrium].** Thomas R. P. Gibb (*U.S. Atomic Energy Commission Publ.*, 1950, **(NEPA-1419)**, 5 pp.).—The dissociation pressure of UH_3 was found to be higher at temp. <350° C. than that reported by Spedding, and the extrapolated value at room temp. was ~1 mm. The effect of N on the Ti–H equilibrium was found to depend on various factors, including the surface condition of the metal and the diffusion rate.—B. W. M.

***Development of Zirconium-Base Alloys: Progress Report.** J. H. Keeler (*U.S. Atomic Energy Commission Publ.*, 1952, **(SO-2506)**, 30 pp.).—Ingots prepared from 99·77% Zr in an improved multiple-hearth arc-melting furnace were of greater workability and lower hardness and U.T.S. than those previously made. These results indicate a lower contamination of the ingots by O and N during melting. Pole figures were determined after cold rolling and annealing at 600° and 900° C. Tensile properties were determined in the temp. range 200°–370° C. for unalloyed Zr and for binary alloys with Cr, Al, and Ta. Heavily cold-worked crystal bar gave unusual results characterized by an early max. load at low strain followed by a continuously decreasing load, which was not associated with any obvious localized necking. Necking finally occurred just before fracture. This material showed markedly increasing rate sensitivity in the temp. range 25°–370° C., whereas the annealed material showed no increase in the same range. 8 ref.—B. W. M.

***Phase Diagrams of Zirconium-Base Binary Alloys: Progress Report.** D. J. McPherson and M. Hansen (*U.S. Atomic Energy Commission Publ.*, 1951, **(COO-30)**, 22 pp.).—The m.p. of iodide Zr, as determined in a high-temp. vacuum induction furnace of the type described by Schramm, Gordon, and Kaufmann (*Trans. Amer. Inst. Min. Met. Eng.*, 1950, **188**, 195; *M.A.*, **17**, 570), was 1852° ± 10° C. A tentative equilibrium diagram is given for Zr–Sn alloys up to 28 wt.-% Sn. An alloy with 46·74% Sn was single-phase, and this would indicate a compound in the Zr_5Sn_3–Zr_3Sn_2 region. A eutectic between this compound and Zr occurs at ~23·5% Sn and a second compound is formed peritectoidally at ~25% Sn. In Mo–Zr alloys, only one compound was identified and was considered to be $ZrMo_2$. A eutectic between Zr and Zr_2Cu occurred at 995° C. and 18% Cu.—B. W. M.

***Phase Diagrams of Zirconium-Base Binary Alloys: Progress Report.** D. J. McPherson and M. Hansen (*U.S. Atomic Energy Commission Publ.*, 1951, **(COO-31)**, 31 pp.).—Tentative phase diagrams are presented for Zr–Sn alloys up to 48 wt.-% Sn, Zr–Mo, and Zr–Cu alloys. A compound was found to exist at 24·2% Sn, corresponding to Zr_4Sn. Mo lowered the m.p. of Zr to a min. at ~30% Mo and ~1525° C.; beyond this the m.p. of the alloys increased steadily up to the m.p. of Mo. A eutectoid was formed at 10% Mo and 750° C., and the solid solubility of Mo in α-Zr was <0·5% at 700° C. and in β-Zr, >12% at 1200° C. Nearly single-phase structures were obtained at compn. corresponding to Zr_2Cu, ZrCu, Zr_2Cu_3, and $ZrCu_3$, and eutectics at 20, 37, 47, 53, and 90% Cu. The solubility of Cu in α- and β-Zr was <5 and <1%, resp. In Zr–Cr alloys, eutectics occurred at 18 and 68% Cr and a compound was identified at 47% Cr, believed to be Zr_2Cr_3 or $ZrCr_2$. The solubility of Si in β-Zr was ~0·2%, and a eutectic was formed between Zr and the first compound at ~3% Si. The solubility of Al in β-Zr was ~7% and three compounds were identified: Zr_3Al_4 (or Zr_2Al_3), $ZrAl_3$, and one of intermediate compn.—B. W. M.

***Phase Diagrams of Zirconium-Base Binary Alloys: Progress Report.** D. J. McPherson and M. Hansen (*U.S. Atomic Energy Commission Publ.*, 1951, **(COO-32)**, 42 pp.).—Zr_4Sn was found to be f.c. tetragonal with $a = 10{\cdot}963$ and $c = 9{\cdot}853$ Å., and to form a peritectoid reaction with the β-solid soln. at 1090° ± 10° C. The eutectic between Zr and $ZrMo_2$ was placed at 31 wt.-% Mo. The solid solubility of Mo in β-Zr decreases from 21% at 1480° C. to 10% at 800° C. and the β-solid soln. decomposes eutectoidally at ~9% Mo and 760° C.; the max. solubility in α-Zr is <0·5%. The solid soln. of Cu in β-Zr decreases from ~4% at 1000° C. to 2% at 850° C., and decomposes eutectoidally between 1 and 2% Cu at ~820° C. The solubility of Cu in α-Zr is <0·2% at all temp. A preliminary survey of the Zr–W system indicated a close similarity to that of Zr–Mo at the Zr-rich end. The only compound identified in Zr–Cr alloys was $ZrCr_2$, which formed eutectics with Zr and Cr at ~18 and 70% Cr, resp. The solid solubility of Cr in β-Zr at 1300° C. was ~6%. Evidence was obtained for the formation of five compounds in Zr–Al alloys, as follows: Zr_3Al_2 (or Zr_5Al_3), Zr_4Al_3, Zr_2Al_3 (or Zr_3Al_4), $ZrAl_2$, and $ZrAl_3$. The solid solubility of Al in β-Zr extended to ~9%, and a eutectic was formed at ~11% Al. Attempts to make Zr–Mg alloys in the A-arc furnace were unsuccessful.—B. W. M.

***The Change in Primary Solubility of Metals in the Solid State Under the Influence of Pressure.** M. I. Zakharova (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (1) 69–71).—[In Russian]. The solubilities of Cu in Al, Zn in Cu, and Al in Cu were investigated at a pressure of 10,000 atm. and in the temp. range 400°–800° C. The pressure was applied to specimens 6 mm. in dia. by a Brinell machine. Annealing times varied from 2 to 30 hr., according to the time taken to reach equilibrium. Equilibrium was assumed to have been reached when the lattice parameter remained const. with increasing annealing time. The thermodynamically derived equation is, $dL/dp : dL/dT = \frac{\Delta V_T}{H}$, where L is the solubility, p is the pressure, T is the temp., ΔV is the change in vol. by the soln. of 1 g.-mol. in the equilibrium state, and H is the heat of soln. The solubility of Cu in Al and Zn in Cu decreased with increasing pressure; that of Al in Cu increased. At atmospheric pressure and 500° C. the alloy contg. 4% Cu in Al appeared to be single-phase; at 10,000 kg./cm.^{2}, this alloy should be two-phase. Microscopical examination after annealing for 2 hr. at 500° C. shows that this is so.—Z. S. B.

***Mechanical Properties of Intermetallic Compounds at Elevated Temperatures.** Robert Lowrie (*J. Metals*, 1952, **4**, (10), 1093–1100).—Cylindrical test-pieces of the intermetallic

compounds CrSb, Cu_5Ca, Cu_2Mg I and Cu_2Mg II, Cu_3P, Cu_3Si, $Ni_3Si + Ni_5Si_2$, Ni_2Si, Al_7Cr, Al_4Cr, $Cu_{17}Al_9$, and Ni_3Si—of which the last four are incongruently melting compounds—were prepared by elect. furnace melting of the constituents (of commercial purity), under a neutral atmosphere of N or He, and centrifugally cast to shape in special sand moulds. The surface finish of the specimens was, in general, quite good, any irregularities being shallow and of relatively large radius of curvature. Tensile testing at elevated temp. was then carried out under a neutral atmosphere, in a specially designed furnace, the specimens being presoaked at temp. for 30 min. The results are shown in tables and graphs, particular attention being paid to the relationship of U.T.S. and elongation with "homologous temp.", T/T_m, where T is the testing temp., and T_m the m.p. (or for incongruently melting compounds, the decompn. temp.), both in °K., and with the "critical homologous temp.", which is defined as that value of T/T_m at which the elongation begins to increase rapidly from negligible values—which generally occurs at ~0·5% elongation. CrSb and Cu_3Si were so fragile that they could not be assembled in the grips, while Al_7Cr and Al_4Cr had low strength and very limited ductility even at high-m.p. fractions. The remaining alloys exhibited quite similar behaviour one with another, in that all had critical T/T_m values; these were generally in the range 0·61–0·68, the exceptions being Cu_5Ca with a slightly higher value of 0·73, and Cu_3P with a surprising low value of 0·35. No correspondence of U.T.S. and T/T_m values was observed other than with the compounds Cu_5Ca and Cu_2Mg, which are composed of metals of similar valence electron configurations, and which possess predominantly metallic binding; these two compounds show, also, close correspondence of critical T/T_m values. U.T.S. and the heat of formation, *d*, or relative decrease in vol. upon formation of the compounds could not be correlated, from which it appears that local conditions favourable to slip or fracture are more important than the overall energy of formation of the compound. Cu_2Mg I, Cu_2Mg II, and Ni_3Si broke with an intercrystn. fracture; the remaining compounds showed transcryst. fractures. The results show that some factor exists which is related to the "degree of brittleness" of intermetallic compounds, and which determines whether a given compound will become ductile at any temp. It is considered that this factor is intimately related to the crystal structure, and further work is proposed on single crystals of intermetallic compounds for its elucidation. A relationship between U.T.S. and crystal structure also appears probable. 30 ref.
—E. N.

***New Supraconducting Compounds.** B. T. Matthias and J. K. Hulm ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 69).—Known supraconducting compounds of Bi include Au_2Bi, several alkali metal–Bi compounds, RhBi, Bi_3Ni, and $CaBi_3$. $SrBi_3$ becomes supraconducting at ~5·5° K., and $BaBi_3$ at slightly above 6° K. Mo_2B, Mo_2N, and MoN are all supraconductors, the transition temp. of MoN being unusually high at ~12·0° K.—J. S. G. T.

***Heat Transfer in Supraconducting Alloys.** J. K. Hulm ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (519), 37–41).—Magnetic field variation of magnetic induction, *B*, elect. resistivity, ρ, and thermal conductivity, *K*, of an In–10% Tl alloy, at 2·6° K., and the temp. variation of *K* of the same alloy and of pure In (0·1% impurity) in the supraconducting and normal states have been determined. Although nearly all the alloy passes from the supraconducting to the normal state in a field of ~10 Oe., where *B* and *K* rise rapidly with increase of *H*, a very small fraction of the whole bulk of the alloy remains supraconducting at higher field strengths and provides a path of zero or low resistance for the current used in measuring ρ for applied fields of ~220 Oe. The graph of *K* against *H* drops slightly as the field is increased beyond 120 Oe. before a steady value characteristic of the normal state is reached. As small regions with greater than average Tl content would probably be overlooked in the chem. test of homogeneity of the alloy, the presence of such regions is a possibility. The results are discussed in terms of a formula for the thermal conductivity of the alloy, derived from the electron theory of alloys, involving two terms, the first attributed to heat transfer by electrons with scattering by impurities, and the second representing heat transfer by lattice waves with scattering by electrons. It is suggested that supraconducting electrons do not scatter lattice waves.
—J. S. G. T.

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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 Interpretation of Etch Patterns Produced by Etching and Chemical Polishing of Aluminium.** A. Politycki and H. Fischer (*Z. Elektrochem.*, 1952, **56**, (4), 326–330; discussion, 330).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. By means of the electron microscope it is established that the etch-patterns appearing on the surface of Al after etching and chem. polishing are not always associated with a cubic cryst. structure as found by Mahl and Stranski (*Z. physikal. Chem.*, 1942, [B], **51**, 319; *M.A.*, **10**, 141); only when gaseous HCl was used as etching reagent did the anticipated cubic-octahedral structure appear in the etch pattern. P. and F. now find that this latter structure appears in the etch pattern, as required by the theory of the disintegration of the oxide-free metal, when an aq. soln. of HI is used as etching agent.—J. S. G. T.

***Elementary Structure and Slip-Band Formation in Aluminium.** D. Kuhlmann-Wilsdorf, J. H. van der Merwe, and H. Wilsdorf (*Phil. Mag.*, 1952, [vii], **43**, (341), 632–644).—A theoretical discussion is given of the electron-microscope observations of W. and K.-W. (*Naturwiss.*, 1951, **38**, 502, and to be published) on the surfaces of single-crystal and polycryst. high-purity Al, which revealed an extensive fine slip structure. The length and amount of glide of the elementary lines is consistent with the possibility that one freely acting dislocation source is responsible for each. The evidence suggests that the structure is confined to the surface. The spacing of the elementary lines is the same as that of the fine structure in the major slip bands, and agrees with the theoretical critical sepn. for rows of dislocations to pass one another. It is suggested that slip bands arise if an elementary line becomes sufficiently strong to clear a corridor on either side of it of dislocations, so that a nearby source can operate without hindrance. Interaction between dislocations to block one another's motion is treated in an appendix.
—P. C. L. P.

***Dynamic Formation of Slip Bands in Aluminium.** N. K. Chen and R. B. Pond (*J. Metals*, 1952, **4**, (10), 1085–1092).—Single-crystal specimens of high-purity (99·997%) Al were made by the method of gradual solidification from the liq. state, using a special graphite mould of sq. cross-section. They were etched, homogenized for 24 hr. at 600° C., gently polished with 0000 emery paper, re-etched, and electrolytically polished, and their crystallographic orientations determined by X-ray methods. More than 30 of these crystals, of various orientations, were extended in a specially constructed microtensile machine having a microload cell and a microclip gauge; a motion-picture camera–metallurgical microscope set-up made possible the co-ordination of phenomena occurring during plastic deformation with load and strain through time. Propagation of slip bands was observed at finite rates, from zero as a min. to ~7600 μ/sec. as a max., in the direction nearly ⊥ the slip direction; strain-hardening had the effect of retarding the rate of propagation; the rate of slip in the slip direction was not determined. The experiments show, conclusively, that a visible slip band does not suddenly form throughout the entire crystal and, therefore, the classical conception of the slip process—that slip is a uniform shear of a plane of atoms (i.e. the slip plane) in the slip direction—must be modified. A definite modification is proposed—that the unit process of slip is a shear of a row, or rows, of atoms in the slip direction, and the growth phenomenon results from progressive participation in the shearing process by successive rows of atoms. A model for this is presented which, it is shown, can also be interpreted from the general theory of dislocations. Characteristics associated with slip propagation, such as the direction of propagation, the " overlap " of slip bands, and the sequential addn. of slip bands are analysed. From the geometry of the addn. of slip bands, it is concluded that one slip band can extend its hardening effect to its neighbours. 10 ref.—E. N.

***Some Observations of Sub-Grain Formation During Creep in High-Purity Aluminium.** Italo S. Servi, John T. Norton, and Nicholas J. Grant (*J. Metals*, 1952, **4**, (9), 965–971).—Creep tests have been carried out, at temp. of 400°–1200° F. (205°–650° C.), on coarse-grained and single-crystal specimens (which were electropolished) of high-purity (99·995%) Al, and the deformed specimens were examined by metallographic and X-ray back-reflection Laue techniques. It is shown that the deformed specimens exhibit: (1) slip bands, the slip systems being (111)[110], (100)[110], or (211)[110], and (2) sub-grain formation both on the surface and in the interior of the specimens; they have (*a*) an average size which is $\cong$ 1/applied stress, and smaller, by at least a factor of 30, than the average slip-band spacing, and approaches the value predicted by Orowan's theory for the min. slip-band spacing, and (*b*) an orientation which, in the same grain, corresponds approx. to a rotation about an axis lying in the slip plane and ⊥ the slip direction. The results suggest that the slip process is the primary mechanism of deformation of the metallic grains even at elevated temp., when the grain-size is large enough to permit slip to become operative; this confirms the earlier results of other workers. Sub-grain formation is not the primary mechanism of deformation of the grains at elevated temp. These sub-grains form by a polygonization process due to local bending caused by non-homogeneous grain-boundary sliding and concurrent annealing. The markedly different behaviour observed in polycryst. specimens tested at low temp. and at high temp. has its origin in the deformation occurring at the grain boundaries. 20 ref.—E. N.

***An Electron-Microscope Study of the Effect of Temperature and Strain-Rate on the Mechanism of Deformation of Aluminium.** R. I. Garrod, J. W. Suiter, and W. A. Wood (*Phil. Mag.*, 1952, [vii], **43**, (341), 677–685).—Electron-microscope observations have been made on Formvar and oxide replicas prepared from deformed polycryst. specimens of 99·98% purity Al. These observations support the conclusion drawn from X-ray and opt.-microscope work that there are three stages in deformation: the slip, cell, and boundary micro-flow processes, found in that order with increasing temp. and decreasing strain-rate (W., Wilms, and Rachinger, *J. Inst. Metals*, 1951, **79**, 159; *M.A.*, **18**, 660). Observations in the slip stage agree with those of other workers. No slip was detected in the cell stage, and though some abnormal deformation of the grain boundaries themselves was found, the cell boundaries were sharp. No grain breakdown was found in the boundary-flow stage, though markings ∥ the boundary occurred. Brief mention is made of the implications of these experiments with regard to the atomic movements involved.—P. C. L. P.

***Structure and Crystallography of Second-Order Twins in Copper.** M. Sharp and C. G. Dunn (*J. Metals*, 1952, **4**, (12), 1344).—A report of the structure and crystallography of second-order twins formed by secondary recrystn. in the f.c.c. matrix of Cu. The origin of the twins may be accounted for in terms of repeated twinning and special growth characteristics. The boundary between the parent grain and its second-order twin involved a {111} plane of the former and a {115} plane of the latter. Although the same combination of planes is not only possible in first-order twins but actually occurs quite frequently, the atoms near a {111} {115} boundary have a configuration in first-order twins which is different from that obtaining in second-order twins. The prevalence of this type of boundary in both types of twinning, together with the necessary occurrence of a large number of common lattice sites at the boundary, is an indication that this combination produces an " energy cusp " boundary. 4 ref. —E. N.

***Fundamentals of Cold Working and Recrystallization [of a Copper–Silicon Alloy]: Progress Report.** B. L. Averbach, M. Cohen, S. Allen, and P. Fopiano (*U.S. Atomic Energy Commission Publ.*, 1952, (**NYO-862**), 3 pp.).—The effective particle size in a single crystal of 98 : 2 Cu–Si alloy after cold rolling by 50% was determined from the Fourier coeff. which describe the X-ray line broadening. In the $[1\bar{1}1]$, [110], and [002] directions, the average particle size was ~1100, ~500, and probably >2000 Å., resp. It is considered that this size represents the average length of a coherent column of cells in a given direction, i.e. the distance between clusters of slip planes or between collections of dislocations which define a grain boundary of small disorientation. The displacements in the [111], [200], and [220] directions were inversely proportional to the Young's modulus in each direction. In cold-rolled single crystals of Al, the Fourier coeff. were almost independent of the plastic reduction, and the line-broadening in each case was quite low.—B. W. M.

***The Growth and Orientation of Single Crystals of Indium.** A. J. Goss and E. V. Vernon (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 905–906).—A letter. Conditions are described for the growth of In single crystals from the melt using the method of Andrade and Roscoe. Electrolytic polishing baths and etchants are also discussed. For these crystals, the angles between the *c*-axis and the rod axis were all between 50° and 90°.—E. O. H.

***X-Ray Line-Broadening from Cold-Worked Iron.** J. H. Auld and R. I. Garrod (*Nature*, 1952, **169**, (4301), 579–580).—Ingot Fe in the form of wire and filings has been examined in a Debye–Scherrer camera in both the strain-free and cold-worked states, using four monochromatic radiations. From the results it appears that, at least for Fe, lattice distortion predominates over particle size as the cause of line-broadening, and there is no essential difference in the deformation process between filings and stressed wires.—R. S. B.

***The Crystal Structures of NpC and Pu_2O_3.** D. H. Templeton and Carol H. Dauben (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECD-3443**), 6 pp.; also (**UCRL-1886**)).—The structures of NpC and Pu_2O_3 were determined from X-ray powder photographs. NpC is f.c.c. with $a = 5{\cdot}004 \pm 0{\cdot}005$ Å. and is isomorphous with UC and PuC. 6 ref.—B. W. M.

***The Unit Cell and Brillouin Zones of $Ni_4Mn_{11}Al_{60}$ and Related Compounds.** Keith Robinson (*Phil. Mag.*, 1952, [vii], **43**, (342), 775–782).—The compounds $Ni_4Mn_{11}Al_{60}$, $Cu_2Mn_3Al_{20}$, and $ZnMn_5Al_{24}$ have been examined by oscillating-crystal and Weissenberg techniques and found to be orthorhombic. The Ni- and Cu-bearing phases are structurally and dimensionally similar. The first Brillouin zone which can be constructed using planes giving a strong reflection can hold a total of 2·04 electrons, while the inscribed sphere to it corresponds to an electron : atom ratio of 1·74. $ZnMn_5Al_{24}$ has a different crystal structure, with a first Brillouin zone which can contain rather fewer electrons. Prominent outer zones of both structures correspond to too many electrons to be significant. The results support the view expressed by Raynor (*J. Inst. Metals*, 1944, **70**, 531; *M.A.*, **11**, 397) that these compounds are partly characterized by an electron : atom ratio of 1·85, resulting from the absorption of valency electrons by the transition metals to fill the vacancies in the atomic orbitals.—P. C. L. P.

***Nature of the Line Markings in Titanium and Alpha Titanium Alloys.** C. M. Craighead, G. A. Lenning, and R. I. Jaffee (*J. Metals*, 1952, **4**, (12), 1317–1319).—A study has been made of the fine-line markings which appear in the microstructures of unalloyed Ti and α-Ti alloys—and which were first noted by J. and Campbell (*Trans. Amer. Inst. Min. Met. Eng.*, 1949, 185, 646; see *M.A.*, 17, 329). A sample of high-purity Ti rod contg. 0·43 at.-% H, and showing line markings, was resistance treated in a vacuum of 0·05 μ Hg for 4½ hr. at 1500° C. During the early stages of heating the pressure increased to 150 μ Hg, indicating that gases were being evolved. After cooling, the H content was found to be 0·03 at.-%, and metallographic examination showed only a transformed β structure, with no evidence of line markings. When a section of the vacuum-annealed rod was cold worked to sheet and annealed for 1 hr. at 700° C. in air, it then showed an equiaxed α structure free of all line markings. Line markings were again observed after the vacuum-treated rod and sheet had been annealed in H. X-ray-diffraction patterns have identified the line markings as a TiH phase. At least 4·5 at.-% H is soluble in α-Ti at 200° C. The room-temp. solubility of the hydride phase is $<0{\cdot}14$ at.-%; it cannot be retained in solid soln. by quenching in cold water, but precipitates as a finely dispersed phase, either during the quench or on subsequent ageing at room temp. The presence of the insoluble TiH phase in commercial Ti (which contains 0·3–0·5 at.-% H) significantly lowers its impact strength, without impairing its tensile properties. Further work has shown that vacuum annealing at as low as 850° C. will effectively remove the H from Ti. 6 ref.—E. N.

***Kinking in Zinc Single-Crystal Tension Specimens.** Jack Washburn and Earl R. Parker (*J. Metals*, 1952, 4, (10), 1076–1078).—When cylindrical single-crystal Zn specimens were tested at a const. load (to give an extension rate of ~0·05%/hr.), kinking was sometimes observed when the tests were carried out above 200° C. The conditions leading to kinking in tension appeared to be: (1) a non-uniform distribution of flow along the gauge-length, caused by one or more of the following factors, non-uniform distribution of impurities, accidents of growth (lineage structure), slight bending or other damage during handling, surface conditions, and strain-ageing characteristics due to dissolved N, combined with (2) restraint imposed by the tensile load. The phenomenon is discussed on the basis of the dislocation theory. Experiments on the stress-induced motion of small-angle boundaries are reported. The conditions under which sharp kink boundaries form during creep of single-crystal tension specimens appear to be similar to the conditions causing individual grains of a polycryst. specimen to break up into cells. The high temp. and low applied stress result in a process which can be considered as simultaneous bending and polygonization, an excess of dislocations of one sign becoming trapped because of shear-stress gradients. Thermal energy aided by the applied stress brings about collection of the dislocations into sharp boundaries. In the case of single crystals, the shear stress gradients may be imposed by tension loading. In the case of an individual grain in a polycryst. sample, a similar restraint may be imposed by similar grains. 13 ref.—E. N.

***Accommodation Kinking Associated with the Twinning of Zinc.** A. J. W. Moore (*Proc. Phys. Soc.*, 1952, [B], 65, (12), 956–958).—Lattice strains around lenticular twins in Zn single crystals are shown to be relieved by the formation of kinks. The surface irregularities on a cleavage face were studied by a profilometer. The angle of the kink is not const., and will depend on the depth of penetration of the twin.—E. O. H.

***Fundamentals of Cold Working and Recrystallization [Polygonization of Zinc]: Progress Report.** B. L. Averbach, M. Cohen, S. Allen, G. P. Conard, and P. Fopiano (*U.S. Atomic Energy Commission Publ.*, 1952, (**NYO-861**), 2 pp.).—Single crystals of Zn were bent to various radii and then annealed to obtain recovery and polygonization. It appeared that polygonization can be effected under a wide variety of conditions in Zn, and tests are being made at different temp. and with different orientations for the crystal. —B. W. M.

Electron-Microscopic Study of Sintering in Thermistor Flakes. —— (*Bell Lab. Record*, 1952, 30, (7), 285–287).—Briefly describes the technique used for the study by the electron microscope of sintering of thermistor flakes.—K. C. R.

Optical Microscopy: An Adjunct to Engineering and Research. F. G. Foster (*Bell Lab. Record*, 1952, 30, (6), 253–260).—Describes the prepn. of specimens and their examination using the opt. microscope.—K. C. R.

Studies on Electrolytic Polishing. Sakae Tajima (*Metalloberfläche*, 1952, [B], 4, (4), 54–58; (5), 73–75).—A somewhat detailed summary of work carried out by T. during and immediately after the last war—and previously reported in the *J. Electrochem. Soc. Japan*—on: (1) the compn. and operating conditions of baths for the electrolytic polishing of (*a*) Al and its alloys—H_3PO_4, NaOH, and H_2SO_4 soln., (*b*) Fe and steel—H_3PO_4–$(COOH)_2$ soln., (*c*) stainless steels—H_2SO_4 soln., (*d*) Cu and its alloys—H_3PO_4–$CO(NH_2)_2$ soln., (*e*) Ag and its alloys—NaCN–$K_4Fe(CN)_6$ soln., and (*f*) Au, Cd, Cr, Mg, Mo, Ni, Sn, Zn, and their alloys; (2) the potentiometric and corrosion testing of electropolished surfaces, and the reasons for the improved corrosion-resistance of such surfaces; and (3) the mechanism of electrolytic polishing—the influence of (*a*) concentration polarization, (*b*) solubility effects at the anode, (*c*) structure of the anode material, (*d*) action of the electrolyte, (*e*) condition, shape, and surface-area of the anode, (*f*) anodic oxidation, and periodic or cyclic effects at the anode, and (*g*) impressed A.C. There is a brief note on the present position of the sci. and indust. appn. of electrolytic polishing in Japan. 45 ref.—E. N.

On the Nature of Eutectic Alloys. [—I]. S. V. Avakyan and N. F. Lashko (*Zhur. Fiz. Khim.*, 1949, 23, (3), 315–323).—[In Russian]. On the basis of thermodynamics and the available experimental evidence, A. and L. enunciate the following rules concerning the structure of multi-phase eutectics: (i) the total assembly of phases forming an eutectic alloy has the same chem. compn. at all points; (ii) in a sufficiently small, arbitrarily chosen, region of an eutectic alloy, all its constituent phases are in contact; and (iii) the probability of nucleation in a liq. eutectic alloy is the same for all the phases. In confirmation of rule (ii), they cite the experimental facts that a binary liq. eutectic alloy crystallizes only when nuclei of both phases are in contact, and that melting takes place in a contact between two metals or alloys when they are heated to their eutectic temp. A. and L. confirmed the latter fact for contacts of Pb with Bi, Pb–Sn alloy with Bi, Pb–Cd alloy with Bi, Pb–Sn–Cd alloy with Bi, and Pb–Sn alloy with Cd–Bi alloy, but could not produce five-phase eutectics by this method. They show that eutectics with >4 phases are not inconsistent with the three rules, but that the probability of simultaneous contact between nuclei of all the phases decreases rapidly with increasing number of phases. (In passing they suggest that the crystal lattices of the phases are strained in the regions where they are in contact in an eutectic, and attribute anomalies in the corrosion-resistance of eutectic alloys, compared with alloys of nearly eutectic compn., to this strain.) A treatment on the lines indicated by Bethe (*Proc. Roy. Soc.*, 1935, [A], **150**, 552; *M.A.*, 2, 425), in which the free energies of binary solid or liq. soln. are calculated in terms of the energies V_{aa}, V_{bb}, and V_{ab} of pairs of nearest-neighbour atoms A–A, B–B, and A–B, resp., gives the following criteria to be satisfied for the formation of an eutectic: $Z\{V_{ab} - (V_{aa} + V_{bb})/2\} < 2kT$ for the liq. and $>2kT$ for the solid, where Z is the co-ordination number in liquid or solid, resp., T is the temp., and k is Boltzmann's const.—G. B. H.

***On the Nature of Eutectic Alloys. II.—On Eutectic Crystallization in the Presence of Surface-Active Substances.** S. V. Avakyan and N. F. Lashko (*Zhur. Fiz. Khim.*, 1949, **23**, (6), 729–735).—[In Russian]. Cf. *ibid.*, (3), 315; preceding abstract. A. and L. examined micrographically the modification of the KCl–$K_2Cr_2O_7$ eutectic, crystallized from aq. soln. by evaporation at 21° C., by surface-active substances added to the soln. in concentrations of 0·005–0·5 wt.-%. Two types of surface-active agent were used, one (agar-agar) which increases the viscosity of the soln. and reduces the crystal size of each component separately, and the other (Na taurocholate, sulphonic acid, and Na dibutylsulpho-succinate), which has no effect on the viscosity and reduces the crystal size only of KCl. A. and L. conclude that the surface-active agents decrease the size and increase the number of those crystals towards which they show surface activity, and that they change the external form of these crystals in the eutectic. The linear velocity of crystn. is decreased in the phase affected by the surface-active substance, while in the other phase it is increased by supersaturation of the soln. Surface-active substances influence the process of eutectic crystn., shifting the eutectic point. —G. B. H.

***On the Nature of Eutectic Alloys. III.—On the Modification of Binary Eutectics.** S. V. Avakyan and N. F. Lashko (*Zhur. Fiz. Khim.*, 1949, **23**, (7), 858–860).—[In Russian]. Cf. *ibid.*, (6), 729; preceding abstract. A. and L. examined the crystn. of the naphthalene–camphor eutectic under the microscope, with and without addn. of stearic acid under similar conditions of supercooling. Without stearic acid, needles of naphthalene form first as the eutectic crystallizes, and camphor dendrites form only when the melt is supersaturated. Addn. of 0·0001–0·001% stearic acid causes the primary naphthalene needles to become finer and greater in number, and with 0·5% stearic acid this effect proceeds so far that the formation of camphor dendrites " dominates " the crystn. of the eutectic. This leads to a smaller crystal size than in the eutectic formed under similar conditions without stearic acid.—G. B. H.

***On the Nature of Eutectic Alloys. IV.—The Single-Crystal Nature of the Phases in Binary Eutectics.** S. V. Avakyan, E. N. Kislyakova, and N. F. Lashko (*Zhur. Fiz. Khim.*, 1950, **24**, (9), 1057–1060).—[In Russian]. Cf. *ibid.*, 1949, **23**, (7), 858; preceding abstract. Having previously shown (*ibid.*, 1949, **23**, 729; abstract above) that in an eutectic reaction both phases crystallize simultaneously from a number of nuclei, and that the " dominant " phase, which crystallizes the more rapidly, forms discrete particles surrounded by a matrix of the " dependent " phase, A., K., and L. investigated whether the two phases have the same crystal orientation throughout the " eutectic grain " formed from one nucleus. In the Ni–Al eutectic, the slip bands produced in the Ni phase (the " dependent " phase) by compression followed the same direction throughout an " eutectic grain ", showing that the matrix was a single crystal. A Debye–Scherrer photograph of a single " eutectic grain " on the surface of a Bi–Sn eutectic ingot, taken by the focusing method, showed Bi lines, but no Sn lines; hence it is deduced that the matrix of Sn (the " dependent " phase) is a single crystal, but the Bi particles have various orientations. Similar conclusions were reached from the same experiment on an ingot of Cd–Sn eutectic, solidified over a period of 14 hr. to give a large " eutectic grain-size "; in this case also the matrix of Sn was a single crystal, while the Cd particles were of various orientations.—G. B. H.

***Kinetics of Ingot Solidification.** V. M. Breitman (*Doklady Akad. Nauk S.S.S.R.*, 1949, **66**, (3), 385–388).—[In Russian]. Simple expressions, derived math. to describe the progress of the solidification process in cylindrical and spherical ingots, are shown to agree with experimental results relating to the solidification of a 9-lb. paraffin wax ingot and to 1100- and 15,500-lb. alloy steel ingots.—J. S. G. T.

***The Self-Perpetuating Step in Crystal Growth from the Melt.** Bruce Chalmers and Ursula M. Martius (*Phil. Mag.*, 1952, [vii], **43**, (341), 686–687).—A letter. A mechanism is proposed whereby a crystal can grow from the melt under the influence of a finite temp. gradient, without the need of either two-dimensional nucleation or lattice imperfections. Forward movement of the supercooled zone maintains a stepped solid/liquid interface.—P. C. L. P.

***Experimental Investigation of the Influence of Ultrasonic Vibrations on the Kinetics of Crystallization.** A. P. Kapustin (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (3), 357–365).—[In Russian]. K. studied the crystn. of thymol in a glass tube surrounded by a water-jacket at const. temp. Vibrations from a quartz plate (Curie cut) 60 mm. in dia. and 4 mm. thick, excited at its natural frequency of 720 kc./s. by up to 6 kV., could be injected into the melt through the bottom of the tube. This reduced the time for crystn. at 30° C. from 30 min. to 2 sec., produced considerable grain refinement, and improved the soundness of the " ingots ". Tensile tests on " ingots " crystallized with and without ultrasonic vibrations gave reproducible values of 50 and 30 kg./mm.^{2}, resp. Vibration of the walls of the tube caused grain refinement only in the directly irradiated region with substances such as *o*-chloro-nitrobenzene, which do not supercool excessively, but with thymol or benzophenone the effect was spread throughout the " ingot ". If a melt of thymol contg. sand is irradiated at such an intensity that the cavitation pressure is approached, the sand is set in random motion, but in weaker fields it forms clots which sink to the bottom. In still weaker fields the fine particles move to the nodes of the stationary wave pattern. A similar effect was obtained in an aq. soln. of gelatin, contg. Indian ink in suspension. A zoned structure, caused by nucleation at the nodes of the stationary wave pattern, was obtained on crystallizing Rochelle salt from the melt in an ultrasonic field. Grain refinement was also produced when the azobenzene–piperonal eutectic was crystallized in an ultrasonic field, applied after crystn. had started. The characteristic columnar crystn. of thymol could be replaced by a fine-grained structure if crystn. was carried out in an ultrasonic field. This effect was observed with a thin disc " seeded " at the centre and with an " ingot " 40 mm. in dia. and 60 mm. high, " seeded " at the mould walls.—G. B. H.

***†Nuclei Formation and Growth in Crystals in Solid Metals.** W. G. Burgers (*Z. Elektrochem.*, 1952, **56**, (4), 318–323).—[In German]. Paper presented at a symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. Experimental results and theoretical deductions relating to nuclei formation and growth in strained metals are critically and authoritatively reviewed, and are used to form a definite conception of the nature of the nuclei produced by recrystn. of the metal. The conventional nuclei-formation theory due to Burke and Turnbull (1951) is briefly discussed, and is followed by a review of Dehlinger's theory of nuclei formation and growth (*Ann. Physik*, 1929, (v), **2**, 749; *J. Inst. Metals* (Abstracts), 1930, **43**, 523) according to which the strained condition of the metal itself constitutes, to a first approximation, the lattice blocks requisite for nuclei growth. Other matters discussed include the location of embryo nuclei at centres of non-homogeneous strain, the atomic significance of grain-boundary dislocations, boundary surface energy as the motive power of such dislocations, crystal growth in the strained structure, polygonization considered as the essential process involved in the process of nuclei formation. Experimental confirmation of polygonization as the origin of nuclei formation and growth is provided by X-ray studies of strained Al crystals. 49 ref.—J. S. G. T.

Nucleation in Phase Transitions. Victor K. La Mer (*Indust. and Eng. Chem.*, 1952, **44**, (6), 1270–1277).—Presented at an American Chemical Society Symposium on Nucleation. A review is given of the conceptual and phenomenological aspects of nucleation theory associated with Gibbs, Ostwald, Farkas, Frenkel, Volmer, Becker and Doering, together with a discussion of more recent contributions by La M. and Pound. Examples are given of homogeneous and heterogeneous nucleation in gaseous, liq., and metallic systems. 33 ref. —D. P. M.

Liquid and Crystal Nucleations. Guy M. Pound (*Indust. and Eng. Chem.*, 1952, **44**, (6), 1278–1283).—Presented at an American Chemical Society Symposium on Nucleation. Good agreement exists between elementary rate theory and experiment for liq. nucleation from vapour. Recent soln. of difficulties attending the study of the kinetics of cryst. nucleus formation in supercooled liquids has led to progress in quant. theory. Developments in heterogeneous and homogeneous nucleation of condensed phases from supersaturated vapour and in nucleation from supercooled liquids are reviewed. 22 ref.—D. P. M.

Nucleation Catalysis. David Turnbull and Bernard Vonnegut (*Indust. and Eng. Chem.*, 1952, **44**, (6), 1292–1298).—

Presented at an American Chemical Society Symposium on Nucleation. A review of a crystallographic theory of crystal nucleation catalysis. Structural relations in nucleation catalysis are discussed with ref. to ice formation, the crystn. of Hg, Al, and other metals, and the formation of oriented overgrowth. The mechanism of nucleation catalysis is reviewed, and a comparison given of theory and experiment in certain cases. 32 ref.—D. P. M.

The Nature of Diffusionless (Martensitic) Transformations. G. V. Kurdyumov (*Doklady Akad. Nauk S.S.S.R.*, 1948, **60**, (9), 1543–1546; *C. Abs.*, 1949, **43**, 1235).—[In Russian]. The mechanism of the transformation of supercooled solid soln. without change in compn. is considered, such as martensitic transformations of one crystal modification of a solid soln. to another. This formation of a new phase results from nucleation and growth processes rather than from a mech. shear process. K. states that: "The mechanism of the martensitic transformation consists in an orderly reorganization of the lattice in which the relative movement of neighbouring atoms does not exceed the interatomic distance." The high rates of nucleation and growth at low temp. are the result of the small atomic movement needed. The expected decrease of these rates with decreasing temp. was observed. To explain why only certain phase changes are martensitic, the hypothesis is advanced that the martensitic reaction occurs only while the coherency of the new and old phase lattices is maintained at the boundaries of the growing crystal. The coherency produces elastic deformation and the breaking of lattice coherency. A small activation energy can be associated with coherency, and thus the high reaction velocity is explained. Two additional conditions for martensitic growth are: (1) a reacting phase with a high elastic limit; (2) a temp. of transformation at which recrystn. does not occur. The total change of thermodynamic potential as the result of the formation of a lamellar martensitic crystallite of length l and thickness δ is, $R = -Al^2\delta + 2\sigma_1 l^2 + 4\sigma_2 l\delta + E(l, \delta)$, where $A(T)$ is the change of thermodynamic potential at temp. T per unit vol., the sum of the 2nd and 3rd terms is the surface energy, and $E(l, \delta)$ is the elastic energy. The practical independence of E of temp. accounts for the temp. hysteresis of the reaction and for the fixed position of the start of martensitic transformation. There is a possibility of "elastic" martensitic crystals similar to elastic twins. 11 ref.

***†Diffusionless (Martensitic) Transformations. II.—Martensitic Transformations in Heating; Reversibility of Martensitic Transformations. III.—On the Nature of Martensitic Transformations.** G. V. Kurdyumov (*Zhur. Tekhn. Fiziki*, 1948, **18**, (8), 1004–1010, 1011–1025).—[In Russian]. [II.—] Studies of martensitic transformations in Cu alloys have shown that the transformations are reversible. On heating, the martensitic phases change to the initial β phase, the transition occurring within the temp. range of the undercooled solid soln. In continuation of previous work the kinetics of transformation of martensitic phases of alloys of Cu with Al, Zn, and Sn were investigated. One of the principal characteristics of martensitic transformations is the practically instantaneous formation of the new phase within a grain of the initial phase, and the fact that the transformation occurs throughout the grain and the alloy, as a result of the genesis of new crystals, and not because of growth of already existing crystals. The transformation has been studied by X-ray photographs of alloy single crystals, and the crystal-geometry and microstructure relations in the transformations are discussed in considerable detail, with many illustrations. Martensitic transformations also occur in Fe–Ni alloys, e.g. the reversible $\gamma \rightleftharpoons \alpha$ transformation in the Fe–Ni alloy contg. 29% Ni. 33 ref. [III.—] The following matters are discussed in detail: (1) the nature of martensitic phases; (2) martensitic transformations as phase transformations in one-component systems; (3) the kinetics of the transformation, including the explanation of the high rate of nucleation and cryst. growth at low temp., the limited growth of martensitic crystals, the rapid decrease of rate of nucleation on arrested cooling or heating, and the limited number of nuclei produced on isothermal heating within the temp. range of the transformation; (4) kinetics of polymorphic transformations in pure metals as a function of the relationship between transformation temp. and recrystn. temp. In a martensitic transformation, the total number of nuclei is governed principally by the degree of supercooling. 50 ref.—J. S. G. T.

***The Thermoelastic Equilibrium in Martensitic Transformations.** G. V. Kurdyumov and L. G. Khandros (*Doklady Akad. Nauk S.S.S.R.*, 1949, **66**, (2), 211–214).—[In Russian]. The theory of diffusionless or martensitic transformations (*Zhur. Tekhn. Fiziki*, 1948, **18**, 999) explains certain basic abnormalities of such transformations, viz. their very high speed, nucleation and growth of crystals, cessation of growth, conditions under which they can occur in solids, and predicts the existence of new phenomena, viz. their slow rate of progress at low temp. and the existence of "elastic" crystals of martensitic phases. The phenomenon has been observed in C and alloy steels (*Doklady Akad. Nauk S.S.S.R.*, 1948, **60**, 1543), and in eutectoid Cu–Sn alloys (K. and Lysak, *Zhur. Tekhn. Fiziki*, 1949, **19**, 525). The occurrence of "elastic" crystals in the martensitic phase of a Cu–Al alloy contg. ~14·5% Al and ~1·0–1·5% Ni after quenching from 900° C. is now definitely established experimentally. "Elastic" crystals in the martensitic phase do not occur in Cu–Sn alloys contg. ~25% Sn. A mechanism of the phenomenon is discussed.—J. S. G. T.

Determination of the Linear Rate of Transformations [Solution of Metallic Phases] in Solid Metals. A. G. Spektor (*Zavod. Lab.*, 1949, **15**, (7), 797–799).—[In Russian]. Hitherto, measurements of the linear rate of crystal grain-growth during transformations in solid alloys have been made by measuring the rate of growth of the largest crystal cross-section in the plane of the metallographic section. Crystals, for this purpose, must be as nearly spherical as possible and completely isolated by the matrix. A unified method of determining rates of crystal growth and dissoln. of phases in solid metals, applicable to crystals of all shapes is briefly discussed, and illustrated by results obtained by studying the soln. of the carbide phase in austenite of a 1% C, 1·5% Cr steel heated at the rate of 7° F. (4° C.)/sec. Essentially, the method consists in determining the surface area, S, of the transformation phase by the method of secants (Saltykov, *ibid.*, 1946, **12**, 816; *M.A.*, **14**, 318), and dividing S into the 3-dimensional rate of transformation of the phase, to give the linear rate of transformation of the phase.

—J. S. G. T.

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5 — POWDER METALLURGY

***Diffusion of Metals in Samples of Sintered Aluminium [SAP].** (Seith and Löpmann). See col. 545.

Chromium Carbide Meets Many Industrial Needs. J. D. Kennedy (*Materials and Methods*, 1952, **36**, (2), 166, 168, 170, 172, 174).—The phys. properties of cemented Cr carbide components contg. 83% Cr carbide, 2% WC, and 15% Ni are reviewed. It is claimed that a range of shapes and sizes similar to those available in WC will soon be obtainable in Cr carbide and these will replace WC for many appn.

—R. P. H. F.

***Role of the Binder Phase in Cemented Tungsten Carbide-Cobalt Alloys.** Joseph Gurland and John T. Norton (*J. Metals*, 1952, **4**, (10), 1050–1056).—WC–Co and WC compacts were prepared from powders of commercial grade and sintered, the former at 1500° C. in a H–N atmosphere, and the latter at 2000° C. in vacuum, after which the following experiments were carried out: (i) electrolytic leaching—to remove the binder from the WC–Co compacts, for the purpose of determining the continuity of the carbide phase, (ii) detn. of the coeff. of thermal expansion, (iii) detn. of the dilatometric cooling curve, (iv) infiltration—to determine the behaviour of a pure WC compact in the presence of the liq. binder metals Co and Cu, (v) sintering variations—to evaluate the influence of compn., temp., and time on the densification of WC–Co compacts, and (vi) wettability—to determine the properties of the solid/liquid interface in WC–Co compacts, measurements were made of the angle of contact of a liquid drop of Co and Cu on the solid surface of WC. The results show that: (1) Densification of WC–Co compacts takes place by a re-arrangement of the carbide particles under the influence of the surface-tension forces of the binder, which tend to eliminate pores present in the compact. (2) Considerable soln. and redeposition of WC takes place during sintering; this leads to grain growth, for the small particles of WC dissolve in the liq. Co and are reprecipitated on the larger ones, so as to minimize the interfacial tension of liquid and solid. (3) The structure of the cemented compact is one of WC particles dispersed in a matrix of Co. A WC skeleton does not form during sintering or on cooling from the sintering temp.; this also follows from theoretical considerations. Liq. Co breaks up and destroys an already existing WC skeleton, whereas liq. Cu preserves it. (4) This structure accounts for some of the characteristic properties of the cemented compacts. The high strength and brittleness is due to the mech. restraint exerted by the WC particles on the thin films of binder, the yield strength of which is correspondingly raised, and to the complex state of stress resulting from the presence of residual stresses of thermal origin. (5) The following are the essential characteristics required for an effective binder for cemented WC compacts: (*a*) it must supply a liq. phase at relatively low temp.—Co appears to do so at $\sim$1320° C., (*b*) WC must be soluble in the binder—the solubility of WC in Cu is insignificant compared with that in Co, and (*c*) the liq. binder must wet the solid WC particles—at $\sim$1400° C. the angle of contact of liq. Co on solid WC is infinitely small, whereas that of liq. Cu is $\sim$20°. 15 ref.—E. N.

The Manufacture of Cemented Tungsten Carbide Dies. B. E. Berry (*Wire Ind.*, 1952, 19, (217), 45–46, 49–52, 55).—The manufacture of cemented carbide, or "hard metal", is outlined. Calcined wolfram is mixed with soda ash and roasted to convert the W content to Na_2WO_4. The soda roast product is leached to form a crude Na_2WO_4 liquor which is concentrated and purified. The pure liquor is treated with $CaCl_2$ to precipitate $CaWO_4$ which is filtered and repulped to form slurry. The conditions under which the pptn. is carried out influence the grain-size of the W powder. The slurry is then acid-treated and the tungstic acid is filtered off and dried. The W metal powder produced therefrom has a tap wt. of 28–32 c.c./100 g. and a grain-size in which $\sim$60% is $<$0·5 μ. Sintering, pressing, and shaping of W carbide and the making of drawing dies and tool tips are detailed. Ref. is made to Elmarid, a German hot-pressed hard metal.—C. P. F.

Powder-Metallurgy Parts in High-Temperature Applications. Julius J. Harwood (*Materials and Methods*, 1952, **36**, (2), 87–91).—The developments of such materials as Cu infiltrated powder mouldings for use as jet compressor parts with U.T.S. of 90,000 lb./in.^{2} and elongations of 8% are reported. The use of Ti alloys produced by powder-met. methods for such appn. as compressor discs and filters for HNO_3 (fuming) is described. The high-temp. strength and other characteristics of sintered Al powder (SAP) are discussed and its possible use as compressor-blade material, considered. The stress–rupture curves of TiC bonded with Ni (20–30%) are given, and the resistance of these materials to oxidation, thermal shock, and fatigue failure is illustrated. The development of porous stainless-steel powder components for use as transpiration, or sweat cooling materials in gas turbines is reviewed. The possibilities of utilizing the high-temp. properties of Mo and its alloys are indicated, provided suitable protective coatings can be developed.—R. P. H. F.

Practical Applications of Hot Pressing. Jerome F. Kuzmick (*Materials and Methods*, 1952, **36**, (1), 84–87).—The advantages and difficulties of hot pressing are discussed in connection with the prodn. of hot-pressed powder components in WC and heavy metal (W 90 and Ni–Cu binder 10%), and also in the diamond dressing, grinding, and drilling fields. Hot pressing is claimed to eliminate sintering operations, and the properties of hot-pressed alloys are similar to those of cold-worked annealed materials. Graphite moulds, used once only, are said to be economical for hot pressing a variety of shapes.—R. P. H. F.

The Plus and Minus of Making Your Own Metal-Powder Parts. Joseph L. Bonanno (*Materials and Methods*, 1952, **35**, (4), 101–105).—The appn. of powder-met. methods to the prodn. of a large number of small components is described. The method is said to be economical on a pilot-plant scale if efficient tool design and tool-making facilities are available. In addn. a well-equipped laboratory to control the characteristics of raw materials is claimed to be an essential feature. —R. P. H. F.

Forming Processes for Metals [The Powder-Metallurgy Process]. W. M. Halliday (*Mass Prodn.*, 1952, 28, (6), 86–92).—Cf. *M.A.*, **20**, 488. H. discusses the dimensional accuracy obtainable and considers the influence of die design on the cost and quality of prodn.—R. W. R.

How Powder Metallurgy Is Progressing. H. W. Greenwood (*Metallurgia*, 1952, **46**, (278), 289–292, 298).—A review, with 31 ref.—F. M. L.

6 — CORROSION AND RELATED PHENOMENA

Aluminium and Water. Max Schenk (*Metalloberfläche*, 1952, [B], **4**, (3), 33–36).—S. discusses the effects of water on Al in relation to: (1) the role of the invisible surface film of Al_2O_3; (2) the electrochem. nature of the corrosion, and the reactions which occur; (3) the influence of impurities and alloying constituents—especially Cu—in the metal; (4) the nature of the corrosion products; and (5) the influence of the nature of the water—differentiating between rain, river, fresh, salt, tap, distilled, and softened water, and condensate.—E. N.

Observations on the Corrosion of Copper in a Closed-Circuit Evaporator. L. W. Haase (*Metalloberfläche*, 1951, [A], **5**, (11), 170–171).—The corrosion of Cu tubes—made of 99·96%-purity Cu free from Cu_2O and CuO—in a low-pressure closed-circuit evaporator is discussed. The corrosion—which took the form of intense pitting immediately below the liquid/vapour interface, i.e. within the meniscus zone, and deposition of rust lower down—is considered to have been caused by the water having a high O content, and, as a consequence, contg. relatively large amounts of hydrated Fe oxides which reacted electrochem. with the Cu.—E. N.

***The Corrosion of Mild Steel and Brass in Chlorinated Water.** F. Wormwell and J. T. Nurse (*J. Appl. Chem.*, 1952, **2**, (12), 685–692).—Investigations carried out at the Chemical Research Laboratory and four outside stations with regard to the corrosion of brass (61·5–64% Cu) and mild steel in chlorinated supply water are reported. The specially designed apparatus used, which is described and illustrated, gave an arrangement by which sheet-metal specimens suitably mounted inside glass vessels were exposed to controlled water flows along their lengths. With low-speed water flow (1·2 ft./min.) corrosion of brass was very slight, the rate of the corrosion being usually $<$25% of that for mild steel, though in one case where the chlorinated water was hard the corrosion rate of brass rose to approx. half that for mild steel. At another station appreciable corrosion was observed on a brass orifice plate in the chlorinated-water line where the rate of flow was high ($\sim$50 ft./sec.), but the water at this station was rather more corrosive towards brass, even at low rates of flow, than at other stations, and it is suggested that chlorination greatly enhanced the corrosion at high flow rates. —J. R.

Corrosion [of Tantalum, Niobium, Zirconium, Tungsten, and Molybdenum]. Mars G. Fontana (*Indust. and Eng. Chem.*, 1952, **44**, (7), 71A–72A, 74A).—Data on the corrosion-resistance of Ta, Nb, Zr, W, and Mo are given in tabular form.—D. P. M.

***The Thermal Oxidation of Tin.** R. K. Hart (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 955).—Electron-diffraction studies of Sn foil, heated in air with a locally applied flame, revealed that there was a gradual transition from the initial Sn, through an amorphous oxide to a cryst. deposit of SnO and finally to SnO_2 in the hottest region. It was found that amorphous films were formed up to $\sim$130° C.—E. O. H.

***Corrosion by Molten Bismuth.** H. A. Pray, R. S. Peoples, and W. K. Boyd (*U.S. Atomic Energy Commission Publ.*, 1952, (**BMI-773**), 23 pp.).—The corrosion-resistance in static molten Bi at 750° and 850° C. was determined for $\sim$50 materials representing low-alloy steels, transformer-grade Si steels, stainless steels, and various heat-resistant metals and alloys. Low-alloy steels were wetted and moderately attacked, but Armco Fe and low-C Fe alloys were severely attacked after relatively short exposures. Transformer-grade steels with 2·9% Si showed good resistance and were considerably superior to any other material tested; a similar alloy contg. 4·5% Si appeared unattacked on visual examination, but some attack was observed on metallographic sections. Stainless steels and heat-resistant alloys were wetted by the Bi and showed alloying or moderate attack. The most severe corrosion appeared to be centred at or near the vapour/liquid interface, but gases in the Bi had little effect on its corrosiveness.—B. W. M.

***Corrosion of Refractories by Tin and Bismuth.** T. A. Coultas (*U.S. Atomic Energy Commission Publ.*, 1952, (**NAA-SR-192**), 23 pp.).—The static corrosion of several refractory materials in Sn and Bi at 1800° F. (980° C.) was estimated by visual and metallographic examinations and by chem. analysis of the residual liq. metals. The materials in decreasing order of resistance to attack were, for Sn: graphite, Al_2O_3, Vycor, W, $MoSi_2$, TaC, Kennametal K-138A, Haynes Metamic LT-1, Ta, Mo; and for Bi: graphite, Al_2O_3, ZrO_2, TaO_2, mullite, Vycor, TaC, MgO, $MoSi_2$, Re, W, Mo, K-138A, SiC, Ta, Metamic LT-1, Ir, and AISI type 303 stainless steel with Mo coating. A simple " convection-diffusion tube " apparatus was described for testing the suitability of liquids for a heat-transfer system. 6 ref.—B. W. M.

***A Parabolic Law for Metal Oxidation Which is Not Controlled by Diffusion.** W. J. Moore (*Phil. Mag.*, 1952, [vii], **43**, (341), 688–689).—A letter. It is suggested that the rate-determining factor in the oxidation of many metals such as Zn, is the reaction between a singly positively charged interstitial Zn ion and a singly negatively charged adsorbed O ion, to form a new cryst. unit of ZnO. The mechanism explains the magnitude of the rate const., and why they vary with O pressure below a limiting pressure.—P. C. L. P.

***The Co-Operation of Phase-Boundary Reactions in the Oxidation of Metals and Alloys at High Temperatures.** K. Hauffe and H. Pfeiffer (*Z. Elektrochem.*, 1952, **56**, (4), 390–398).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. Results relating to the oxidation of Fe to FeO at 1000° C., and to the sulphiding of Ni to NiS at 670° C., lead to the conclusion that the parabolic time-law and, therewith, the Wagner *Zunder* theory (*Z. physikal. Chem.*, 1933, [B], **21**, 25; 1936, [B], **32**, 447; *M.A.*, **1**, 132; **3**, 293) is not applicable to these reactions, as boundary surface reactions control the velocities of the reactions. Moreover, the presence of metallic oxide vapours, e.g. MoO_3 and LiO_2 in the surrounding atmosphere can increase or decrease the rate of oxidation of metals and alloys, which accords with the disorder theory of heterotype mixed phases. 29 ref.—J. S. G. T.

Dry Oxidation and Wet Corrosion. U. R. Evans (*Nature*, 1951, **168**, (4281), 853–855).—The substance of a lecture delivered on 26 October 1951, at the University of Birmingham, in which E. correlates the high-temp. oxidation of metals with their wet corrosion at low temp., and shows that under certain conditions the processes concerned are similar. The effect of inhibitors, S, and impurities on the two processes are discussed. 19 ref.—R. S. B.

New Views on the Relationship Between Electrochemical and Chemical Corrosion. A. Bukowiecki (*Chimia*, 1952, **6**, (7), 156–165).—B. discusses the relative importance of electrochem. and purely chem. attack of various reagents on metal surfaces, mainly Zn in weak salt soln. He shows that electrochem. couples can be formed on the surface of a metal even when the liquids in contact with it have poor elect. conductivity. In many cases, however, the action is purely chem. or a combination of both. Electrochem. attack is not necessarily more dangerous than purely chem. attack, except when it leads to localization of attack through the formation of small anodes and large cathodes. Such local concentrations can take place even when the attacking liquid has poor elect. conductivity.—W. F. H.

***The Breakdown of Oxide Films on Metal Surfaces in Acid Atmospheres and the Mechanism of Atmospheric Corrosion.** W. Feitknecht (*Chimia*, 1952, **6**, (1), 3–13).—F. discusses the more important factors governing the corrosion of Fe, Zn, Cd, Ni, and Cu in atmospheres contg. HCl vapours. Three kinds of breakdown of the primary oxide film can be distinguished: (1) (Fe and Ni) the primary film breaks down only at isolated points; (2) (Zn and Cd) the whole surface layer breaks down and an even layer of basic salt is formed; (3) (Cu) the natural oxide film at first thickens, and this is followed by the formation of a layer of CuCl which eventually leads to accelerated corrosion at isolated centres. The progress of these different processes was studied by means of the electron microscope. An electrochem. mechanism is put forward to account for the breakdown of the primary oxide film, leading to atmospheric corrosion, and the effect of the chem. and electrochem. properties of the corrosion products of each of the metals on the final results are discussed.

—W. F. H.

***Cathodic Processes in the Corrosion of Metals.** N. D. Tomashov and T. V. Matveeva (*Zhur. Fiz. Khim.*, 1950, **24**, (11), 1281–1293).—[In Russian]. T. and M. studied the rates of evolution of H ($H^+ + \ominus \rightarrow H$; $H + H \rightarrow H_2$) and of absorption of O ($O_2 + 2H_2O + 4\ominus \rightarrow 4OH^-$) during the corrosion of various metals and alloys immersed in 0·5*N*-NaCl soln. at 25° C. under an atmosphere of air for periods of 100–230 hr. The specimen (previously cleaned with grade 0 glass-paper, washed, and dried) and the soln. were contained in a closed vessel connected to a const.-pressure gas analysis apparatus; the vol. of H evolved was measured by the decrease in the vol. of gas when a Pt spiral inside the vessel was heated to incandescence, and the vol. of O absorbed was deduced from this decrease and the net increase in the vol. of gas during corrosion. Al (contg. Si 0·03, Cu 0·1, Fe 0·01%) and electrolytic Cu both showed only absorption of O under these conditions; Cd, Armco Fe, and Zn all showed an evolution of H $\ll$ the absorption of O, and Mg showed only evolution of H. Several Al-rich alloys showed mainly absorption of O, but the evolution of H increased with the content of alloying elements, the resp. relative rates σ and h of the two processes being as follows: for commercially pure Al (contg. Cu 0·004, Fe 0·14, Si 0·12%), σ 3·8, h 0·4; for Al–Mn alloy (contg. Cu 0·2, Mg 0·05, Mn 1·5%), σ 4·8, h 0·45; for Al–Mg alloy (contg. Mg 2·4, Mn 0·30, Fe <0·5, Si <0·5%), σ 8·4, h 0·4; for commercially pure Al (contg. Si 0·20, Cu 0·30, Fe 0·26%), σ 9·9, h 0·5; for Al–Si–Mg alloy (contg. Si 0·6, Cu 0·2, Fe 0·17, Mg 0·8%), σ 11·3, h 1·6; and for Duralumin (contg. Cu 4·0, Mg 1·5, Mn 0·49, Fe 0·66, Si 0·40%), as aged, σ 15·6, h 1·3. After annealing for 2 hr. at 350° C. to precipitate $CuAl_2$, the Duralumin specimen gave much more rapid evolution of H with unchanged absorption of O, the relative rates being σ 17, h 5·0. Study of this alloy under varying conditions showed that, under conditions which facilitated diffusion of O through the soln. (reduced depth of immersion in the soln., alternate immersion and withdrawal of the specimen, stirring of the soln., increase of temp., substitution of O for air as an atmosphere), the rate of absorption of O increased as expected, and the rate of evolution of H also increased, so that the ratio between them remained const. T. and M. attribute this fact to the "negative difference effect", i.e. increase in the working efficiency of the micro-cathodes in the alloy with increased diffusion of O to them. Addn. of H_2O_2 to the soln. increased the evolution of H from Duralumin more than the absorption of O; this also is explained by the "negative difference effect", since the reaction $H_2O_2 + 2\ominus \rightarrow 2OH^-$ contributes to the rate of corrosion without being observed as a change in gas vol. T. and M. showed experimentally that, under certain practically important conditions (complete immersion in distilled water, wetting with distilled water, suspension of dry specimen above 0·5*N*-NaCl soln.), Mg is corroded with absorption of O and evolution of H at roughly equal rates.—G. B. H.

Control Galvanic Action: Stop Costly Failures. E. C. Reichard (*Materials and Methods*, 1952, 36, (2), 83–86).—The theory of galvanic corrosion is discussed and examples of this type of corrosion, together with the estimated costs of replacement of corroded materials, are given. The position of various metals in the galvanic series and the reversible properties of stainless steel and Zn are used to illustrate that only slight changes in conditions of service may result in marked alterations in corrosion conditions. Cathodic protection and other methods of preventing corrosion are also discussed.—R. P. H. F.

***Accelerated-Cavitation Research.** W. J. Rheingans (*Trans. Amer. Soc. Mech. Eng.*, 1950, 72, (5), 705–719; discussion, 719–724).—Results of accelerated cavitation-pitting tests on a machine of a vibrating type using varying amplitudes of vibration, varying depths of submergence in test liquids of alkalies, acids, and oils. Tests showed that Ampco bronzes, and Colmonoy might have distinct advantages over present materials for use in hydraulic machinery. Repair by welding influences the resistance to pitting, whereas preheating the base metal has very little effect on the resistance. Hardness has a definite effect on the resistance to pitting for all materials tested.—H. Pl.

A Critical Look at Salt-Spray Tests. F. L. LaQue (*Materials and Methods*, 1952, 35, (2), 77–81; and (abridged) *Métaux, Corrosion-Ind.*, 1952, 27, (321), 233–236).—LaQ. discusses the limitations of salt-spray tests, and points out the fallacy of the assumption that a number of hours' exposure to a salt spray is equivalent to so many months' or years' service in some other environment. In addn. it is stressed that the objectives of the salt-spray test, such as the measurement of the thickness of protective coatings, measurements of the relative protective values of various metallic and non-metallic coatings, &c., are not realized. 11 ref.—R. P. H. F.

***Polarization Method for the Speedier Testing of Metals for Corrosion-Resistance in Sea-Water.** L. V. Elin and E. S. Ryt (*Zavod. Lab.*, 1949, 15, (7), 811–813).—[In Russian]. Two identical specimens of the same metal are immersed in electrolyte (sea-water) and polarized by a battery. By varying the current, *I*, between the electrodes and measuring the p.d., Δu, between them, the differential polarization curve $I = f(\Delta u)$ is determined, and therefrom the comparative resistance of the metal to corrosion in the electrolyte is ascertained by the method proposed by Akimov ("Theory and Methods of Testing the Corrosion of Metals", 1945, p. 354). Preferably, Δu is determined by a condenser discharge through a ballistic galvanometer. Illustrative examples are given of results obtained with 4 samples of steel and Cu, and it is claimed that the method gives relative qual. information relating to corrosion-resistance in a much shorter time than ordinary methods.—J. S. G. T.

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7 — PROTECTION

(Other than by Electrodeposition)

New Protective Treatment for Aluminium Simplifies Processing at Reduced Costs. R. Stricklen (*Materials and Methods*, 1952, 35, (2), 91–95).—The use of the Iridite chromate-coating process for Al is discussed. The results of salt-spray tests are given to indicate the relative protective qualities of the coating as compared with anodized coatings, &c. In general, the salt-spray resistance is said to be inversely proportional to the Cu content of the alloy. The chromate coating provides a good base for painting, and the elect. resistance of the coating is said to be satisfactory for certain appn. The methods of surface prepn. and appn. of the coating are described.—R. P. H. F.

The Patent Literature in the Field of the Anodic Oxidation of Aluminium. Max Schenk (*Metalloberfläche*, 1952, [B], 4, (3), 43–46; (4), 59–61; (6), 91–92).—Patent numbers—mainly German, but with some British, French, Italian, Swiss, and U.S. equivalents—appn. dates, names of patentee, and abridged descriptions of the salient features of the patents, are given under the groupings: (1) electrolyte compn.—36 ref.; (2) processes involving special features or characteristics—43 ref.; (3) electro-brightening and polishing—13 ref.; (4) dyeing processes—16 ref.; (5) sealing processes—28 ref.; (6) special effects, i.e. direct printing and photographic processes—18 ref.; and (7) special appn. of the anodized surfaces—69 ref.—E. N.

***Polishing of High-Purity Aluminium [and Its Alloys with 0·5, 1, and 2% Magnesium] by Repeated Anodizing and Re-Solution of the Anodic Films.** (Lattey). See col. 601.

Grinding and Polishing of Aluminium and Aluminium Alloys, Especially as Pretreatments for Anodic Oxidation (Eloxal Treatment). (Erdmann). See col. 602.

How Lead-Clad Steel and Copper Extend Lead Applications. Alfred P. Knapp (*Materials and Methods*, 1952, 36, (4), 115–117).—The cladding of steel (Ferrolum) and Cu (Cupralum) with Pb by the pouring, burning, and expanding methods is described. It is stated that the strength of the bond between Pb and the base metal is increased by rolling and drawing. The appn. of Pb-clad Cu given include the handling of H_2SO_4, cooling coils in anodizing baths and anodes in Cr plating baths. The use of Pb-clad steel in X-ray equipment, in nuclear fission, and in the handling of Ra is discussed.—R. P. H. F.

Tin- or Lacquer-Coated Steel? W. Richards (*Tin-Printer*, 1950, 26, (304), 8–9).—The advantages and limitations of Sn and lacquer coatings for steel are reviewed. A lacquered tinplate seems to be the soln. to many problems, but much improvement of this material is necessary.—S. R. W.

Tin Coatings and Compounds. W. Richards (*Tin-Printer*, 1950, 26, (302), 8–9).—A brief description of the effect of Sn coatings on the properties of tinplate is given.—S. R. W.

Surface Defects on Tinplate. T. Fielding (*Tin-Printer*, 1952, 28, (327), 8–9; (328), 6–7).—A reprint. See *ibid.*, 1948, 24, (281), 10; *M.A.*, 16, 494.—S. R. W.

Microstructure of Tinplate. T. Fielding (*Tin-Printer*, 1952, 28, (325), 8–9; (326), 6–7).—A reprint. See *ibid.*, 1948, 24, (278), 8–9; (279), 8–9; *M.A.*, 15, 466.

—S. R. W.

Thin, Ductile Carbide Coatings Possible with New Method [Flame Plating]. F. E. King (*Materials and Methods*, 1952, 36, (4), 112–114).—Cf. *M.A.*, 20, 497. The method of flame-plating developed is not described. It is claimed that powdered WC can be deposited in the form of a coating 0·0005–0·02 in. thick by this method. No welding or inter-diffusion of the carbide coating is said to occur during flame-plating, and the max. temp. of the operation is claimed to be ~400° F. (205° C.). The coatings are said to be ductile and to exhibit greater resistance to wear than sintered WC. A number of appn. of flame-plating are given.—R. P. H. F.

***Determination of the Temperature of the Internal Surfaces of Iron Pipe-Lines [for Acetylene Gas] During Flame-Spraying with Zinc.** Otto van Rossum (*Metalloberfläche*, 1951, [A], 5, (11), 172).—The external surfaces of tubes, 80 and 600 mm. in dia., and 3·5 and 4 mm. wall thickness, have been flame-sprayed with Zn, and the rise in temp. of the internal surfaces has been determined. The wall temp. inside the tube was found to rise to ~80° C. and well above 100° C. for the large- and small-dia. tubes, resp., when deposits of the usual thickness (0·12–0·15 mm.) were applied without interruptions for the work to cool. It is concluded that the interiors of C_2H_2 containers and pipe-lines must be thoroughly cleaned to remove all traces of gas, in order to prevent any explosion hazard when their external surfaces are subsequently flame-sprayed with Zn for corrosion-protection purposes.—E. N.

Modern Developments in Metallizing. Sam Tour (*Weld. J.*, 1952, **31**, (3), 199–207).—Wire spray guns, surface prepn., types of coatings, the mech. properties of coatings, and corrosion appn. are reviewed.—K. B.

Metallic and Non-Metallic Coatings for Grey Iron.—I.–II. Charles O. Burgess (*Foundry*, 1950, **78**, (12), 76–79, 194, 196, 198, 202–203; 1951, **79**, (1), 96–99, 225–236).—[I.—] Accounts are given of the prodn. and appn. of the different metallic and non-metallic coatings used to protect grey Fe castings from corrosion or wear or to improve their appearance. Among the metallic coatings considered are sprayed-metal coatings, and coatings of Sn, Zn, Pb, Pb–Sn alloy, and Al produced by hot-dipping. 34 ref. [II.—] B. describes the prodn. and appn. of the following types of coating on grey Fe castings: (*a*) hard-facing materials (ferrous alloys, non-ferrous alloys, and carbides); (*b*) cementation and diffusion coatings (carburizing, chromizing, and sherardizing); and (*c*) electrodeposited coatings of Cu, Ni, Cr, Cd, Zn, Pb, and Sn. 69 ref. —R. W. R.

Cathodic Protection and Its Application to the Protection of Buried Pipe-Lines. Heinrich Steinrath (*Metalloberfläche*, 1952, [A], **6**, (4), 49–52).—The prevention of attack on coated Fe pipe-lines and on cables by maintaining the line or cable cathodic in respect to the adjacent earth is discussed. Two systems which are commonly employed are described and illustrated: (1) Sacrificial anodes—of Zn, Al, or Mg, and (2) external e.m.f. in conjunction with insoluble anodes (e.g. graphite), A.C. from generators or windmills being stepped down by transformers and rectified to give the necessary D.C. 14 ref.—E. N.

***Cathodic Protection of Steel.** F. Wormwell, T. J. Nurse, and H. C. K. Ison (*Chem. and Ind.*, **1952**, (40), 972–974).—Results from hitherto unpublished work carried out at the Chemical Research Laboratory with regard to elect. current and potential requirements for the cathodic protection of steel, using Zn and other sacrificial anodes, are presented, and are considered with ref. to previously reported studies. Protection against sulphate-reducing bacteria is discussed by W. and *T. W. Farrer*. 37 ref.—J. R.

Nichtmetallische anorganische Überzüge. (Machu). See col. 606.

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8 — ELECTRODEPOSITION

Electroplated Coatings on the Light Metals. Philip O'Keefe (*Materials and Methods*, 1952, **35**, (6), 119–130).—Materials and Methods Manual No. 83. The use of the H_3PO_4 anodic process, the Zn-immersion coating process, the wet-blasting process, and the rubber-coating process for preparing Al and its alloys for plating are described. Of these the Zn-immersion process is claimed to be the most widely used, the Work process has not been widely used owing to the difficulty of control in prodn., and the H_3PO_4 anodic process has a number of drawbacks, including sensitivity to alloy compn. and the difficulty of plating large areas and irregular surfaces. The properties and uses of electrodeposits on Al and its alloys are also reviewed, but details of plating procedures are not included. The surface prepn. of Mg and its alloys, together with the methods of plating, properties, and appn. of electrodeposits are discussed. Finally the development of likely methods for plating Ti and its alloys is reviewed.—R. P. H. F.

***Amperometric Determination of Certain Main Components [Nickel, Cadmium, Copper, Zinc, Aluminium, and Sulphate Ions] of Electroplating Baths.** N. G. Chovnik, N. N. Kuz'mina, A. N. Galkina, and B. Y. Starik (*Zavod. Lab.*, 1949, **15**, (5), 517–522).—[In Russian]. The appn. of the apparatus and amperometric method previously described (*Zhur. Obshch. Khim.*, 1947, **17**, 625) for the detn. of constituents of electroplating baths, involving the use of a Hg dropping electrode to obtain complete titration curves of the electrolyte is discussed. Gelatin is added to the electrolyte to suppress polarographic max. $K_3Fe(CN)_6$ is a suitable reagent for the detn. of Ni and Cu; $K_4Fe(CN)_6$ is suitable for the titration of Zn. A bath contg. SO_4 ions is titrated with $PbNO_3$, and a somewhat similar process can be used for the detn. of Al in H_2SO_4 anodizing baths.—J. S. G. T.

Palladium [Electro-]Deposits. R. Erdmann (*Metalloberfläche*, 1952, [B], **4**, (1), 6–7).—The compn., operation, and control of Pd-plating baths are briefly reviewed.—E. N.

Silver- and Gold-Plated Parts Meet Tough Specifications. T. C. Du Mond (*Materials and Methods*, 1952, **36**, (5), 114–115).—The use of Ag and Au electrodeposits 0·0007 in. thick on components such as elect. contacts, elect. connectors, &c., is discussed. The necessity for the good adhesion of the extremely thin deposits is stressed. The use of severe bend tests to evaluate the adhesion of deposits and their resistance to blistering or cracking at 500° F. (260° C.) is described. —R. P. H. F.

The Electrodeposition of Tin Alloys. Edmund R. Thews (*Metalloberfläche*, 1952, [B], **4**, (1), 5–6).—A brief review.—E. N.

***Influence of the Wave Form of Rectified Alternating Current on the Growth of Electrodeposited Coatings [and the Operation of the Bath].** Max Ed. Beckmann and Friedrich Maas-Graefe (*Metalloberfläche*, 1951, [A], **5**, (11), 161–169).—A study has been made of the effects of using rectified A.C., as compared with D.C. obtained from batteries, on the operation of various types of plating baths—alkaline and acid Zn and Cu, acid Ni, and hard Cr. The A.C., at 380 V., 50 c./s., was subjected to 1-, 2-, 3-, and 6-phase rectification, to produce currents similar to that obtained from the batteries (20 V. and 20 amp.). The operating characteristics of the baths, and the columnar nature of the deposits obtained were generally unaffected by variations in the current supply, but the following exceptions were noted: (1) pure D.C. supply from batteries usually gives the highest cathodic current efficiencies; (2) alkaline Zn baths operating from 1-phase rectification produce a slightly modified columnar deposit; (3) the cathodic current efficiency of alkaline $Cu(CN)_2$ baths decreases by ~10% on changing from 1-phase to 6-phase rectification; (4) acid Cu baths contg. organic reagents, such as glycol or acridine, as grain refiners, are more efficient when working on current supplied from 6-phase rectification than from batteries, an effect which is reversed when the organic reagents are omitted; (5) 6-phase rectification produces the lowest current efficiencies in acid Ni baths, but the deposits obtained are the least porous; and (6) hard-Cr baths (*a*) produce grey matte coatings, in place of the usual bright deposit, when operating on 1-phase rectification; this appears to result from the formation of the hexagonal modification of Cr in place of the usual cubic lattice structure, and (*b*) operating on 2-phase rectification have, for some inexplicable reason, cathodic current efficiencies of only ~15%, compared with the customary 18–20% obtained with the other forms of current-supply.—E. N.

Automatic Control of Plating Operations. Leo Walter (*Machinery* (*Lond.*), 1950, **76**, (1951), 387–389).—A brief survey is given of common control problems encountered in a plating shop, particularly the temp. control of the baths. —S. R. W.

Defects in Barrel and Vat Electroplating.—I.–II. Otto Niedermeyer (*Metalloberfläche*, 1952, [B], **4**, (1), 10–11; (2), 24–25).—Faults and their remedies, in degreasing, pickling, and plating procedures are briefly discussed.—E. N.

Review of Recent Developments in Electroplating. H. Silman (*Metallurgia*, 1952, **46**, (278), 283–288, 298).—A review, with 28 ref.—F. M. L.

9 — ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

***Investigation of the Exchange and Diffusion Processes Occurring at Silver Electrodes by Means of Radioactive Indicators.** Heinz Gerischer and Wolf Vielstich (*Z. Elektrochem.*, 1952, **56**, (4), 380–386).—[In German]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. A study of the exchange and diffusion processes between activated metallic Ag atoms and Ag^+ ions in $AgNO_3$ soln., using the radioactive isotope ^{110}Ag as indicator, shows that diffusion of ions from the electrolyte to the electrode and their incorporation in the uppermost exchange layers are the factors determining the velocity of the interaction. After an initial rapid exchange in which only the uppermost exchange layers are concerned, there follows a further slow increase of activity due to a purely diffusive process. The diffusion coeff. of the metal, derived from the results, is $\sim 10^{-14}$–10^{-15} cm.^{2}/day.—J. S. G. T.

Electrometallurgy of Tin and Its Alloys. E. S. Hedges and J. W. Cuthbertson (*Chem. and Ind.*, **1952**, (52), 1250–1254).—Presented at a Symposium organized jointly by sections of the Society of Chemical Industry and the Royal Institute of Chemistry. Sn is not normally extracted from its ores by electrolysis, but such processes are important in the prodn. of tinplate, simultaneous deposition of Sn with other metals, detinning of scrap, and, to a lesser extent, refining of Sn metal. Resp. electrochem. and practical aspects are discussed. 14 ref.—J. R.

***The Production of Zirconium by Fused-Salt Electrolysis.** Merle E. Sibert and Morris A. Steinberg (*U.S. Atomic Energy Commission Publ.*, 1952, (**NYO-3118**), 24 pp.).—A description of further work on the double-fluoride process for the electrolytic prodn. of Zr, including the design of a bath for continuous operation. By vacuum drying of the K_2ZrF_6, a pre-electrolytic step is eliminated and better-quality deposits are obtained. There is evidence that recrystn. of the starting material results in elimination of impurities and that the use of $KZrF_5$ as a source material is an economic improvement. Arc-melted Zr can be cold worked by various techniques and has been rolled to 0·005 in. sheet and drawn to 0·03 in. wire. The compn. varied from 99·2 to 99·9% Zr with O, C, and N as the chief impurities.—B. W. M.

***The Electrochemical Series of Potentials for Metals in Molten Sodium Bromide and Potassium Bromide as Solvent.** Yu. K. Delimarsky and A. A. Kolotti (*Zhur. Fiz. Khim.*, 1949, **23**, (3), 339–341).—[In Russian]. D., Skobets, and Ryabokon' (*ibid.*, 1947, **21**, 843) had previously measured the decompn. potentials of several metallic bromides dissolved in molten equimolecular NaBr–KBr mixture, and the present paper augments this work with measurements on $MnBr_2$, $CaBr_2$, TlBr, and $SnBr_2$ in the same solvent at 700° C. (5 mole-% soln.). The NaBr and KBr were prepared by recrystn. of analytical reagents, the TlBr was prepared by pptn. from Tl oxalate soln. with KBr, and the prepn. of the other reagents has been described elsewhere (*Zapiski Inst. Khim. Akad Nauk U.R.S.R.*, 1947, **9**, 176). The method of measurement was as described by D., S., and R. (*loc. cit.*) with a glass diaphragm and electrodes of Pt and graphite. The temp. coeff. of the decompn. potentials were measured in the range 650°–800° C. At 700° C. the decompn. potentials were: $CaBr_2$, 2·86; $MnBr_2$, 1·72; TlBr, 1·56; $SnBr_2$, 0·92 V. The temp. coeff. were: $CaBr_2$, −0·0014; $MnBr_2$, −0·0010; TlBr, −0·0010; $SnBr_2$, −0·0006 V./°C. $SbBr_3$ was found to be insoluble in the NaBr–KBr mixture. From these and previous measurements of decompn. potentials, the electrode potentials of the metals, both in the pure bromides and in soln. in NaBr–KBr mixture, were calculated by Nernst's formula. Hence the following electrochem. series were established: for the pure molten bromides at 700° C., Na, Ca, Mn, Al, Tl, Zn, Cd, Pb, Sn, Ag, Cu, Co, Bi; and for soln. in NaBr–KBr mixture at 700° C., Na, Ca, Mn, Zn, Al, Cd, Fe, Cu, Pb, Co, Sn, Ag, Ni, Hg, Bi.—G. B. H.

***Chemical Polarization by Adsorption and Cathodic Deposition of Alloys from Non-Complex Electrolytes.** M. A. Loshkarev and M. P. Grechukhina (*Zhur. Fiz. Khim.*, 1950, **24**, (12), 1502–1510).—[In Russian]. L. and G. examined the electrodeposition of Pb, Sn, and Cu on a Pt cathode at 18° C. from 0·25*N*-soln. in 1*N*-aq. phenol-sulphonic acid, with and without addn. of surface-active agents to the soln. With soln. saturated with α-naphthol and diphenylamine, and contg. 1 g./l. of gelatin, the deposition of Pb or Sn was subject to considerable chem. polarization. Sn and Pb could be deposited simultaneously from such a soln. to form a dense fine-grained alloy deposit, contg. 25·9–33·9% Sn. The cathodic polarization curve for deposition of Cu, however, was unaffected by addn. of α- and β-naphthol, thymol, xylol, and phenylamine to the soln., with or without gelatin; this anomaly may be caused by (*a*) structural peculiarities of the layer adsorbed on Cu, or (*b*) increased permeability of the layer to Cu^{++} ions, because of sp. interaction forces with the adsorbed mol. With simultaneous deposition of Cu and Sn, there was considerable inhibition of both electrode reactions, so that dense fine-grained deposits of alloys contg. 10–15% Sn could be obtained at low polarization voltages; this indicates that the lack of adsorption effects with pure Cu was caused by the discontinuity of the adsorbed layer. At higher polarization voltages, just below the limiting c.d., alloys contg. <20% Cu were deposited; this indicates that the discharge of Cu^{++} was inhibited more than that of Sn^{++} in this range. Similar results, *mutatis mutandis*, were obtained in the deposition of Cu–Pb alloys from the same electrolyte.—G. B. H.

***A New Method of Studying Cathodic Passivation.** A. T. Vagramyan and Z. A. Solov'eva (*Zhur. Fiz. Khim.*, 1950, **24**, (10), 1252–1261).—[In Russian]. When electrodeposition of a metal is interrupted for some time and then started again, the polarization voltage sometimes rises to a max. and then falls to its former steady value. This effect is attributed to passivation of the electrode surface by surface-active materials in the electrolyte, so that its effective area is reduced and the true c.d. is increased. The effect can be separated from the effects of concentration polarization by stirring the electrolyte and using a very small cathode. V. and S. built an automatic apparatus to study the effect, in which the current to an electrolytic cell is interrupted by a motor-driven commutator and a relay in the following sequence: 5 min. on, 3·5 sec. off, 5 min. on, 10 sec. off, 5 min. on, 30 sec. off, 5 min. on, 1 min. off, 5 min. on, 5 min. off, 5 min. on. The polarization voltage is measured during this process by a valve voltmeter and a short-period galvanometer (frequency 200 c./s., sensitivity $1{\cdot}5 \times 10^{-7}$ amp.). The light-spot of the galvanometer traces a record on photographic paper mounted on a rotating drum driven from the commutator shaft. By means of an electromagnetic clutch and a shutter, recording is confined to periods of 30 sec., beginning 5 sec. before the current is switched on each time. Measurements with this apparatus showed considerable passivation of Ag in 0·2*N*-$AgNO_3$ soln., but none with Ag in 0·2*N*-$KAg(CN)_2$ soln. (surface area of cathode $1{\cdot}96 \times 10^{-3}$ cm.^{2}, current $5{\cdot}6 \times 10^{-6}$ amp., temp. 18° C. in both cases). Immersion of Ag in KCN soln. "activates" it, so that passivation no longer occurs in $AgNO_3$ soln., but the effect disappears after 5–10 min. immersion in the latter soln. With Zn in a soln. of 2*N*-$ZnSO_4$ + 0·6*N*-Na_2SO_4 (surface area of cathode $7{\cdot}8 \times 10^{-3}$ cm.^{2}, current $1{\cdot}6 \times 10^{-5}$ amp., temp. 18° C.), no cathodic passivation was observed, but addn. of octyl alcohol caused passivation. By comparison of the passivation measurements with the polarization curve obtained by Kudryavtsev's method (*ibid.*, 1949, **23**, 849; *M.A.*, **19**, 853) the true c.d. and hence the effective surface area of the cathode could be calculated at each stage of passivation. It was found that with the pure electrolyte the effective area was 14·0% of the total, and that 0·0115 g./l. of octyl alcohol reduced this to 3·07% after 5 min. immersion.—G. B. H.

11 — ANALYSIS

Paper Chromatographic Analysis of Aluminium Alloys. G. Venturello and A. M. Ghe (*Anal. Chim. Acta*, 1952, **7**, (3), 268–273).—[In English]. The method described in previous work (*ibid.*, p. 261; *M.A.*, **20**, 590) is adapted to the detn. of Cu, Fe, Ni, Pb, and Zn in Al alloys. 5 ref.

—R. W. R.

***2-(*o*-Hydroxyphenyl)Benzoxazole as a Volumetric Reagent for Cadmium.** Joseph L. Walter and Henry Freiser (*Analyt. Chem.*, 1952, **24**, (12), 1985–1986).—The reagent, described by W. and F. for the grav. detn. of Cd (*ibid.*, (6), 984; *M.A.*, **20**, 191) is adapted to the volumetric detn. using a bromate-bromide titration with dead-stop end-point.—F. M. L.

***Use of Perchloric Acid in Removing Interferences Prior to the Electrolytic Determination of Lead as Lead Dioxide.** George Norwitz and Irwin Norwitz (*Metallurgia*, 1952, **46**, (278), 318).—Interference of chloride, bromide, As, Sb, Sn, and reducing organic material in the electrolytic deposition of PbO_2 is eliminated by fuming the soln. with $HClO_4$ in the presence of HBr and HNO_3. Electrolysis is carried out from a $HClO_4$–HNO_3 electrolyte.—F. M. L.

***Analytical Chemistry of Niobium and Tantalum: Atmospheric Chlorination of Hydrolysed Oxide Precipitates.** C. F. Hiskey, Leonard Newman, and R. H. Atkinson (*Analyt. Chem.*, 1952, **24**, (12), 1988–1991).—Difficulties have been encountered in the chlorination of hydrolysed and ignited precipitates of mixed earth acid oxides with titania and zirconia, using octachloropropane (cf. *ibid.*, (3), 477, 480, 484; *M.A.*, **19**, 796). The reasons have been investigated. Ignition at high temp. causes surface shrinkage and formation of new phases resistant to chlorination. Pure TiO_2 and Ta_2O_5 cannot be chlorinated completely, but addn. of a very small amount of Nb_2O_5 causes quant. chlorination. ZrO_2 inhibits chlorination. Generally, ease of chlorination increases as the acidity of the oxide increases.—F. M. L.

***Photometric Determination of Microgramme Quantities of Palladium with Beta-Furfuraldoxime.** Eugene W. Rice (*Analyt. Chem.*, 1952, **24**, (12), 1995–1997).—The pink colour produced when an acetic acid soln. of the Pd complex of β-furfuraldoxime is treated with *p*-bromoaniline is used to determine very small amounts of Pd.—F. M. L.

***Colorimetric Determination of Ruthenium with *p*-Nitrosodimethylaniline.** J. E. Currah, Alice Fischel, W. A. E. McBryde, and F. E. Beamish (*Analyt. Chem.*, 1952, **24**, (12), 1980–1982).—After sepn. of Ru by conventional distillation methods, it is determined colorimetrically by warming the soln. buffered to *p*H 4·1, with *p*-nitrosodimethylaniline.

—F. M. L.

***Determination of Small Amounts of Chlorides in Titanium Sponge.** Maurice Codell and James J. Mikula (*Analyt. Chem.*, 1952, **24**, (12), 1972–1973).—1 g. Ti is dissolved in HF and H_3BO_3 is added to the soln. to neutralize excess HF. 5 c.c. of dil. HNO_3 are added to the soln., which is warmed with const. stirring, so as to avoid hydrolysis, until it is colourless. After dilution, $AgNO_3$ is added and the soln. set aside in a dark place overnight. The precipitate of AgCl is filtered off, washed with dil. HNO_3, and then dissolved in NH_4OH. Na_2S is added to the soln., which is then made up to vol. and an aliquot taken for spectrophotometric detn. of the Ag and, hence, the Cl.—F. M. L.

***Spectral Analysis of Steels and Alloys.** K. A. Sukhenko (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 590–597).—[In Russian]. S. briefly describes a rapid method for the quant. spectrochem. analysis of high-alloy steels during melting. He also gives details of the elect. circuit, electrodes, line-pairs used, and the accuracy attained in the quant. analysis of Mg-base alloys for Al (spark), Mn (spark), Ni (spark and arc), and Ce (spark and arc). Similar data are given for the detn. of Mn, Al, Ti, Cr, and Fe in Ni-base alloys by spark spectra. In the quant. spark spectrum analysis of Al-base alloys for Mg, Mn, and Fe, the influence of Cu content on the result was shown to be negligible, since the calibration curves were the same for "Avial" alloys, contg. 0·01–0·75% Cu, and Duralumins, contg. 1·8–5·0% Cu. Similarly, it was shown that the Ni or Mn content of a bronze does not affect the calibration curve for spectrochem. detn. of Fe. Calibration curves were compared for the detn. of Cr, Ni, and Mn in Fe-, Co-, and Ni-base alloys, for several line-pairs, with the following results: (*a*) the calibration curves for Cr in Co-base alloys and in steels were parallel, but that for Cr in Ni-base alloys was not parallel to them; this discrepancy was connected with the high Si content of the Ni alloys; (*b*) the curves for Ni in Co-base alloys and in steels were parallel, but those for Mn in the same alloys were not; (*c*) curves for Fe in Co- and Ni-base alloys were parallel.

—G. B. H.

***Spectrographic Investigation of Calibration Curves for Solid Alloys and Solutions.** N. N. Sorokina (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 567–571).—[In Russian]. S. determined calibration curves for quant. spectrochem. analysis of Fe–Cr (0·34–24·4% Cr), Fe–Mn (0·48–27·88% Mn), and Ni–Cr (1·78–47·44% Cr) alloys by measurements with solid alloys, soln. of the alloys in acids, and synthetic soln. of the same compn. The alloys had been analysed chem. in several laboratories. Spectra were taken with a Feussner spark source, without an interrupter, at 10,600 V., with a capacity of 6500 cm. and an inductance of 0·012 millihenry. For solid alloys the upper electrode was a Cu rod 8–9 mm. in dia., tapering to 2 mm. dia. at the tip. For soln. two 6-mm. C rods soaked in the soln. were used as electrodes. Spectra were photographed on a Q. 24 medium-dispersion quartz spectrograph. Results are expressed as curves of $\log_{10} (I_1/I_2)$ (intensities of lines from alloying element and main constituent, resp.) against $\log_{10} C$ (concentration of alloying element in %) for the following detn.: (1) Cr in Fe from Cr 2677·159 and Fe 2684·751; (2) Cr in Fe from Cr 3124·978 and Fe 3154·206; (3) Mn in Fe from Mn 2886·678 and Fe 2880·756; (4) Mn in Fe from Mn 2939·304 and Fe 2880·756; (5) Mn in Fe from Mn 3482·909 and Fe 3465·863; (6) Cr in Ni from Cr 2668·712 and Ni 2545·900; (7) Cr in Ni from Cr 2830·468 and Ni 3012·004 Å. In all cases the curves for solid alloys and soln. were parallel, and curves for soln. of alloys and synthetic soln. were coincident.—G. B. H.

***On Displacements of Calibration Curves in Spectral Analysis.** I. S. Fishman (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 586–589).—[In Russian]. Previous measurements have shown that the slopes of calibration curves in spectrochem. analysis are affected only by the γ of the photographic plate. Curves of $(\Delta S/\gamma)$ against $\log C$ are usually parallel for all values of the other variable factors. F. studied the calibration curves for Mg 2852·13 Å. in spark spectra of specimens of alloy AL-4, and of aq. soln. contg. 4 g. NH_4 alum in 25 c.c. soln., together with small quantities of $MgSO_4.7H_2O$. For the soln. the curve changed slope at concentrations of 0·15–0·4% Mg, but for the alloys it was straight. To prove that the effect was not caused by differences in the evaporation conditions in the discharge, measurements made on the line Mg 2790·79 Å. were shown to give parallel curves for soln. and alloys. Hence F. deduces that the effect is caused by self-reversal, which depends on the physico-chem. state of the element. If this is so, the slopes of calibration curves should be the same at equal abs. line intensities, since self-reversal $\propto$ line intensity. This was shown to be true, when the two branches of the soln. curve were compared with points of equal line intensity (equal blackening with standard exposures) on curves for AL-4 alloys and Duralumin resp. Comparison of the arc and spark spectra of the AL-4 alloys showed that, the higher the abs. line intensity, the smaller the slope of the calibration curve. Hence F. deduces that, given a calibration curve for an element in one form, the calibration curve in another form can be deduced from one standard specimen.

—G. B. H.

Qualitative and Quantitative Paper Chromatography of Inorganic Ions. G. Venturello and A. M. Ghe (*Anal. Chim. Acta*, 1952, **7**, (3), 261–267).—[In English]. Chromatographic methods are described for the qual. and quant. detn. of Cu, Fe, Ni, Zn, and Pb; the chromatogram is compared with a series of standard chromatograms prepared from known soln. The chromatographic technique is that devised by Rutter (*Nature*, 1948, **161**, 435). Each of the above elements may be determined in the presence of the others. 16 ref. —R. W. R.

The Role of Micro-Analysis in Metallurgy. W. Koch (*Angew. Chem.*, 1952, **64**, (7), 187–195).—In recent years micro-chem. methods of analysis have become of increasing importance in the field of metallurgy. A description of some of the apparatus used is given, with some notes on methods of handling and the principal fields of appn. 16 ref. —W. F. H.

***Electrolytic Titration Without External Current.** W. Oelson, H. Haase, and G. Graue (*Angew. Chem.*, 1952, **64**, (3), 76–81).—In electrolytic titration a chem. compound is either neutralized, oxidized, or reduced at the cathode by an elect. current, the quantity of the substance being the product of current and time. A const. current is essential, and this involves the use of complicated apparatus. This can, however, be avoided if the current is produced in the reaction vessel itself by employing a Zn electrode, separated from the soln. by means of a diaphragm and short-circuited to a Pt electrode suspended in the soln. The strength of the current can be controlled by varying the degree of immersion of the Zn electrode.—W. F. H.

***High-Sensitivity Recording Polarograph.** Myron T. Kelley and Hugh H. Miller (*Analyt. Chem.*, 1952, **24**, (12), 1895–1899).—A polarograph intended to extend the useful range of polarography to the 10^{-6} *M* region is described.—F. M. L.

Analysis for Industry [: Use of New Reagents]. M. Kapel (*Indust. Chemist*, 1952, **28**, (333), 466–468).—The use is described of a number of new analytical reagents contg. Cu, Ni, or Ag, in which the metal is in an abnormal valency state. Methods for the detn. of Ca, As, Sb, Cr, and Pb are briefly described. 13 ref.—R. W. R.

Analysis for Industry. [: Application of Complexones.—I.–II.]. M. O. Lawson (*Indust. Chemist*, 1952, **28**, (334), 512–514; (335), 559–561).—[I.—] The appn. are discussed of ethylenediaminetetra-acetic acid (E.D.T.A.) and other complexones in the analysis of various metals including Co, Ni, Mn, Mg, W, Mo, Be, Ti, Cu, and Cr. 10 ref. [II.—] L. discusses the use of complexones in titrimetry. The titration of numerous metals with uramildiacetate or E.D.T.A. is briefly considered; the use of this reagent for the estn. of Co and Mn (in the presence of Co) is described in more detail. 10 ref.—R. W. R.

Analysis for Industry [: Redox Titrations in Non-Aqueous Media]. T. S. West (*Indust. Chemist*, 1952, **28**, (332), 415–417).—The use of redox titrations in non-aq. media for the detn. of various substances, including a number of metals, is described. 4 ref.—R. W. R.

Recent Developments in Metallurgical Analysis. I.—Electrochemical Analysis. T. S. West (*Metallurgia*, 1952, **46**, (278), 313–316).—A review, with 58 ref.—F. M. L.

Visual Lines for Spectroscopic Analysis. (Smith). See col. 607.

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13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

European Research on the Behaviour of Metals Yields Promising Results. Thomas J. Dolan (*Materials and Methods*, 1952, **35**, (6), 93–96).—D. reviews a number of fatigue-testing methods such as rotating-beam and axial-load fatigue tests at elevated temp., axial-load and torsional fatigue tests on decarburized surfaces, flexural fatigue strength of gear teeth, &c. A new Swiss creep-testing machine for testing 36 specimens simultaneously is mentioned, together with developments in other methods of testing. A comparison of American and European methods of testing is given. —R. P. H. F.

The Electrical Wire Resistance Strain Gauge and Its Applications. F. C. Widdis (*Machinery* (*Lond.*), 1950, **77**, (1975), 259–264).—The operation and types of resistance strain gauges are briefly described, and the effects of humidity and temp. on them are indicated. The methods of measuring the electrical signals for both static and dynamic strain measurements are also described.—S. R. W.

Requirements of Machine-Shop Inspection [of Castings]. C. Walker and F. H. Greenwood (*Found. Trade J.*, 1952, **92**, (1860), 433–439).—Read before the East Midland Branch of the Institute of British Foundrymen. The first part of the paper is mainly concerned with the machinist's viewpoint of the quality of castings, from the design, jigging, accuracy, machinability, and inspection angles. The second section consists of examples of inspection methods and illustrations of jigs, fixtures, and records which are kept.—G. J. S.

The Principles of Technical Control in Metallurgical Manufacture. A. R. E. Singer (*J. Inst. Metals*, 1952–53, **81**, (7), 329–340).—The effects of economic and tech. factors on the quality of manufactured products are examined, and the characteristics of controlled processes analysed in terms of basic principles. Some aspects of statistical control of plant are considered in relation to present practice and probable future developments.—Author.

***Brightness Measurements for the Evaluation of Metal Surface Quality.** Joh. Elze and H. Grüss (*Metalloberfläche*, 1952, [A], **6**, (2), 17–23).—A description is given of the construction and operation of an instrument—consisting of a light source, series of lenses, and a photocell—for measuring the intensity of specular and diffusion reflection, and also any astigmatism of such diffuse reflection caused by bending of the specimen during previous working processes (e.g. polishing). The instrument gives very precise results which are completely reproducible. It can be used for the evaluation of electroplated or polished surfaces, and for following the tarnishing of metal surfaces—by measuring the decrease of reflectivity with increasing time of exposure to the corrosive atmosphere.—E. N.

***Study of the Distribution of Surface Potential by Means of Radioactive Deposits.** T. Westermark and L.-G. Erwall (*Nature*, 1952, **169**, (4304), 703–704).—Autoradiographs of pure metal surfaces which had previously been exposed in the presence of an elect. field (50 V./cm.) to emanations and recoil atoms from a radio-Th source, showed a uniform activity over the surface. A Pb plate with various metals embedded in it, and treated in the same way, gave an autoradiograph with blackened regions near to the embedded metals when the source was positive relative to the plate. Reversal of the field produced reversal of the autoradiograph. It is suggested that positive ions travel along the lines of force and are deposited on negative (high-work-function) regions, when the plate is negative. The method is applicable to the study of minerals, biological materials, and alloys. —R. S. B.

Ultrasonic Inspection Checks Quality of Brazed Joints. H. Greenberg (*Materials and Methods*, 1952, **35**, (6), 102–105).—The ultrasonic testing of brazed joints between Cupaloy (i.e. an age-hardening alloy contg. Cu 99·5 and Cr 0·5%) and a contact alloy (i.e. W 65 and Ag 35%) is discussed, and the close correlation between ultrasonic testing and destructive shear testing illustrated. The advantages and disadvantages of ultrasonic methods for testing rotors, thrust-bearing shoes, turbine discs, and miscellaneous forgings of all types are given.—R. P. H. F.

Black Light Spots Defects in Materials. N. E. Walters (*Materials and Methods*, 1952, **36**, (5), 105–107).—The use of ultra-violet light in the non-destructive testing of metals and non-metals, after processing with special fluorescent soln., is discussed. Among the advantages claimed for this method of testing are flexibility, economy, speed, and sensitivity. The methods of inspection, sources of ultra-violet light and its appn. are reviewed.—R. P. H. F.

Where Radioisotopes Are Finding Industrial Use. Philip O'Keefe (*Materials and Methods*, 1952, 36, (3), 87–89).—The appn. of radioactive tracers and radiation sources to the measurement of thickness, wear testing, radiography, chem. analysis, and surface-cleaning methods are reviewed. The use of absorption and backscatter gauges using Co^{60} and Sr^{90} are described.—R. P. H. F.

15 — FOUNDRY PRACTICE AND APPLIANCES

Aluminium Castings for Polished and Anodized Articles, Especially Fittings. Maria Buckeley-Liesegang and August Buckeley (*Metalloberfläche*, 1952, [A], 6, (3), 33–41).—Brief notes are given on the compn. of Al alloys suitable for the sand- and die-casting of door fittings which are to have an anodized finish. This is followed by an illustrated review of the various types of casting defects and their relation to faults and blemishes in the surface appearance of the finished articles. Speedy identification of their origin is the most effective method for their elimination. Ref. is made to some recent or little-known methods for improving the final surface finish, viz. (1) special lacquers, and (2) Ni and/or Cr plating—more particularly when the fittings are for use in the transport industry. 8 ref.—E. N.

Aluminium Alloy Die-Castings Used in Complex Automotive Parts. T. C. Du Mond (*Materials and Methods*, 1952, 35, (1), 80–81).—The reasons for adopting the die-casting technique in preference to the plaster-mould casting and other methods, for the prodn. of 30 different parts in Al–Si A.S.T.M. alloy B 85 and S.A.E. alloy E 308 for the new Ford automatic transmission system, are discussed. It is claimed that die-castings give satisfactory components with thin walls, sharp corners, and close tolerances. The deciding factors in choosing Al alloys is attributed to their high thermal conductivity, lack of resonance, good corrosion-resistance, excellent machinability, and casting properties.—R. P. H. F.

The Control of Quality in the Melting and Casting of Aluminium Alloys for Working. R. T. Staples and H. J. Hurst (*J. Inst. Metals*, 1952–53, 81, (7), 377–391).—The special problems arising in the melting and casting of Al alloys and the best methods of quality control are considered in relation to 9 standard groups of alloys intended for use in wrought forms. The effects of the alloying addn. and of impurities and incidental contaminants on foundry practice, on the setting up of standards of quality, and on the ultimate properties of the material, are treated under the sp. alloy groups. The limitations of the melting and casting processes available, and in particular of the direct-chill method, are outlined, and the degree of control which it is possible to exercise through routine phys. and chem. inspection at the various stages is indicated.—AUTHORS.

History of Aluminium-Alloy Casting. A. P. Fenn (*Found. Trade J.*, 1952, 92, (1852), 223–228).—Read before the London Branch of the Institute of British Foundrymen. F. reviews the developments that have taken place in the Al industry in the last 40 years. The moulding and pouring of a 4¾-ton Al casting is described and illustrated.—G. J. S.

Phenomenal Growth in Light Alloys. E. G. West (*Found. Trade J.*, 1952, 92, (1854A), 65–71).—W. traces the development of light-alloy founding through the past 50 years and lists several met. problems which have been overcome.
—G. J. S.

The Control of Quality in Melting and Casting Copper and High-Conductivity Copper-Base Alloys. J. Sykes (*J. Inst. Metals*, 1952–53, 81, (7), 351–363).—Control is necessary at all stages during the prodn. of high-conductivity Cu shapes for subsequent working. The selection of raw materials and their blending into suitable furnace charges are discussed in relation to the type of furnace to be used. The fire-refining process for the prodn. of high-conductivity Cu wire-bars is outlined, and control methods are described in detail. Photographs show the changes occurring in the structure of button and tube samples during the course of oxidation and the later stages of the reduction process. The removal of As by soda ash–lime fluxing is described, and a method is given for its control by conductivity testing. Desirable qualities required for mould dressings are indicated, and ref. is made to the influence of pouring speed and O content of the Cu on the set of the final casting. Typical pouring temp. and speeds are given. Methods used in the prodn. of P-deoxidized Cu and Cu–Te, Cu–Cd, and Cu–Ag alloys are outlined, with ref. to the time of alloy addn. and the control of O content in the bath. Typical casting defects and their causes are listed, and inspection and analytical methods outlined.—AUTHOR.

How to Use Copper Alloys in Die-Castings. L. F. Spencer (*Iron Age*, 1952, 169, (2), 90–91; also *Found. Trade J.*, 1952, 92, (1851), 205–206).—S. gives the chem. composition and mech. properties of 4 Cu-base die-casting alloys. The effect of several elements and various size and design limitations are described. The chem. compositions of two suitable die materials are also given.—G. J. S.

Production of Castings in Aluminium Bronze, D.T.D. 412. L. Hargreaves (*Found. Trade J.*, 1952, 92, (1862), 487–489).—The specification D.T.D. 412 is given and H. suggests raising the Mn content to 1·5%, keeping the Al content at 11%, and casting the test-bars separately in a cast-Fe chill mould. The following founding procedure is recommended: a normal non-ferrous foundry sand with a moisture content of 8·5%; gating by means of a reverse horn-gate, chills in the mould, and the use of exothermic sleeves. A high pouring temp., 1200° C., is advised for thin-sectioned castings.—G. J. S.

Some Casting Problems of Copper–Lead Alloys. P. D. Liddiard and P. G. Forrester (*Found. Trade J.*, 1952, 92, (1863), 509–515).—Paper read to a Joint meeting of the London Branch of the Institute of British Foundrymen and the Institute of Metals. L. and F. consider the Cu–Pb equilibrium diagram and discuss the effect of other elements on the form of the miscibility gap, and some of the problems that arise. The controversial subject of the efficacy of a fine dispersion of Pb in a Cu matrix is examined, and the three known methods of lining bearings by (*a*) static, (*b*) centrifugal and (*c*) continuous strip lining processes are summarized. Casting problems such as adhesion of bearing metal to the steel strip, Pb dispersion and shrinkage, and blow holes are described. The effect of an addn. of a small quantity of another element is examined briefly.—G. J. S.

The Control of Quality in the Production of Brass Ingots and Billets. Maurice Cook and C. L. M. Cowley (*J. Inst. Metals*, 1952–53, 81, (7), 341–350).—After ref. to relevant general considerations important in the formulation of quality-control procedures, the paper is mainly concerned with the appn. of basic principles and practical knowledge to eliminate or minimize the incidence of defects and so ensure the regular prodn. of high-quality brass castings for the manufacture of alloys of this type in wrought forms. The control of raw materials used in their prodn. is dealt with, and the various factors involved in melting processes are considered with particular ref. to L.F. elect.-furnace practice. Following some general observations on casting which include data concerning casting temp. for different alloys and comments on cast structures, the significant factors relating to moulds and mould materials are reviewed. Pouring and the subsequent operation of feeding, which together constitute the process of casting, are in turn considered, and finally brief ref. is made to inspection of cast products.—AUTHORS.

Better Brass-Mill Products Produced by Continuous Casting. Edward L. Wolff and Frederick M. Barry (*Materials and Methods*, 1952, 35, (4), 94–97).—The continuous-casting machine described consists of three 10,000-lb./hr. Ajax–Scosnet induction melting furnaces, a 30,000-lb./hr. brass holding furnace, a charging crane, a crane ladle, a casting machine, cut-off saw, and roller delivery table. The casting machine produces a 7¾-in.-dia. billet, 31½ in. long, every 2 min., or a 2200 lb. slab, 25½ in. wide by 2½ in. thick and 10 ft. long, every 6 min. The following advantages are claimed for the process: uniformity of compn. and structure, good mech. properties, excellent free-flowing and hot-working characteristics, and good machinability of leaded alloy due to the uniform distribution of Pb.—R. P. H. F.

Consolidation of Bronze and Brass Founding. Frank Hudson (*Found. Trade J.*, 1952, 92, (1854A), 49–55).—H. discusses the progress in layout and techniques in bronze and brass founding during the past 50 years.—G. J. S.

The Control of Quality in Melting and Casting Magnesium Alloys for Hot Working. R. G. Wilkinson and S. B. Hirst (*J. Inst. Metals*, 1952–53, 81, (7), 393–400).—A description is given of present practices in the control of quality and inspection in the melting and casting of the three principal types of Mg alloy used for wrought purposes, viz. Mg–Mn, Mg–Al–Zn–Mn, Mg–Zr–Zn. The Zr-contg. alloys are considered separately, not only because special procedures are involved in their melting, but also because they are always cast by the semi-continuous direct-chill process. Except on a laboratory scale, rolling, extrusion, and forging are not carried out in the authors' works, and thus the use of commercial-scale hot working as a running routine "inspection" method is not possible. The inspection processes applied before the cast stock is sent out to other works are therefore crucial.—AUTHORS.

*__Origin of Porosity in Castings of Magnesium–Aluminium and Other Alloys.__ E. J. Whittenberger and F. N. Rhines (*J. Metals*, 1952, 4, (4), 409–420; discussion, (11), 1189–1190).—A mechanism for the formation of casting porosity is proposed, which is based on a nucleation and growth process with shrinkage and gas pptn. as co-operative driving forces. The principal features of the process are: (1) Single nucleation is produced by the hydrostatic tension developed by shrinkage; this nucleus, according to its location, grows into either an internal pipe or a type of interdendritic pore that may be sub-microscopic. (2) Multiple nucleation is produced by gas pptn., giving rise to visible microporosity in all its varieties. (3) Late nucleation occurs mainly in the intergranular channels, where the highest concentration of gas is developed, and microporosity is, accordingly, mainly intergranular. (4) The mode of growth is determined by whether or not liquid feeds to the interior of the casting; if so, pptn. of gas controls cavity growth; if not, shrinkage determines the extent of growth, but gas pptn. determines the relative rates of growth among the cavities that are in hydrostatic communication. (5) The linear growth of gas pores may be decelerating or sustaining, depending upon the restraint imposed by the solid phases; where the solid phases offer so little restriction that growth is essentially three-dimensional, the growth rate diminishes as the gas content of the adjacent metal declines and isolated pores are formed; if the solid phases restrict growth essentially to one dimension, so that the bubble advances continually into undepleted liquid, growth is sustaining, and a high degree of restraint by the solid phases results in channel porosity. The major premise that microporosity is nucleated by gas pptn. is confirmed by experiments which have been carried out by induction melting, under controlled atmospheres, of binary alloys of: (*a*) Al with Ag, Cu, Mg, Si, and Zn, (*b*) Cu with Ni, Pb, Sb, Si, Sn, and Zn, (*c*) Mg with Sn, and Zn, and (*d*) Bi with Sb. The results show that: (I) Visible microporosity does not form in completely degassed Al–Cu alloys, Mg, or Mg–Al alloys upon solidification from the melt. (II) Gases capable of producing visible porosity in Mg and its alloys, listed in their descending order of effectiveness, are: wet H, dry H, H_2S, and H_2O vapour (all forming gas holes and channels); CH_4, dry N, dry SO_2, and dry CO_2 (which form channels); wet A, which produces channels, but no visible porosity is observed with the dry gas. (III) Visible microporosity, due to gas evolution, is observed in a Bi–Sb (90 : 10) alloy which expands on freezing. (IV) In visually sound castings of Mg and its alloys, there is definite evidence of an invisible (sub-microscopic) porosity amounting to a few hundredths to a few tenths of 1%, and which can be caused to disappear by working and annealing. (V) The vol. of gas porosity varies inversely with the freezing velocity, and increases with the temp. range of freezing of the alloys. (VI) Parallel curves are obtained on plotting alloy compn. against porosity, and against freezing range. (VII) Microporosity is intergranular, with only a slight penetration among the dendrite arms. It is pointed out that the metallographic method of determining porosity is relatively insensitive and, therefore, unreliable due to smearing; the most sensitive methods are microradiology, examination of fractured surfaces, and, for quant. measurements, *d* detn. 17 ref.—E. N.

The Control of Quality in the Casting of Zinc and Zinc Alloy Rolling Slabs and Extrusion Billets. C. W. Roberts and B. Walters (*J. Inst. Metals*, 1952–53, 81, (7), 365–375).—The various techniques and procedures adopted commercially for the prodn. of rolling slabs and extrusion billets in Zn and Zn alloys are briefly described, and the met. principles involved are examined. The origins of the defects normally encountered in the castings are indicated, and the means adopted to minimize the incidence of such defects are discussed. The importance of control of chem. compn. both in raw materials and in the final products is emphasized, and it is concluded that, although defects can be associated with particular methods of casting, several methods are available whereby satisfactory slabs or billets can be produced provided that supervision of the operations is adequate.—AUTHORS.

*__Approach to Melting Reactive Metals Eliminates Use of Refractory Containers.__ T. T. Magel, P. A. Kulin, and A. R. Kaufmann (*J. Metals*, 1952, 4, (12), 1286–1288).—See *M.A.*, 20, 203.—E. N.

Basic Principles of [Die-Casting] Die Design. [—I.–II.] H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 76, (1948), 273–280; (1953), 468–470).—[I.—] The essential features of the casting cycles with goose-neck and cold-chamber machines are outlined, and the characteristics of centre-gated and edge-gated dies are briefly indicated. The disposition of the castings in the die is shown to depend on the "effective working area", the factors governing which are discussed. Various methods of arranging parting lines in a simple two-piece die are considered. [II.—] The design of ejector systems to produce a balanced thrust, the choice of registration methods, the importance of arranging the parting line with regard to trimming operations, and the need for anchoring the casting to the ejector half of the tool during opening, are discussed.—J. C. C.

Basic Principles of [Die-Casting] Die Design: Sprues, Runners, and Gates for Die-Castings. H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 76, (1957), 608–613).—The influence upon the quality of the castings produced, of the form and disposition of the runners in relation to the cavities is considered, and the way they are connected to the feed and gate is indicated. The proportions of the sprue and the position of the gate are important.—S. R. W.

Basic Principles of [Die-Casting] Die Design: Venting Arrangements and Ejector Mechanisms. H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 76, (1961), 752–759).—Good practice in venting (which, when it can be provided, makes for sounder castings), the use of a rubber insert, multiple centre gating, the design of ejector mechanisms and pins, and the causes of buckling of ejector shanks, are among the matters discussed.—J. C. C.

Basic Principles of [Die-Casting] Die Design: Ejector Mechanisms. H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 76, (1966), 927–932).—A short description is given of two-stage ejection, non-circulator ejectors, sleeve ejectors, ring ejection, stripper-plate ejection, and oblique ejection. —S. R. W.

Basic Principles of [Die-Casting] Die Design: Cores and Core Mechanisms. [—I.–III.] H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 77, (1970), 113–120; (1975), 268–274; (1978), 363–365).—[I.—] The advantages and limitations of cavities made by the monobloc and the built-up

construction are discussed. The mounting of cores and inserts in the cavities is described in some detail. [II.—] The advantages and disadvantages, together with the restrictions imposed by the machine and core construction, of the methods of moving cores, are given. Examples involving the use of finger cams, cams and rollers, toggle mechanisms, and racks and pinions are considered. [III.—] The use of hydraulic core withdrawal and the difficulties encountered with curved and helical cores are discussed.
—S. R. W.

Basic Principles of [Die-Casting] Die Design: [I.—] Disposition of the Die Cavity. [II.—] The Use of Loose Die Inserts. H. K. and L. C. Barton (*Machinery* (*Lond.*), 1950, 77, (1985), 568–574; (1989), 727–731).—[I.—] Alternative arrangements for casting a simple component in edge- or centre-gated machines and in a six-cavity die are discussed. [II.—] Examples of the use of loose inserts when casting screw threads are described. The use of a number of threaded cavities in a single loose die element, and of a single loose cavity plate for very small threaded parts are also discussed. Ref. is made to the use of loose plates for loading large numbers of studs and for centre-gating individual cavities. Other methods of handling small inserts are illustrated.
—J. C. C.

Basic Principles of [Die-Casting] Die Design: The Location of Inserts in Die-Casting Dies. H. K. and L. C. Barton (*Machinery* (*Lond.*), 1951, 78, (1993), 157–164; (1997), 335–342).—Methods of locating inserts of many different forms in die-casting dies are surveyed.—J. C. C.

Experiences with the Investment-Casting Process. D. F. B. Tedds (*Found. Trade J.*, 1952, 92, (1855), 297–310).—Read before the Birmingham Branch of the Institute of British Foundrymen. After discussing briefly the possibilities and limitations of the process, T. gives detailed examples of special components for which it has been adapted on an experimental basis. Various features of investment casting are taken severally, and the present state and lines of possible future development are discussed at some length. Three tables listing pattern, mould and casting defects, resp., are given.—G. J. S.

Symposium on Vermiculite in the Foundry. [I.—Nature and Uses. II.—For Ferrous and Non-Ferrous Castings. III—For Steel Castings.] (I) J. E. Laschinger, (II) G. Butler, (III) A. Hirst (*Found. Trade J.*, 1952, 92, (1847), 89–92).—Organized by the South African Branch of the Institute of British Foundrymen. [I.—] L. describes the structure, possible chem. compn., and some commercial uses of the material. [II.—] B. discusses the uses of vermiculite in non-ferrous founding. A mixture of 50 : 50 red sand and vermiculite is a good lagging and lining material for coke-fired furnaces. Vermiculite is also used as a mould, chill, and core wash, and it is claimed to eliminate penetration in phosphor-bronze castings when molasses is used as a binder, and to lengthen the life of chills. Further appn. are: (i) filter screens—the use of metal gauze vermiculite-sprayed, enables feeder heads to be knocked off without damaging the casting, and (ii) feeding flux-vermiculite sleeve round the riser. [III.—] H. describes uses in steel founding; 50 : 50 mixture of vermiculite and moulding sand is used for feeder cups and considerably reduces the weight of the feeder head required. H. suggests using bentonite as a bond.
—G. J. S.

" Perpetual Interest " of Non-Ferrous [Foundry] Research. G. L. Bailey (*Found. Trade J.*, 1952, 92, (1854A), 39–43).—B. indicates the advancement of knowledge as a result of research work. Several major problems, including solidification of alloys, feeding, and grain refinement, are discussed, and the results of recent investigations are given briefly. 30 ref.
—G. J. S.

Further Mechanical Aids for the Foundry. A. S. Beech (*Found. Trade J.*, 1952, 92, (1854), 279–283; (1856), 331–335).—B. describes new mech. and semi-mech. devices put into practice in the foundry between 1942 and 1949. Mass-prodn. plants, sand prepn., and the appn. of mechanization to the jobbing foundry are discussed and illustrated.—G. J. S.

Ways and Means to Increased Productivity [in the Foundry]. John Hunter (*Found. Trade J.*, 1952, 92, (1861), 457–462).—Paper read before the Scottish Branch of the Institute of British Foundrymen. H. lists the general factors affecting productivity in the foundry industry. The possible improvements in the use of patterns, moulding boxes, manual effort, conveyors, overhead hoppers, sand slingers, and rubbish disposal are examined in detail, and the inter-relation of plant utilization is emphasized.—G. J. S.

Evolution of the [Foundry] Equipment Industry. J. Blakiston (*Found. Trade J.*, 1952, 92, (1854A), 23–33).—B. describes and illustrates the development of foundry equipment during the past 50 years.—G. J. S.

Cross-Section of a Non-Ferrous Jobbing Foundry. J. Gorman (*Found. Trade J.*, 1952, 92, (1851), 189–199; (1852), 229–231).—Read before the Scottish Branch of the Institute of British Foundrymen. G. describes the layout and plant in a non-ferrous jobbing foundry. Techniques used for making several types of castings, including the lost-wax process, are described and illustrated.—G. J. S.

[50 Years' Development in the Foundry]. From the Editor's Viewpoint. V. C. Faulkner (*Found. Trade J.*, 1952, 92, (1854A), 9–16).—F. reviews developments in the foundry industry, both ferrous and non-ferrous, during the past 50 years.—G. J. S.

Gestaltung von Druckgussteilen (Spritz- und Pressguss). (Lieby). See col. 608.

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16 — SECONDARY METALS: SCRAP, RESIDUES, &c.

[Secondary] Copper-Base Non-Ferrous Ingots. W. G. Mochrie (*Found. Trade J.*, 1952, 92, (1848), 121–124).—M. discusses various refining problems encountered in reclaiming non-ferrous scrap. The identification and removal of various undesirable impurities from the secondary metals is described.
—G. J. S.

***High-Purity Sponge-Like Copper from Waste Pickling Sludge.** Robert L. Rusher and George W. Blum (*Indust. and Eng. Chem.*, 1952, 44, (3), 660–664).—A highly acidic metallic salt sludge produced in the acid pickling baths of brass and Cu products, which constitutes a severe water-pollution problem, was used in the investigation. A method based on the Cu/Zn displacement reaction is described for recovery of Cu in sponge-like, highly surface-active form, suitable for use as a catalyst in gas- and liq.-phase reactions. The economic feasibility of the work primarily undertaken as a laboratory investigation, is considered. 14 ref.—D. P. M.

***Liquid Industrial Wastes. Brass and Copper Industry: Cation Exchanges for Metals. Concentration from Pickle Rinse Waters.** F. X. McGarvey, R. E. Tenhoor, and R. P. Nevers (*Indust. and Eng. Chem.*, 1952, 44, (3), 534–541).—Presented at an American Chemical Society Symposium on Liquid Industrial Wastes. The concentration and recovery of alloy metals from the dil. pickle rinse soln. encountered in the Cu and brass industries has presented a problem for many years, which has not been solved by the conventional pptn. and coagulation procedures. Recent studies on typical mill rinses have shown that cation exchangers operating in the Na and H cycle might be employed successfully to concentrate Cu, Zn, and Cr soln. The economics of the exchange in the H cycle indicate that profitable recovery of Cu might be accomplished in conjunction with an electrolytic deposition process in the Cu pickle itself. Flow rate and influent concentration have a marked effect on the economics.—D. P. M.

Recovery of Copper and Sulphuric Acid from Pickling Baths. —— (*Electroplating*, 1952, 5, (7), 227).—Details are given of a unit capable of recovering 2½ tons of Cu/year from H_2SO_4 baths used for pickling Cu. This consists of an electrolytic recovery cell through which the soln. is pumped by means of a centrifugal glandless vertical pump and thence to the pickling tank.—H. A. H.

Gold Scrap Recovery Methods Applicable to Scarce and Costly Materials. Bernard Blackman and Carl Heinzman (*Materials and Methods*, 1952, 35, (3), 102–104).—B. and H. claim that the difference between profit and loss in the manufacture of watch cases as a whole is determined by the percentage of less easily reclaimed scrap. The daily burning of all combustible material (e.g. emery cloths, buffs, brushes, &c.), the pptn. and filtering of all wash water from watch-cleaning baths, wash basins, and laundry, &c., and the leaching of any Au remaining in soln. from plating baths, &c., are the methods of recovery reviewed.—R. P. H. F.

The Sorting and Sampling of [Scrap] Metals [and Alloys]. Josef Blanderer (*Z. Erzberg. u. Metallhüttenwesen*, 1952, 5, (7), 257–262; discussion, 262–263).—The successful recovery of scrap metals and alloys depends, to a large extent, on accurate sampling of the bulk material. This almost invariably necessitates sorting, which, after magnetic removal of Ni and tramp Fe, can be done, very efficiently, by a skilled operator noting variations in: (1) colour—for differentiating between Al, Pb, Sn, Zn, and their alloys, and for Cu–Zn alloys of varying compn., and (2) form—which enables castings, forgings, extrusions, and various types of turnings to be differentiated, thereby, also, giving a clue to their compn. B. deals with these aspects of sorting at some length. In the discussion the use of portable spectroscopes (modified steeloscopes) and of potentiometric methods is mentioned.—E. N.

Metal Recovery by Ion-Exchange. C. F. Paulson (*Plating*, 1952, 39, (12), 1330–1334, 1338).—Cf. *M.A.*, 20, 36. The recovery of various metals by ion-exchange from a number of indust. wastes, including the wastes from plating and anodizing processes, is described. 8 ref.—G. T. C.

19 — WORKING

High-Strength Aluminium Impact Extrusions. M. P. Meinel (*Materials and Methods*, 1952, 36, (5), 110–113).—M. discusses recent improvements in the techniques of impact extrusion which have led to the prodn. of a variety of shapes in high-strength Al alloys. Factors influencing design, such as reduction of area, capacity of press, overall length of component, wall thickness, blank size, &c., are reviewed. The efficient design of dies and punches is stressed, and the disadvantages of impact extrusion at elevated temp. are enumerated. Punch load/reduction of area curves for various alloys, together with their extrusion characteristics, are given.—R. P. H. F.

Impact Extrusion [of Aluminium and Zinc]. P. Mabb (*Machinery* (*Lond.*), 1950, 76, (1946), 192–196).—The steel for the punches and dies, the lubrication, and the mechanism of the process are discussed. Only pure Al and Zn are impact extruded on any appreciable scale, and a number of micrographs and hardness diagrams indicate the effect of this process on these metals.—S. R. W.

Use of Densified-Wood Dies for Forming and Drawing Aluminium. J. Milliken (*Machinery* (*Lond.*), 1950, 76, (1957), 598–601).—Densified-wood dies have high strength (15,000 lb./in.^{2}) and can be quickly and easily made or altered. The low coeff. of friction avoids scratching of the metal being formed, and ensures even metal distribution. A short description of the working and the appn. of the wood dies is given.—S. R. W.

The Machining of [Light Metal] Pistons for I.C. Engines. H. Barth (*Automobiltechn. Z.*, 1952, 54, (4), 88–90).—A critical survey of methods employed for machining light metal pistons, showing the importance of dimensional accuracy and surface finish in general and for certain parts in particular. Special methods of measuring accuracy are described, including light-interference gauges, and various methods of attaining the high degree of accuracy specified.—W. F. H.

***Influence of Surface Activity of the Liquid Medium Upon Cutting Processes and Machinability of Metals [Aluminium].** N. A. Pleteneva and P. A. Rehbinder (*Doklady Akad. Nauk S.S.S.R.*, 1948, 62, (4), 501–504).—[In Russian]. It has been shown that at high values of initial sp. work of drilling Al, in a non-polar medium, under sufficiently high axial pressures, the addn. of surface-active substances to hydrocarbon liq. lubricants, e.g. petrolatum oil or kerosene, effects a very considerable reduction in the work of drilling. Thus the addn. of 0·25–0·50% of cetyl alcohol to kerosene reduces the work of drilling Al to $\frac{1}{8}$th. There exist two sharply delimited ranges of influence of the surface-active components of the lubricant upon the deformation of metal in layers adjoining the surface. At small depths there occurs the effect of adsorption-conditioned reduction in the effort of deformation. At greater depths addn. of a surface-active substance have the reverse effect of facilitating sliding of the cutting tool, and the usual lubricating action predominates. Additives which increase the oiliness of a lubricant will, under high pressures, produce a consistent increase in wear. The critical pressure which delimits the two ranges will, in a series of metals, vary according to the Y.P. or other strength characteristic of the metal. When an optimum addn. of surface-active substance is introduced into non-polar kerosene, the metal subjected to mech. working acquires a much smoother and more brilliant surface.—J. S. G. T.

***Strain-Hardening of Aluminium During Cutting in Inactive and Surface-Active Cutting Fluids.** N. A. Pleteneva, L. A. Shreiner, and P. A. Rehbinder (*Doklady Akad. Nauk S.S.S.R.*, 1948, 62, (5), 653–655).—[In Russian]. The strain-hardening of annealed Al, in the process of cutting, was determined by measuring the micro-hardness, H_μ, of the flat surface at the bottom of a cylindrical hole drilled with a plume-shaped cemented-carbide drill fed under a const. axial thrust. The results showed that with an initial H_μ of the Al of 24 kg./mm.^{2}, H_μ of the metal drilled in the presence of non-polar kerosene reached 115 kg./mm.^{2} and more, whereas in the presence of kerosene contg. the optimum concentration of a surface-active substance it was only 57–65 kg./mm.^{2}. The relation between the degree of strain-hardening and concentration of the surface-active substance was studied by drilling Al in pure toluene and in soln. of stearic acid in toluene. An increase in the concentration of stearic acid produced a reduction in micro-hardness, and hence a lower degree of strain-hardening. During cutting in a non-polar medium, the sp. work done increases sharply with hole depth, whereas only a slight increase of sp. work occurs in an active medium. The increase in sp. work is associated with an increase of strain-hardening. H_μ, after drilling in non-polar kerosene increases with increasing hole depth; during drilling with surface-active substances H_μ is const. The major part of the strain-hardening effect manifests itself during the first turn of the drill. Mechanisms of the strain-hardening produced in drilling in active and inactive media are suggested. —J. S. G. T.

Fabricating from Continuous [Tinplate] Strip. W. Richards (*Tin-Printer*, 1950, 26, (306), 5–6).—The advantages of continuous coils of tinplate are considered.—S. R. W.

Some Hints on Fabricating Titanium. Thomas A. Dickinson (*Steel Processing*, 1952, 38, (7), 333–334, 339).—A short general review, including a table of phys. properties of Ti and a table showing typical compn. of Ti alloys.—S. R. W.

Hot Stamping the "Superalloys". William Van den Akker (*Aero Digest*, 1950, 60, (6), 33, 83).—"Superalloys" are defined as alloys with or without predominant quantities of Fe and usually contain a relatively large number of other alloying elements. "Kirksite" tooling can be used to stamp superalloys preheated to $>$2000° F. (1095° C.) when: (1) the punches and dies are shielded with viscous lubricants, (2) stock materials are handled with tongs and similar types of forging tools so that the preheated stock materials will be in contact with the dies for no more than a few sec. during each stamping impact, and (3) stampings are carefully cleaned and pickled so that any extraneous metal particles adhering to them will be removed; the pickling soln. comprises 20% HNO_3 and 2% HF heated to a temp. of 160° F. (71° C.).—H. Pl.

***Forming of a Plastic Sheet Between Fixed Cylindrical Guides with Coulomb Friction.** H. I. Ansoff (*Trans. Amer. Soc. Mech. Eng.*, 1950, 72, (1), 145–152).—The purpose of the investigation described was to obtain a rigorous analysis of the drawing of a plastic sheet between two fixed cylindrical surfaces. In this treatment the assumption is made that the material between the cylinders is in a state of plane plastic flow and behaves according to the Saint Venant–von Mises theory.—H. Pl.

***Aqueous Colloidal Clay Slurries as Lubricants and Coolants in the Deep Drawing of Metals.** S. Ya. Veiler, N. N. Petrova, and P. A. Rehbinder (*Izvest. Akad. Nauk S.S.S.R.*, **1946**, [Tekhn.], (4), 625–628).—[In Russian]. The deformation of metals in the process of deep drawing is considerably facilitated by the presence of small amounts of surface-active substances in the lubricants and coolants. Addn. of graphite, C black, S, chalk, talc, mica, ZnO, and colloidal suspensions of Pb have been so used. It has now been shown that the addn. of bentonite or Florida clays in the form of aq. colloidal suspensions considerably improves the processes involved in metal drawing by extending the tool life, improving the quality of the product, and reducing considerably the consumption of the surface-active component, soap.—J. S. G. T.

***A New Method for Studying the Effectiveness of Lubricants in Deep Drawing.** S. Ya. Veiler and L. A. Shreiner (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (2), 325–328).—[In Russian]. A simple and convenient method of studying the effectiveness of lubricants in deep-drawing experiments is described. Cylindrical Cu specimens of dia. 20 mm. and height 15 mm., with a hole of radius 6·200 mm. drilled through the axis, were prepared. A steel ball of radius 6·316 mm. was pressed through the hole and the force necessary to do this was measured. After the first passage, which required 705 kg., when the extra surface layer of Cu was forced out, the ball was pressed either dry or with a lubricant of 5% soap (59% fatty acid) in aq. soln. The force necessary for the second passage was 138 kg. without and 80 kg. with lubricant. After 14 passages without and 5 with lubricant, the forces became const. at 72 and 45 kg., resp. This was interpreted as the number of passages required to complete the process of plastic flow, leaving only elastic deformation to oppose the movement of the ball. It was concluded that the effect of the lubricant was to ease the process of plastic deformation and to lower the frictional force. Similar experiments on brass gave essentially the same results.—Z. S. B.

[Making] Collapsible Tubes. —— (*Tin-Printer*, 1950, **26**, (306), 8–9).—A description of a new process devised by the American Type Founders Inc., U.S.A., of producing tubes from foil, laminated or coated on both sides with plastic. The tubes are known as ATF tubes. Cf. *M.A.*, 18, 308; 19, 82.—S. R. W.

Non-Metallic Dies for Aircraft Fabrication. J. Delmonte (*Aero Digest*, 1950, **60**, (1), 52–53, 97).—A table is given which presents a summary of the properties of flexible and semi-rigid thermoplastic materials of high impact resistance which are used for non-metallic dies. Strong durable tools having complex contoured surfaces can be prepared in one day. Satisfactory matte reinforced Calcerite dies have been used for elevated-temp. tooling.—H. Pl.

Notes on Press-Tool Maintenance. John W. Langton (*Tin-Printer*, 1950, **26**, (307), 8–9; (308), 7–8).—Notes on a system for tool maintenance and the min. toolroom plant essential are given.—S. R. W.

Mass-Producing Turbine Blades. —— (*Aeroplane*, 1951, **80**, (2072), 417).—Hamilton Standard have developed a machine which uses an endless abrasive belt, the surface of which is varied in consistency, for blade grinding. Blade surfaces may be finished on the machine after rough grinding, or alternatively blades may be processed from rough forgings. The machine will remove up to 3/32 in. of stock in 15 min. or less. Stock is removed accurately at high speed without distorting the blade or setting up damaging surface stresses. The machine works on the same principle as the grinder used by the firm during the war for the high-speed contour grinding of solid Duralumin airscrew blades.—H. Pl.

Hot-Machining Methods for Difficult-to-Machine Metals. A. A. Gaminada (*Materials and Methods*, 1952, **36**, (1), 98–100).—The hot-machining techniques reviewed are experimental developments only, but the results indicate that these methods have a number of advantages over cold machining and are particularly suitable for machining hard alloys that are either difficult or impossible to machine by conventional methods. The methods reviewed include turning, cutting, grinding, &c., and a comparison of various heating methods, such as flame, induction, and arc heating, are made.

—R. P. H. F.

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20 — CLEANING AND FINISHING

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

***Polishing of High-Purity Aluminium [and Its Alloys with 0·5, 1, and 2% Magnesium] by Repeated Anodizing and Re-solution of the Anodic Films.** Richard Lattey (*Metalloberfläche*, 1952, [A], **6**, (3), 41–48).—A detailed study has been made of the brightening of Raffinal (high-purity Al) and Reflectal (Raffinal + 0·5, 1, or 2% Mg) by alternate anodizing and stripping of the surface film obtained; the reflectivity of the finished surfaces was evaluated by comparison with that obtained from an evaporated Ag film on glass (98% total reflectivity). The following procedure was found to give the best results: (1) Anodizing in an electrolyte contg. $(COOH)_2$ 56, and CrO_3 0·5 g./l., at 20°–35° C., and using A.C. of 1·5 amp./dm.^{2} for 10 min. and D.C. of 3 amp./dm.^{2} for 5 min.; this produces a film ~9·5 μ thick. (2) Removal of the oxide film without damaging the polished metal surface at its base, by treatment in a mixture of 80% HNO_3 (53%) and 20% HF (71%), for 12–40 sec. (3) Three-fold repetition of this cycle, i.e. (1) and (2). (4) Preservation of the mirror surface by final anodizing for 10 min. in an electrolyte contg. H_2SO_4 21·5% (by vol.), at 18°–22° C., using D.C. of 1·5 amp./dm.^{2}; this provides a durable surface, consisting of an oxide film 4–5 μ thick, with little loss in reflectivity. After such treatment: (*a*) the total and the specular reflectivities of Al specimens increases from 50 to 87, and from 18 to 75%, resp., as the purity of the metal rises from 99·54 to 99·98%, and (*b*) the specular reflectivity of reflectors made from Raffinal is 71·4%, Reflectal (0·5% Mg) 76·4%, and Reflectal (2% Mg) 73·5%; the total reflectivity of each is ~85%.—E. N.

Grinding and Polishing of Aluminium and Aluminium Alloys, Especially as Pretreatments for Anodic Oxidation (Eloxal Treatment). Richard Erdmann (*Metalloberfläche*, 1952, [B], **4**, (3), 36–42).—E. discusses: (1) the properties of Al and Al alloys—in the cast and wrought forms—affecting machinability; (2) machining speeds and feeds, cutting compounds and lubricants; (3) mech. grinding, scurfing, and polishing procedures and materials; sand- and shot-blasting, scratch-brushing, and emery finishing; (4) barrel polishing; (5) chem. polishing; (6) chem. and electrochem. cleaning and degreasing; and (7) blemishes in the final anodized surfaces, resulting from faults in the above-mentioned operations.—E. N.

New Drawing Brass Cuts Finishing Costs. S. G. Kelley (*Materials and Methods*, 1952, **35**, (3), 82–83).—" Formbrite ", the so-called new brass, is a standard brass manufactured by a new technique, with an exceedingly fine grain-size of 0·012 mm. or less. The brass is said to have excellent polishing and finishing characteristics, high tensile strength, and increased hardness combined with good ductility. The material can be produced with a fine grain-size only in the thinner gauges.—R. P. H. F.

Cleaning, Dow Treating, and Finishing Magnesium Products. James D. Breneman (*Indust. Finishing* (*U.S.A.*), 1952, 28, (11), 20–22, 24, 26, 28, 30, 32, 34).—Surface treatment of Mg alloy parts is described.—K. C. R.

Chemical Cleaning of Heat-Exchange Equipment. C. M. Loucks and C. H. Groom (*Trans. Amer. Soc. Mech. Eng.*, 1949, **71**, (7), 831–836).—The use of chem. solvents in cleaning heat-exchange equipment is described on the basis of the types of deposits, and the metals encountered. The appn. of the solvents is discussed, and a number of case histories which include the turbine condensers of power plants and the wash oil coolers in steel-mill coke plants, are described, illustrating chem. cleaning in the various industries.—H. Pl.

Modern Electrolytic Degreasing. A. Pollack (*Metalloberfläche*, 1952, [A], **6**, (2), 27–28).—Recent developments in the compn. and operation of electrolytic degreasing baths are briefly reviewed, and the advantages of the process are discussed.—E. N.

Wet Blasting Performs Many Cleaning and Finishing Operations. John L. Everhart (*Materials and Methods*, 1952, **35**, (4), 98–100).—Blasting with an abrasive such as silica suspended in a liquid is described. This wet blasting method provides a flexible means of surface prepn. before plating and coating, the removal of grinding and polishing marks before finishing, removal of oxide scale resulting from annealing and heat-treatment, and deburring. It is also applicable to the cleaning of parts before welding and to the removal of flux after welding.—R. P. H. F.

Band-Polishing Processes. Otto Schleppi (*Metalloberfläche*, 1952, [A], **6**, (2), 23–26).—Various types of machines, both fixed and portable, for the band-polishing of plain and profiled articles are described and illustrated.—E. N.

Abrasive-Belt Finishing. W. A. Papworth (*Machinery* (*Lond.*), 1950, **77**, (1967), 12–18).—The types of machines—platen, contact roll, and free-belt grinding—and the abrasive belts for various materials, are described.—S. R. W.

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Technical Survey. Published semi-annually by the Copper Development Association, Kendals Hall, Radlett, Herts. No. 1 appeared in January 1953. (Gratis).

25 — BOOK REVIEWS

An Outline of Metallurgical Practice. By Carle R. Hayward. Third edition. 9 × 6 in. Pp. viii + 728, with 445 illustrations. 1952. New York: D. Van Nostrand Co. ($10.00); London: Macmillan and Co., Ltd. (75*s.*)

The second edition of Hayward's book (1940) will presumably be familiar to many of those who are likely to be interested in the present edition. A substantial proportion of the text of the former has been retained without alteration, but the third edition contains much-enlarged chapters on titanium and zirconium, and a chapter on uranium is included. In addition, other parts of the book have been altered somewhat. For instance, the chapter on copper includes fuller discussion of dust collection, references to the use of oxygen in smelting, and notes on mechanical twyer-punching and secondary copper recovery. The chapter on lead includes "slag-fuming", that on aluminium contains more information on methods of alumina production, and controlled cooling of nickel–copper mattes, magnesium recovery from sea-water, and the Pidgeon process are referred to.

The work is concerned mainly, though not exclusively, with American practice. In recommending supplementary reading, particularly, Professor Hayward displays an inexplicable insularity. Few people would, one supposes, claim that the contributions to lead metallurgy made by Australian metallurgists are negligible, that Belgian, German, and British journals contain no information on zinc smelting, or that German metallurgists have produced no worth-while literature on secondary-metal recovery. Can it be, then, that such publications are unobtainable in the United States of America?

G. L. Evans.

Photoconductivity in the Elements. By Trevor Simpson Moss. 8¾ × 5¾ in. Pp. x + 263, with 56 illustrations. 1952. London: Butterworths Scientific Publications. (50*s.*)

Those who partake intimately in the rapid expansion of a restricted branch of physics must keep abreast of developments by constant reference to new papers appearing in the scientific journals. The early assessment of the implications of new developments in one field for other fields is an important concern to those working in the latter. To obtain this assessment by a study of the detailed and highly specialized original papers is at the least a lengthy and arduous task. Here, general survey papers or books play an important part, particularly if the surveys are written in a style which is "readable" by specialists in any of the other physical sciences.

The book under review immediately falls into this category. It is eminently a readable book. The subject of photoconductivity is treated *ab initio.* It can be read with little or no previous knowledge of semi-conductor theory, and the first chapters give an excellent introduction to the subject for the non-specialist. A simple development from the energy levels of a single atom to energy band spectra and forbidden zones is given in non-mathematical language.

A description of the difference between metals and insulators and of the conception of semi-conductors is followed by an explanation of the different types of semi-conductors and the mechanisms of excitation and electrical conduction. This first chapter takes the non-specialist a great step forward. He can now appreciate much of the vocabulary of his semiconductor specialist colleague. "*P*-type", "*n*-type", "donor and acceptor centres" take on a reality perhaps not before appreciated. The complete absence of any attempt to include mathematical theory in the introductory chapter is almost startling in this subject so often clothed in wave-mechanical arguments.

The next two chapters introduce the reader to the Hall effect and conductivity processes, and to optical absorption effects in semi-conductors with only a modicum of classical mathematics, still by way of introduction to the main subject. Photoconductivity is next expounded from the simple single excitation process. The place of activation energies and dependence on the radiation wave-length is followed by more detailed accounts of the different phenomena associated with the photoconductive process.

Part two of the book deals with the experimental methods and measurements. Separate chapters are devoted to boron, in thin films and bulk, diamond, silicon, germanium, grey tin, phosphorus, arsenic, antimony, sulphur, selenium, tellurium, and iodine. These chapters, which constitute the rest of the book, except for a final discussion, give details of methods of preparation, and summarize measurements made on the Hall effect, photoconductivity, the effects of impurities, electron and hole mobilities, activation energies and temperature dependences, &c.

The final discussion summarizes the measured properties and attempts to relate these with some of the known properties of the different elements.

There is little discussion on the possible application of the photoconductors described, but the information given is quite adequate for those who may wish to consider these applications. It is too much to expect the data given to be completely up to the minute in these days of ever-accelerating progress, but the recent dates quoted in the quite extensive bibliography assure us that reference has been made to very recent literature.

M. E. Haine.

Nichtmetallische anorganische Überzüge. Von Willi Machu. 25 × 17·5 cm. Pp. xii + 404, with 152 illustrations. 1952. Wien I: Springer-Verlag. (97*s.*)

The production of inorganic coatings on metal surfaces, whether for their intrinsic resistance to corrosion, as a means of increasing the adhesion of paints, varnishes, and lacquers to the parent metal, for purely decorative purposes, or for assisting cold working, is now a most important part of the metal-finishing industry. Professor Machu's companion volume to his book on metallic coatings is therefore of considerable interest, as he has brought together within some 400 pages a great deal of information on these coatings and coating processes, ranging from the well-known anodizing and phosphating treatments to the use of sulphur and selenium coatings, or the old Bower–Barff process for the treatment of iron and steel.

With such a wide field to cover, it is gratifying to note that particular attention has been given to the arrangement and method of presentation of the subject matter. The book has been divided into four main sections, dealing with: (I) The electrochemical and chemical oxidation of the light metals (140 pp.); (II) Metal colouring (55 pp.); (III) Phosphate coatings (75 pp.); (IV) Vitreous enamel coatings (120 pp.).

The material in each section is grouped under major subject headings of which there are 78; references to the relevant patent and scientific literature are given at the end of each of these sub-divisions and total 1800. This classification system, however, collapses completely at the end of the book, as topics considered in the last 8 sub-sections under the main heading of "vitreous enamel coatings" include graphite or carbide coatings, protective coatings on zinc, tin, and copper, and increased corrosion resistance by nitriding. Indeed, it is most doubtful whether nitriding comes within the scope of the book at all and as only 7 references are given, the most recent of these being 10 years old, it looks as if this subject was included to bring the total number of references up to a round 1800.

The book should prove useful to metal finishers, chemists, metallurgists, and engineers who wish to obtain a broad general picture of this field, whilst at the same time the bibliography provides a good foundation for more detailed reading. As one would expect, however, there is a strong bias towards German practice and processes, in preference to treatments which are quite well known in this country or the United States. For example, in discussing chemical surface treatments for aluminium, 4 pages are devoted to the M.B.V. process, whereas the Alrok process is dismissed in 4 lines, and the Alocrom and Pylumin processes are not even mentioned. Similarly, the Cronak process for zinc surfaces receives only 12 lines' consideration.

The section dealing with "Metal colouring" is particularly interesting, as it contains a good account of the production of protective decorative oxide coatings on iron and steel surfaces by controlled heat-treatment, anodic oxidation, and molten salt or alkali solution treatment, as well as recommended methods for colouring copper, zinc, tin, aluminium, &c. On the other hand, the treatment of phosphate coatings is frankly disappointing, being, as far as one can see, condensed directly from the author's book "Die Phosphatierung: Wissenschaftliche Grundlagen und Technik" (1950). It is especially surprising that no details are given of the use of phosphate coatings for assisting wire drawing or the cold extrusion of steel, both of which were German developments. Similarly, the phosphating of aluminium surfaces receives no mention.

The book is well produced and clearly printed, but its value to the practical man would have been greatly enhanced if more indications could have been given as to which of the many patents quoted have any real commercial or technical significance.

H. A. HOLDEN.

Visual Lines for Spectroscopic Analysis. By D. M. Smith. Second, revised edition. 8½ × 5½ in. Pp. 102, illustrated. 1952. London: Hilger and Watts, Ltd. (16*s.*)

In his preface to this second edition, the author expresses the opinion, with which many will agree, that "photographed spectra are more generally useful than visual observations in a restricted wavelength range". Nevertheless, occasions arise in all laboratories having some familiarity with spectra, when the presence or absence of a specified element must be settled immediately by reference to its visual spectrum. The first Edition of "Visual Lines" was both useful and reliable in such circumstances. It has now been amplified, revised, and brought up to date; it includes 19 more elements, and an improved standard of presentation and printing has been adopted. One may disagree with the author on certain minor points, such as the inclusion of doubtful intensity values from the M.I.T. tables, which might sometimes be misleading, but the book can be recommended, nevertheless, as an indispensable adjunct to spectroscopic identification of elements.

M. MILBOURN.

Gestaltung von Druckgussteilen. (Spritz- und Pressguss). Von Gustav Lieby. 23·5 × 16·5 cm. Pp. 157, with 135 illustrations. 1949. Stuttgart-O: Franckh'sche Verlagshandlung. (Geb., DM 15.—.)

Authoritative books on foundry subjects are scarce, and this gives a particular significance to the appearance of Lieby's book, for it fills a great need and makes at the same time a very good contribution to the foundry literature.

It is very seldom that a book on a technological subject could please at the same time a student as well as a specialist in the field. For the former, the writing should be lucid and simple; for the latter, brief, factual, analytical, and at a high level. Lieby manages to combine the two requirements skilfully, as well as to treat the metallurgy and engineering of pressure die-casting without injustice to either but with a proper emphasis on both.

The book deals preponderantly with the design of pressure die-castings. While this is the central theme, which both the manufacturer and the user, the metallurgist and engineer are likely to find useful and interesting, a number of other topics are also properly related to this central theme: casting methods and alloys, fettling machinery, tolerances, and economics. The book thus attains a good balance, and it is a great pity that the subjects of die-casting machines and die manufacture were not included. With these and one or two other minor topics, the book could have been a text-book as well as a handbook on pressure die-casting.

V. KONDIC.

Industrial High Vacuum. By J. R. Davy. 8½ × 5½ in. Pp. viii + 243, with 82 illustrations. 1951. London: Sir Isaac Pitman and Sons, Ltd. (25*s.*)

In his introduction to this book, the author remarks that he wrote it with the particular object of making contact with the industrial technician and the student. On reading this, the reviewer was stimulated to refresh his memory as to the succinct definition of a "technician" given in the recent admirable report on "The Education and Training of Metallurgists", published by the Joint Committee of our various Institutes and Institutions. Technicians, it says, are "persons employed in the application of existing techniques. Usually they have little knowledge of the basic sciences, but are skilled in the expert conduct of some branch of production work." In as much as the present book is essentially practical in character, and makes do with a minimum of simple theory, it fulfils the author's avowed purpose admirably. It is very readable and is profuse in experimental tips which will be found useful by those unversed in the difficulties of high-vacuum work.

The book covers the usual subjects of pumps, gauges, seals, valves, vacuum mechanisms, leaks, and leak-finding rather briefly. The author is on his own ground, however, in discussing evaporation techniques, and the chapters dealing with the deposition of metallic, anti-reflecting, and other types of films are the best in the book. It should be made clear, in fact, that the book is written almost entirely from the point of view of the "bell-jar" man; the title is a little misleading, since large-scale industrial plant is only touched upon in a very perfunctory manner.

There is a chapter headed "Vacuum Metallurgy" which perhaps calls for notice in a review for the *Metallurgical Abstracts*. It is short (13 pages), does not contain any glaring inaccuracies, and will probably prove of some interest to the non-metallurgist.

C. E. RANSLEY.

METALLURGICAL ABSTRACTS

METALLURGICAL ABSTRACTS AUTHOR INDEX

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Members are reminded that they can subscribe through the Institute to the new international journal *Acta Metallurgica* at the privileged rate of £3 6*s*. 0*d*. per annum. A leaflet describing *Acta Metallurgica* and a subscription form were distributed with the February issue of the *Journal*.

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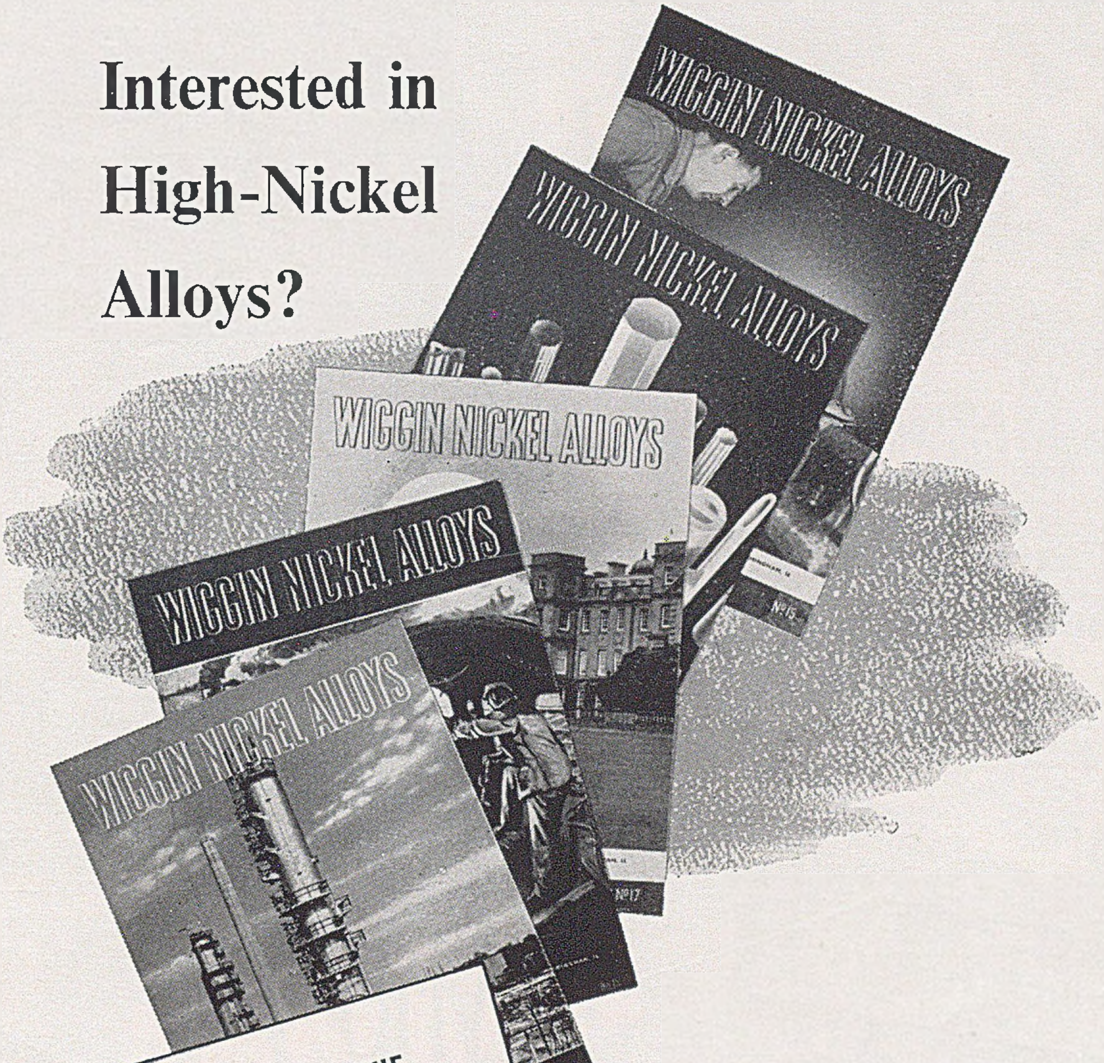
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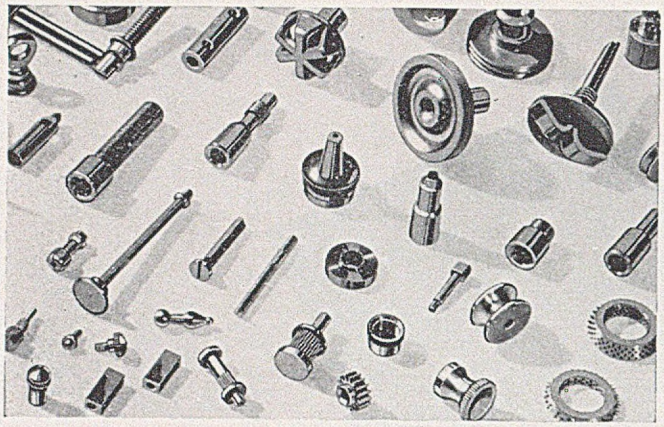

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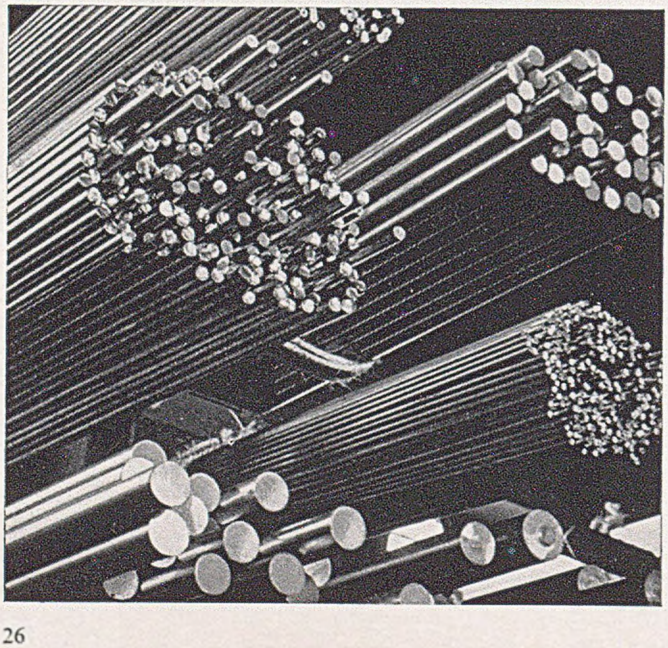

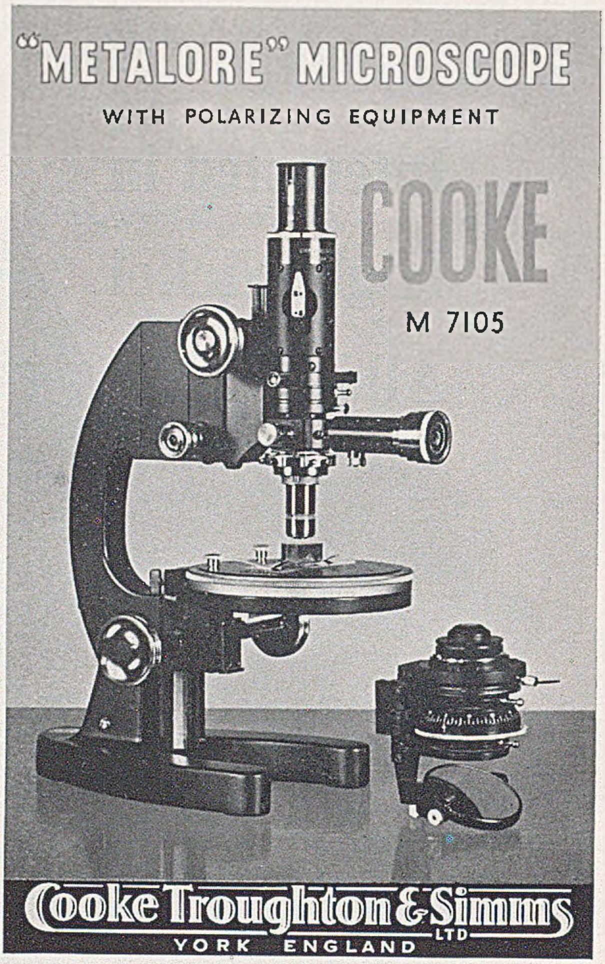

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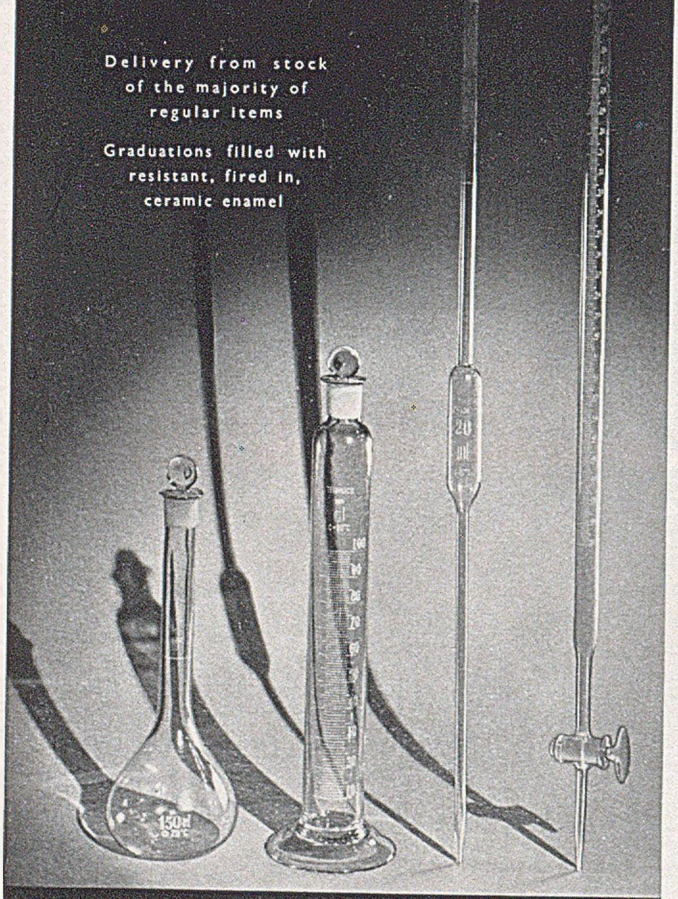

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CONTENTS: Electric Furnaces for the Thermal Treatment of Non-Ferrous Metals and Alloys; Gas Equipment for the Thermal Treatment of Non-Ferrous Metals and Alloys; Batch and Continuous Annealing of Copper and Copper Alloys; Bright Annealing of Nickel and its Alloys; Batch Thermal Treatment of Light Alloys; Flash Annealing of Light Alloys; Continuous Heat-Treatment of Aluminium Alloys of the Duralumin Type.

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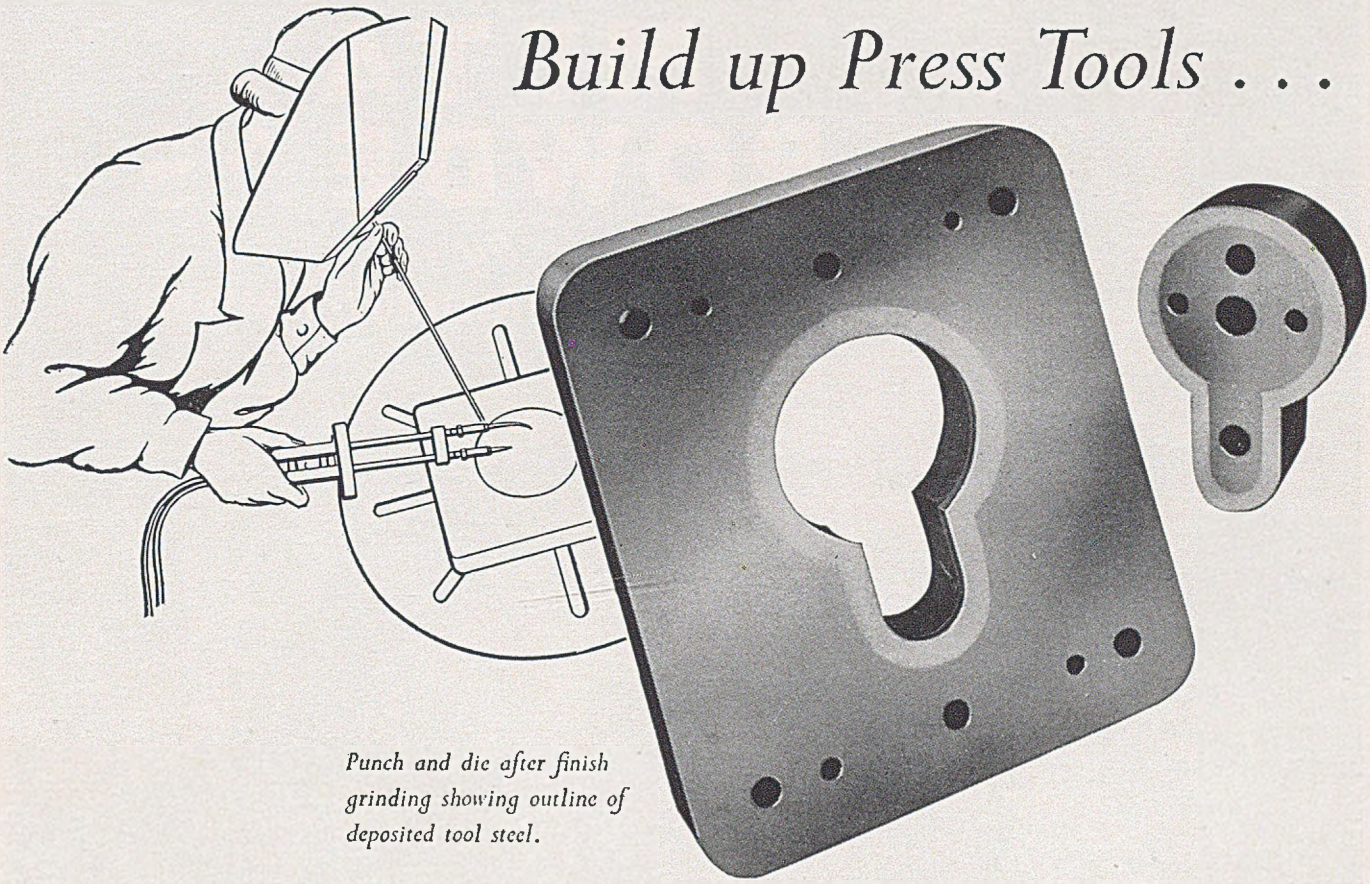

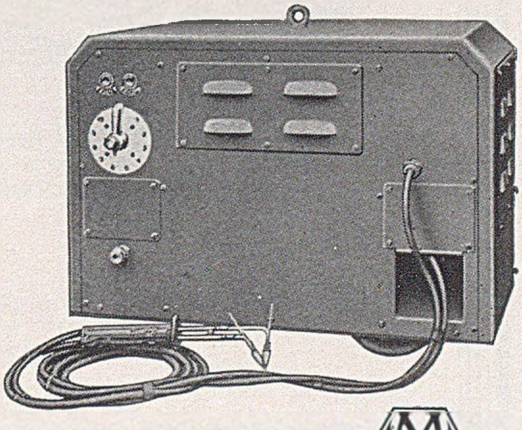

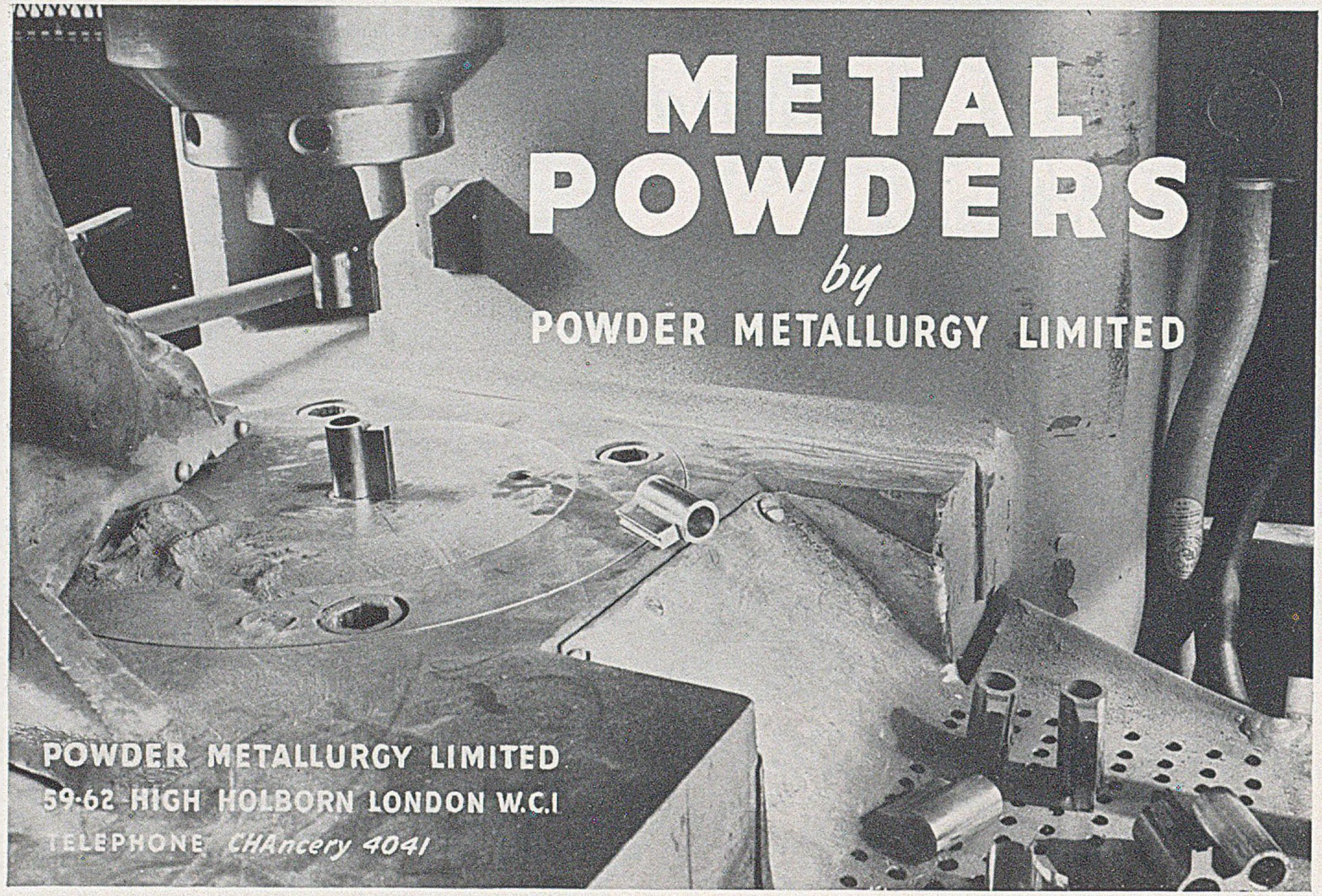

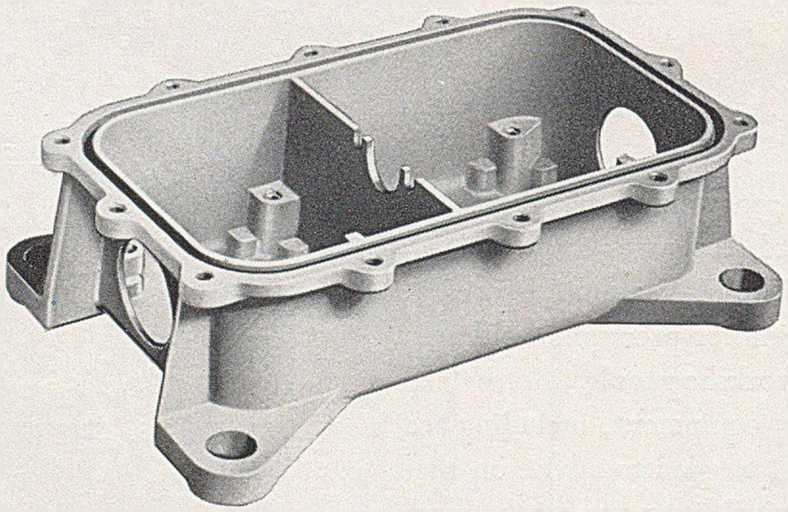

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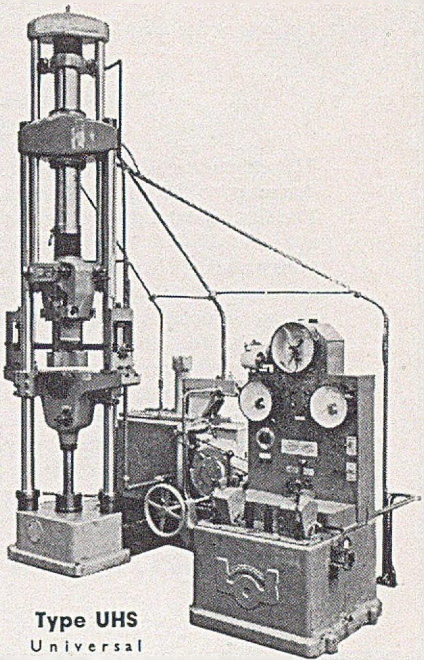
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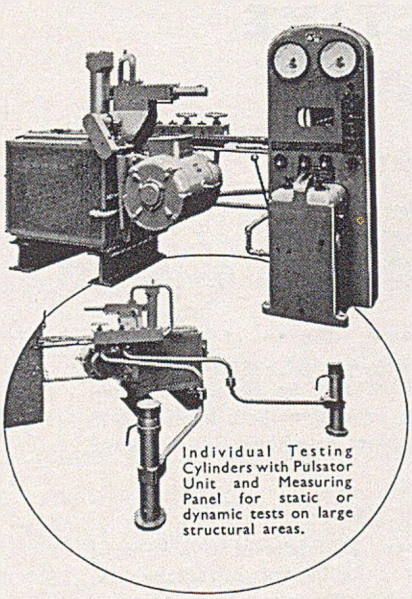
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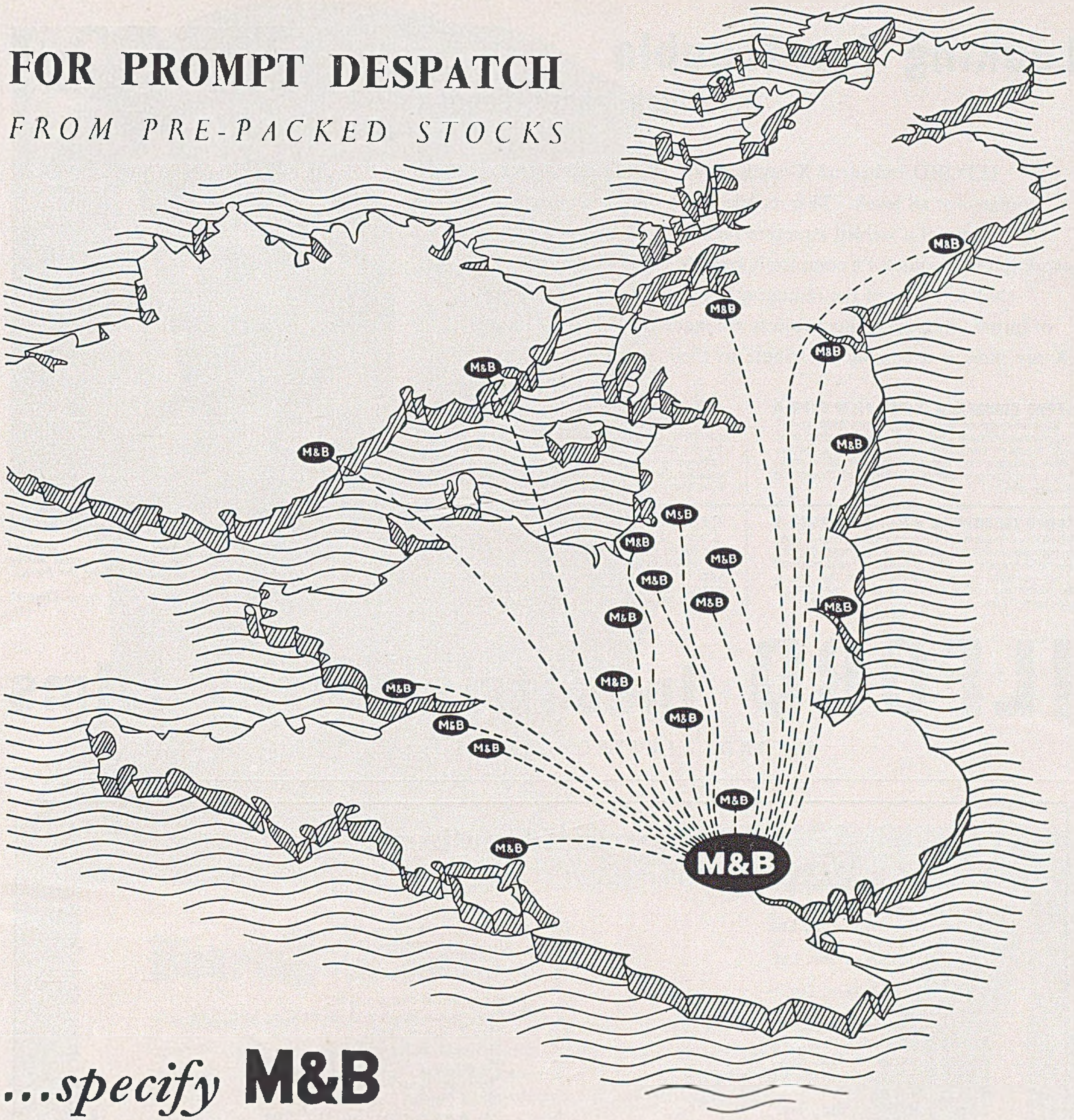
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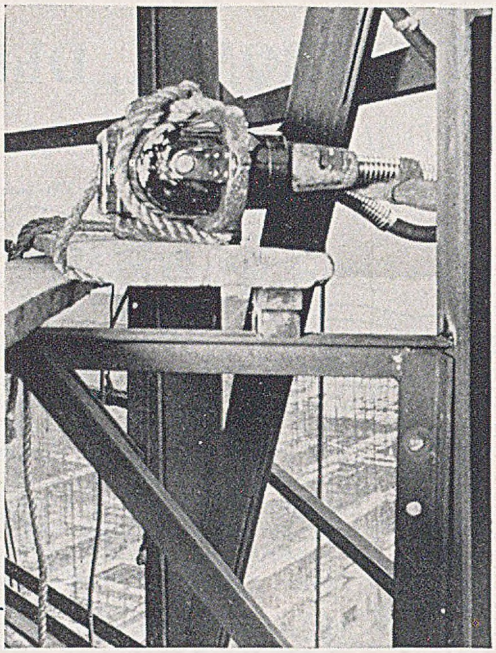

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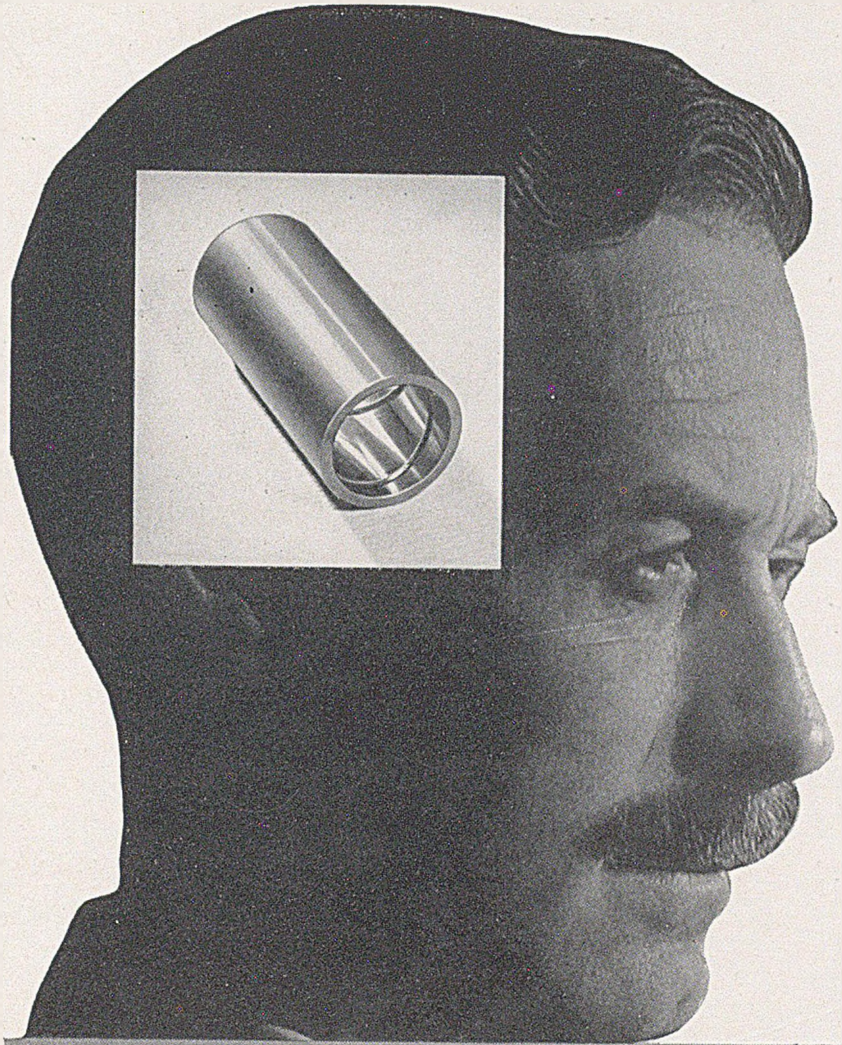

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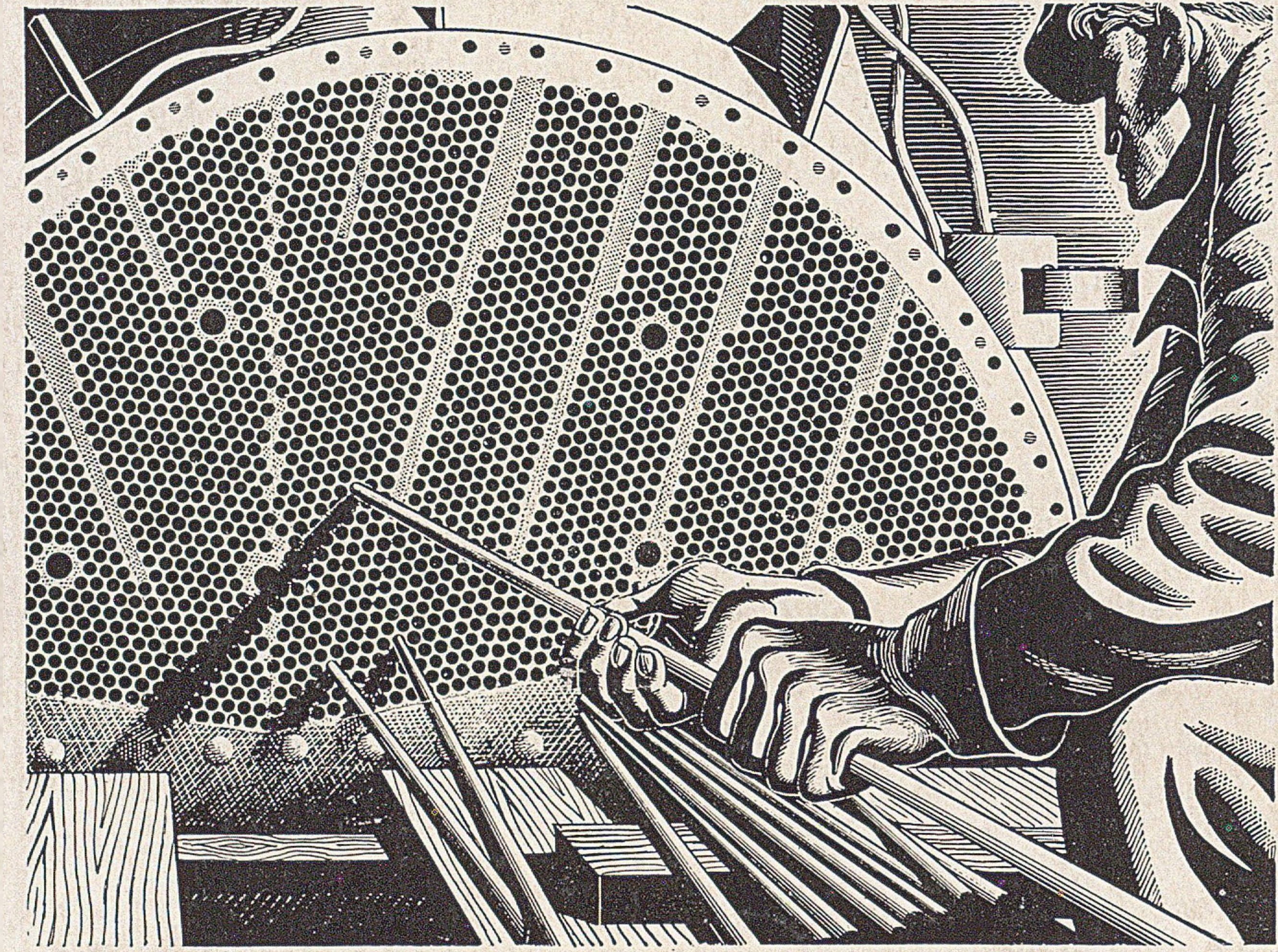

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