THE JOURNAL OF THE INSTITUTE OF Metals

WITH THE BULLETIN AND METALLURGICAL ABSTRACTS

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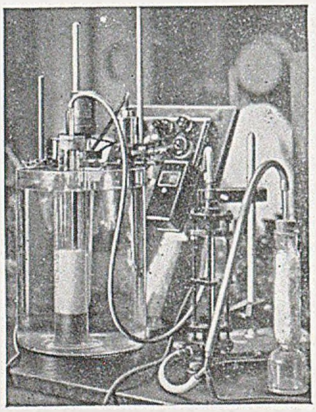
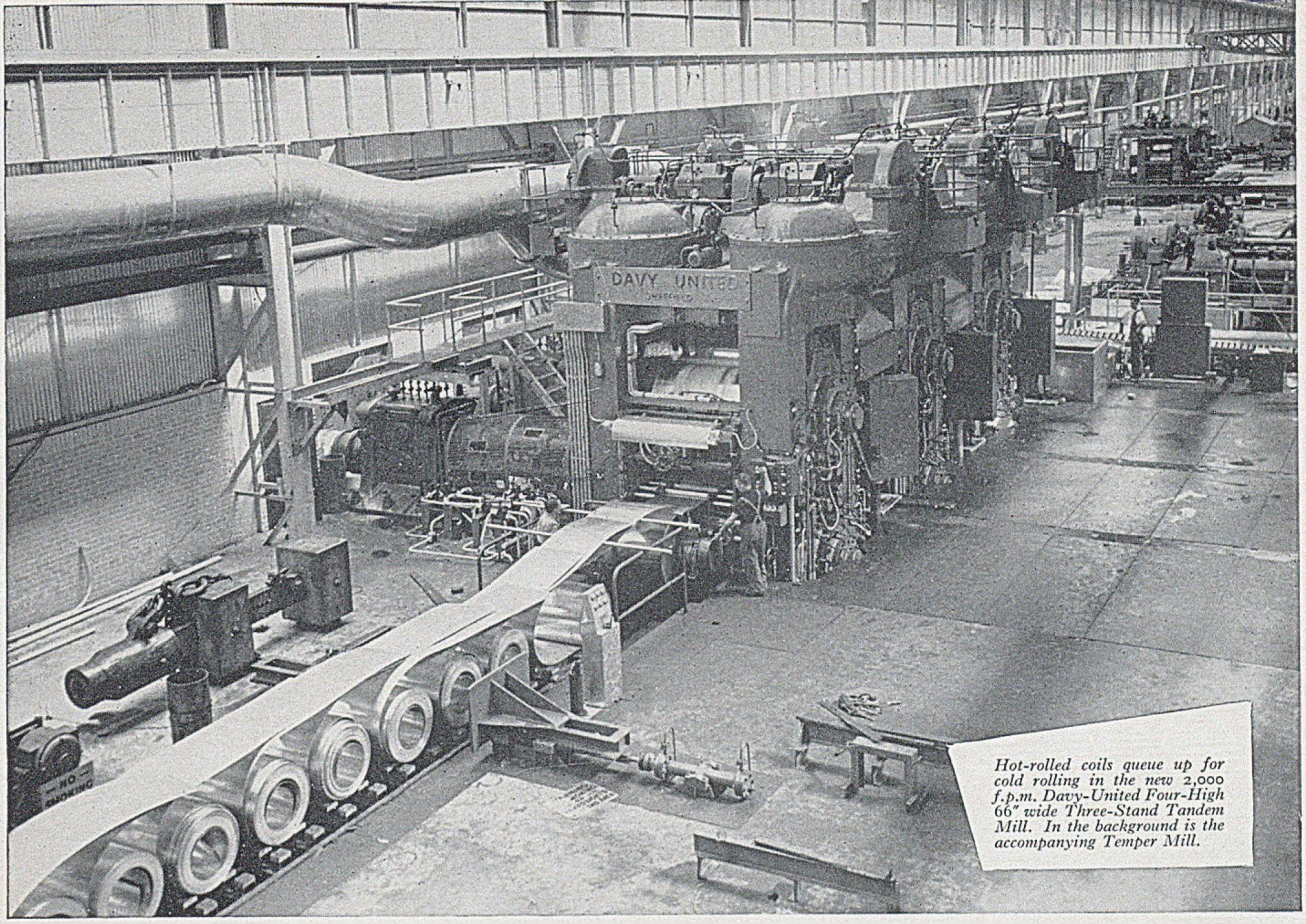
SEPTEMBER 1951

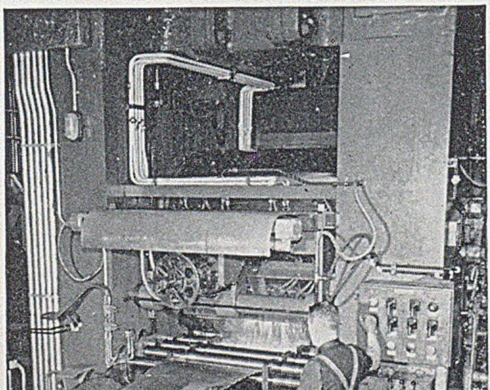

What has this to do with hydraulics?





Hot-rolled coils queue up for cold rolling in the new 2,000 f.p.m. Davy-United Four-High 66″ wide Three-Stand Tandem Mill. In the background is the accompanying Temper Mill.

An Integral part of a Great Production Project



Close-up of the ingoing end of the tandem mill, with a coil well advanced on its single run through the mill. Ingoing strip thickness ranges from .08″ to .02″ and outgoing thickness between .01″ and .064″, in finished widths up to 56″.

The new continuous strip mill plant at the Rogerstone Works of Northern Aluminium Company Limited challenges comparison with any in the world. The 50,000 tons of aluminium sheet and strip which it will produce in one year will increase Britain's present capacity for that material by more than one-third.

Key units in the cold rolling section of this 7½-acre plant are two new Davy-United Mills—a 66″ wide 4-High 3-Stand Tandem Mill, and a following 36″ wide 4-High Temper Mill, both of which roll at the unprecedented speed of 2,000 f.p.m., the fastest production rate on aluminium in the world. All previously hot-rolled material in this plant, whatever its ultimate application, passes through this tandem mill which can handle coils up to 5,000 lb. in weight, larger than any aluminium coils previously produced in this country. A single run of a coil through the mill will effect reductions of from 60% to 90% in strip thickness and provide a length of strip of up to 3,500 feet, in finished widths up to 56″. Feeding, control and discharge equipment are believed to be the most complete ever installed.

This is yet another instance in which Davy-United engineering has helped to reinforce the industrial power of Britain.

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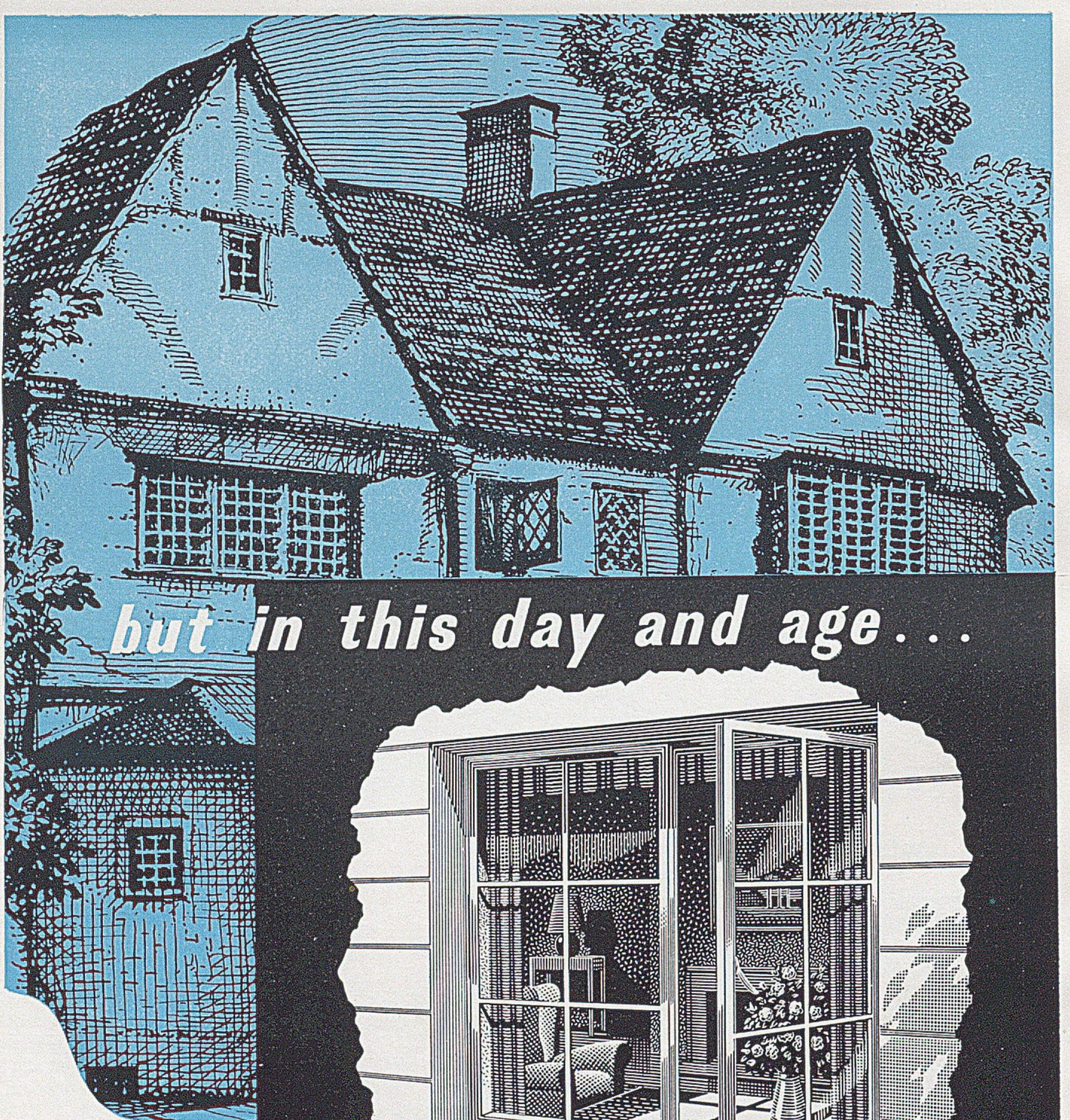
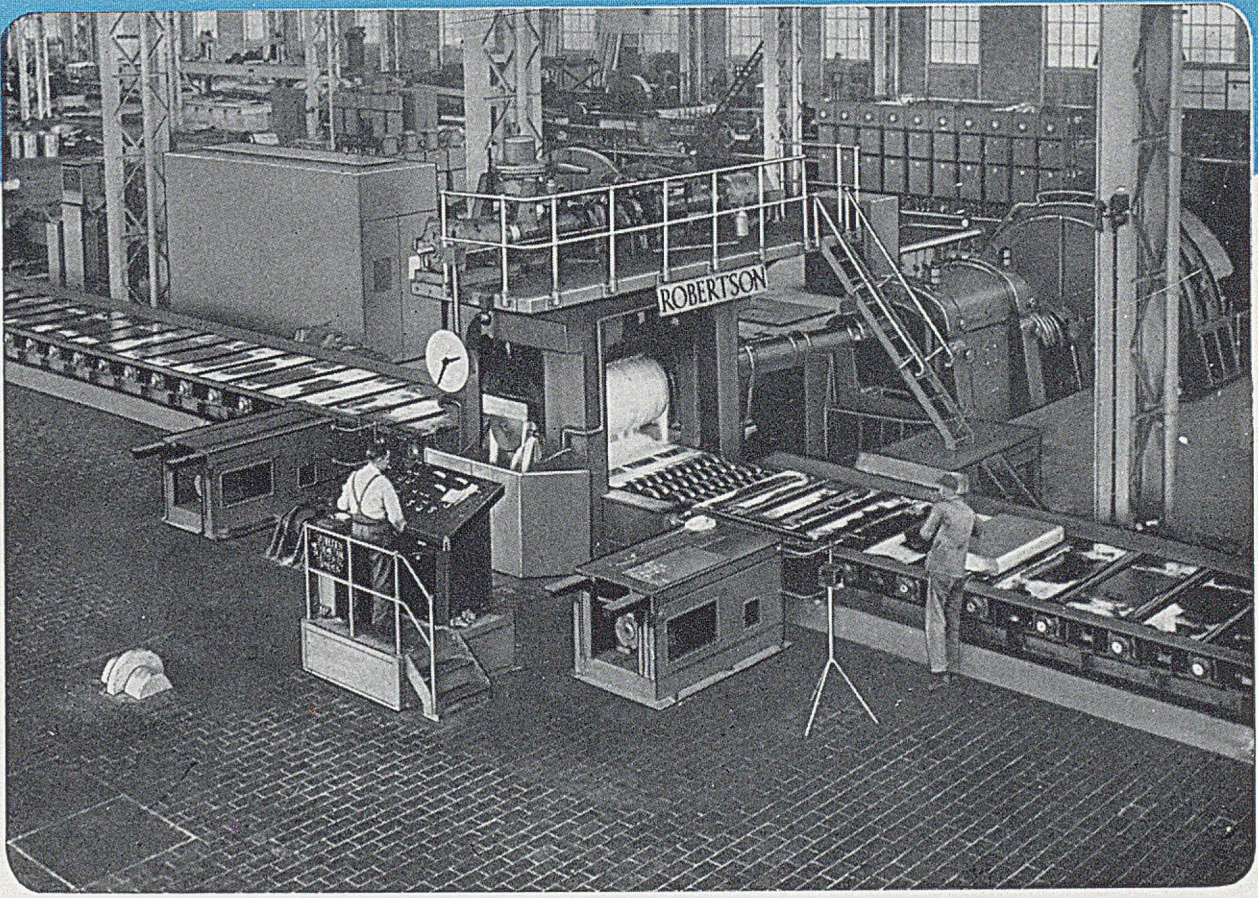

but in this day and age . . .

ROBERTSON

HOT ROLLING MILLS



Photograph by courtesy of Société Industrielle de l'Aluminium, Duffel, Belgium

Illustration shows a 32″ two-high reversing hot breaking down mill for aluminium and light alloy slabs, which incorporates the latest developments

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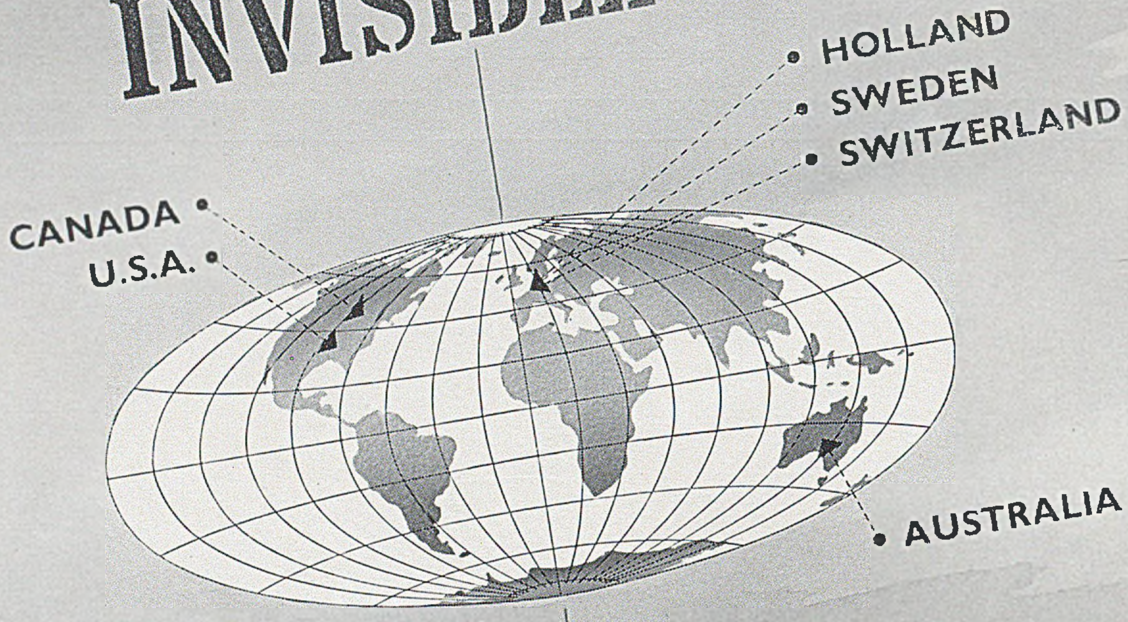
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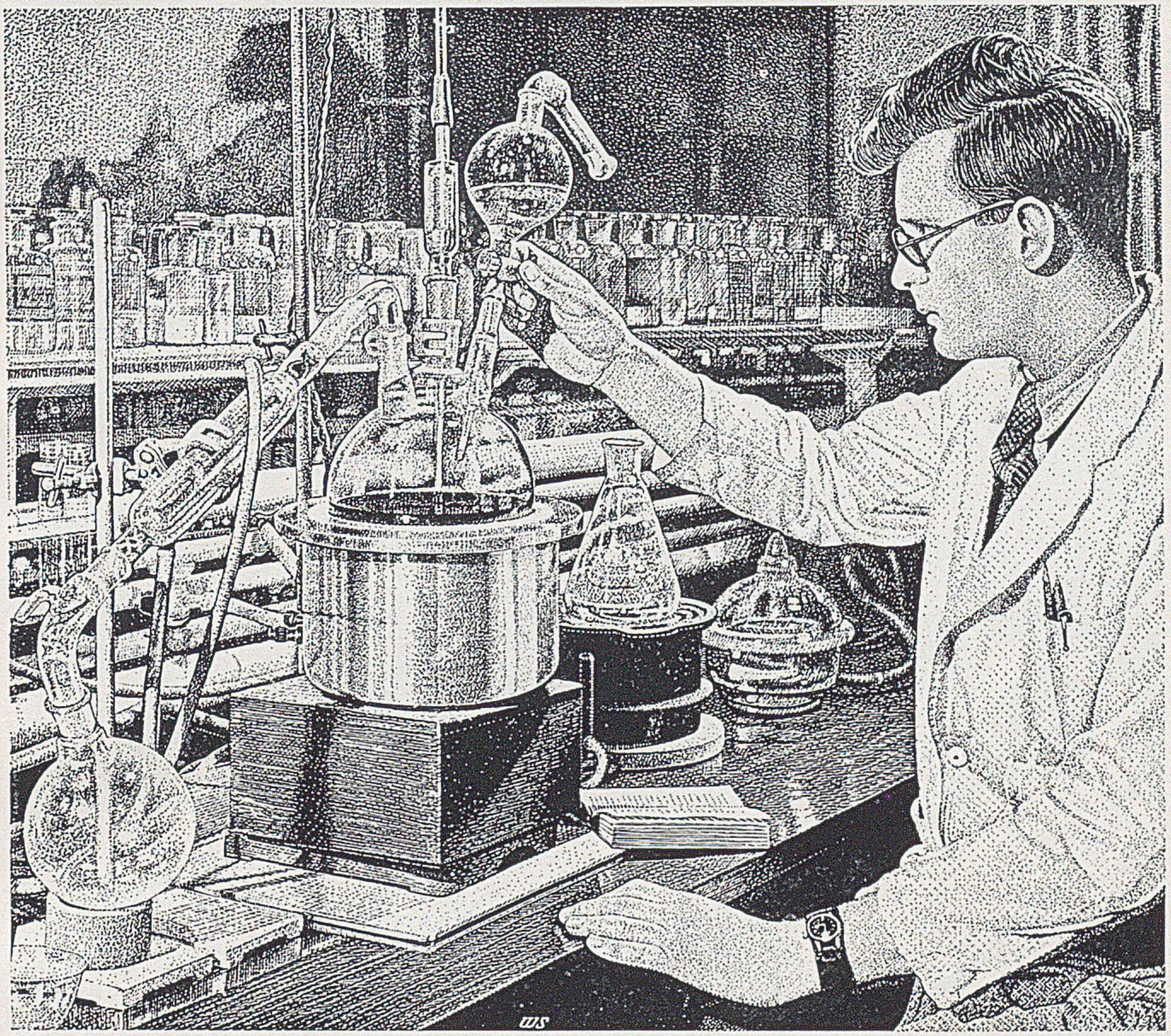
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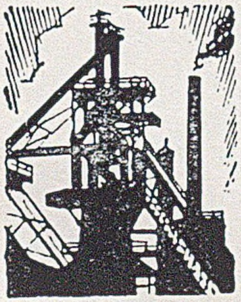


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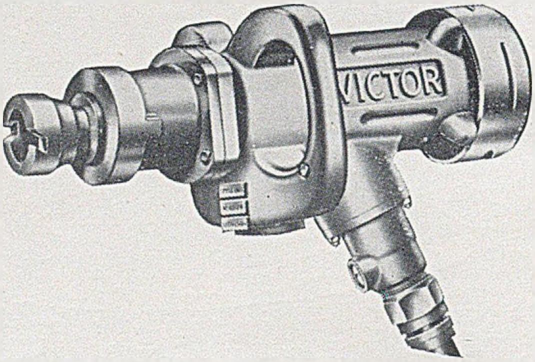

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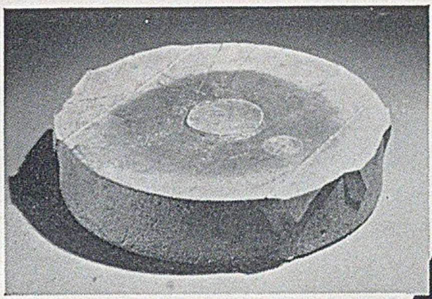

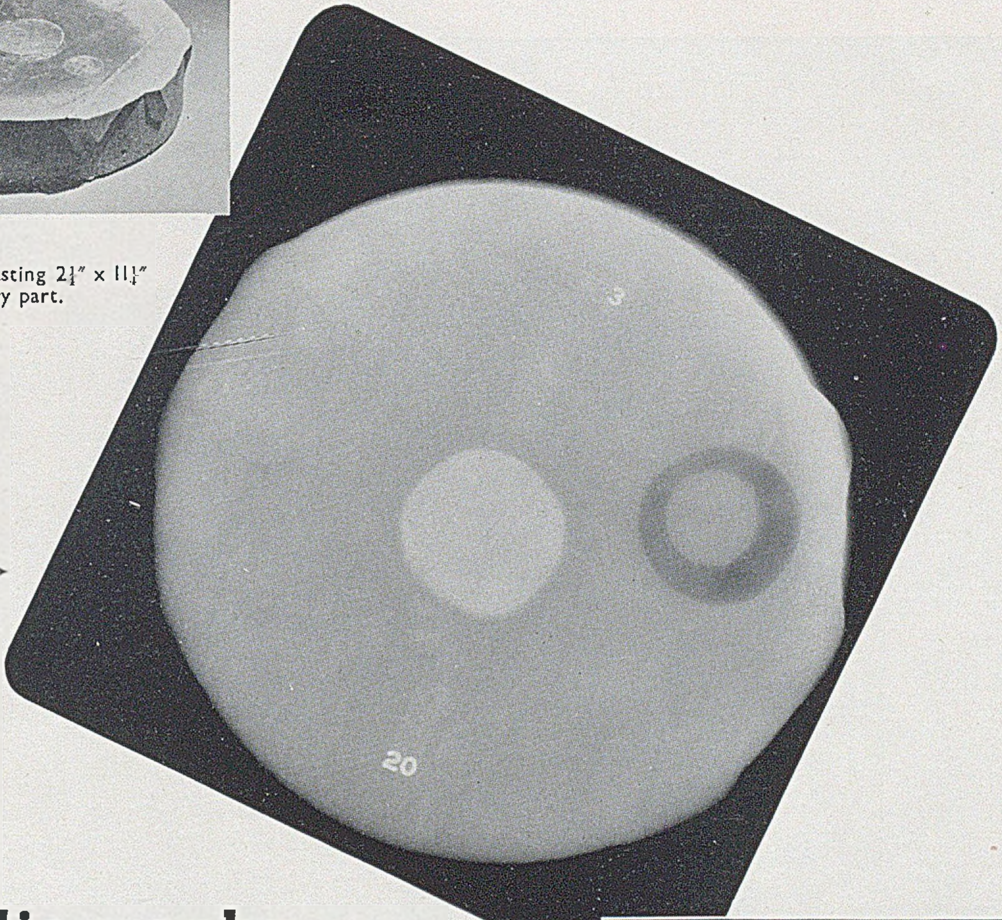
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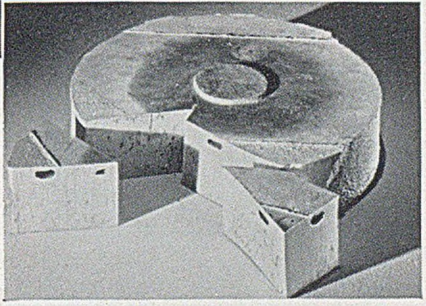

RADIOLOGICAL EXAMINATIONS

The fluorescent method of crack detection, involving the impregnation of the castings with a fluorescent material and their subsequent examination in ultra-violet light, constitutes a useful supplement to X-ray inspection. The application of this method to a light alloy casting is shown above.

and "X-ray quality" castings

THE RADIOLOGICAL EXAMINATION of castings, originally developed to meet the requirements of the aircraft industry for highly-stressed light alloy components, has long since been established on a routine basis. Our own very completely equipped radiological department has been in operation for more than ten years.

Other industries have now come to recognise the value of this penetrating supplementary inspection service and, by specifying "X-ray quality", have been able with advantage to extend their use of non-ferrous castings and reap the full advantage of light alloys in their designs. . . . Substantial savings can also be made by the radiological examination of castings calling for a considerable amount of machining: hidden faults are discoverable ***before*** the casting reaches the machine shop—not on the last operation.

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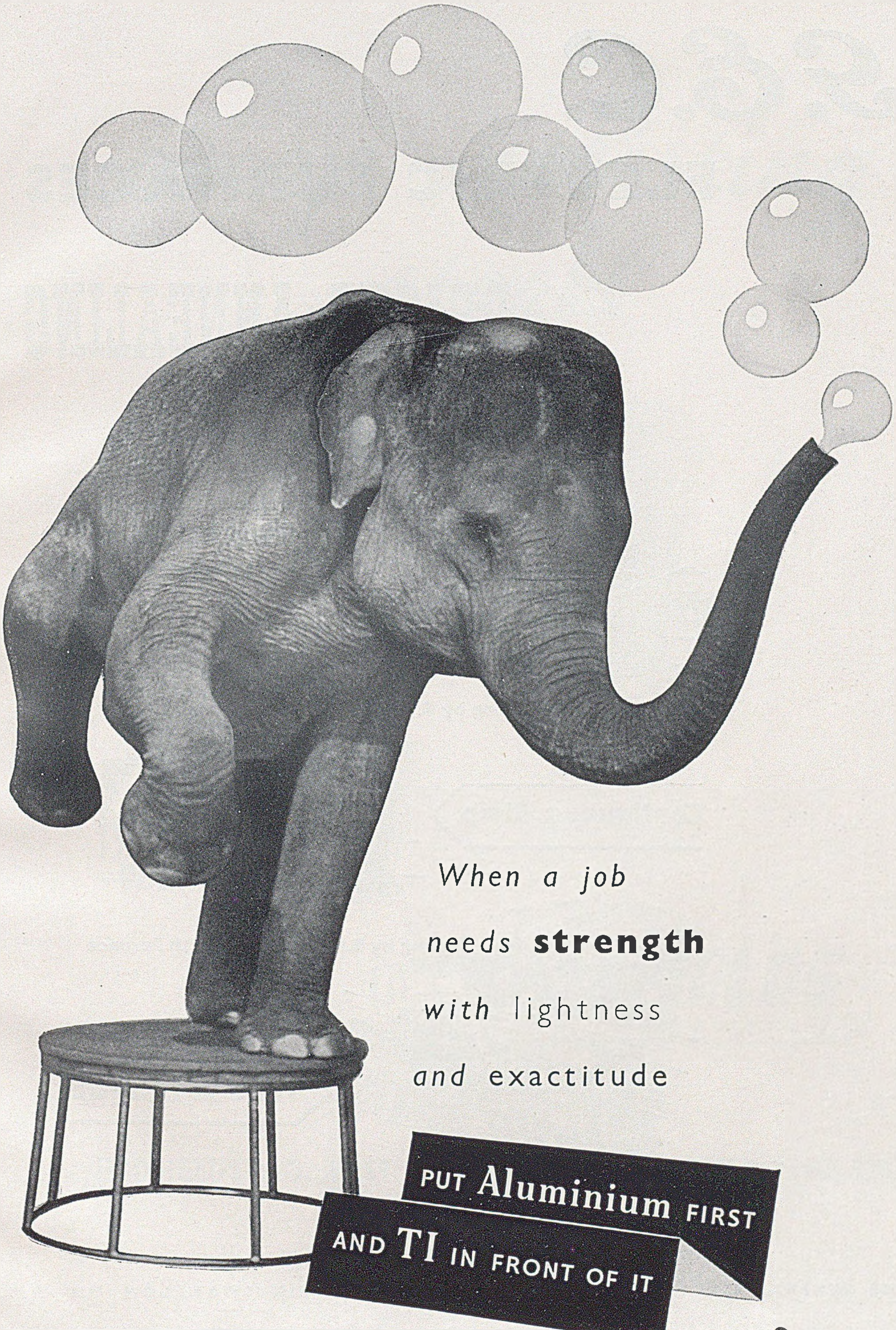

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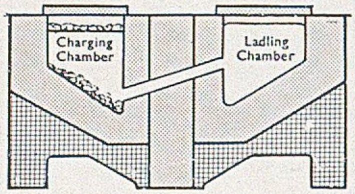

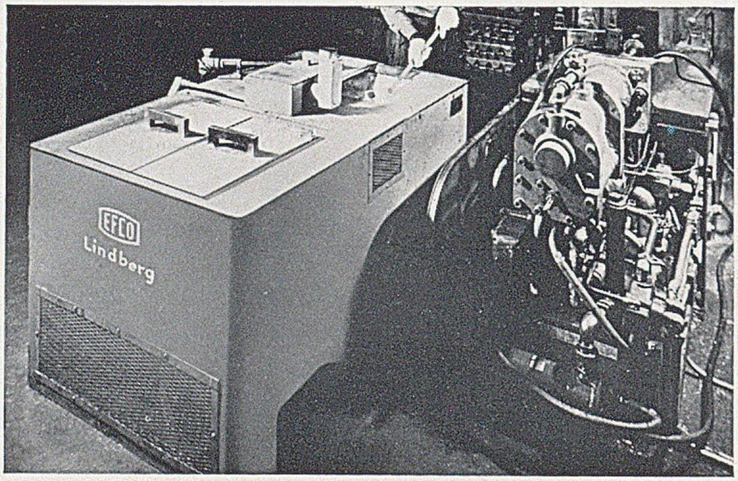

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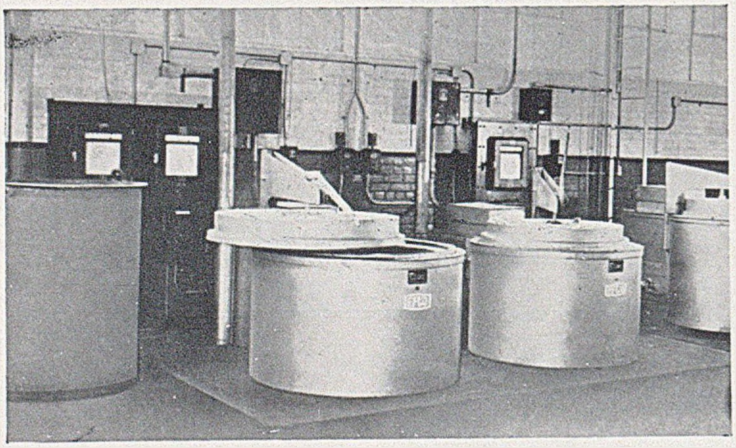

The vertical type cyclone is ideal for batch heat treating aluminium or magnesium castings and forgings. The charge is handled in and out of the furnace in baskets or fixtures. If the work is to be quenched it remains on the fixture and is quickly transferred from the furnace to the quench tank in a matter of seconds.

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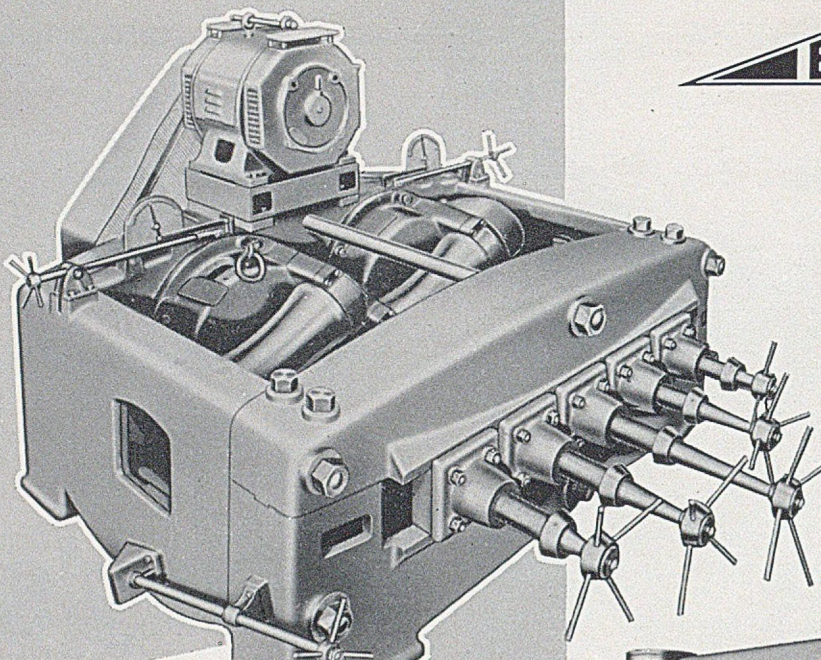
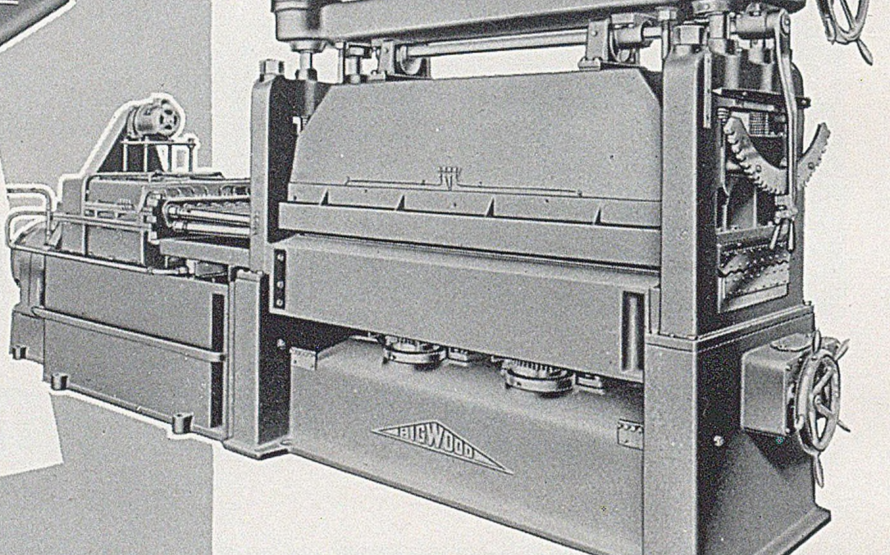
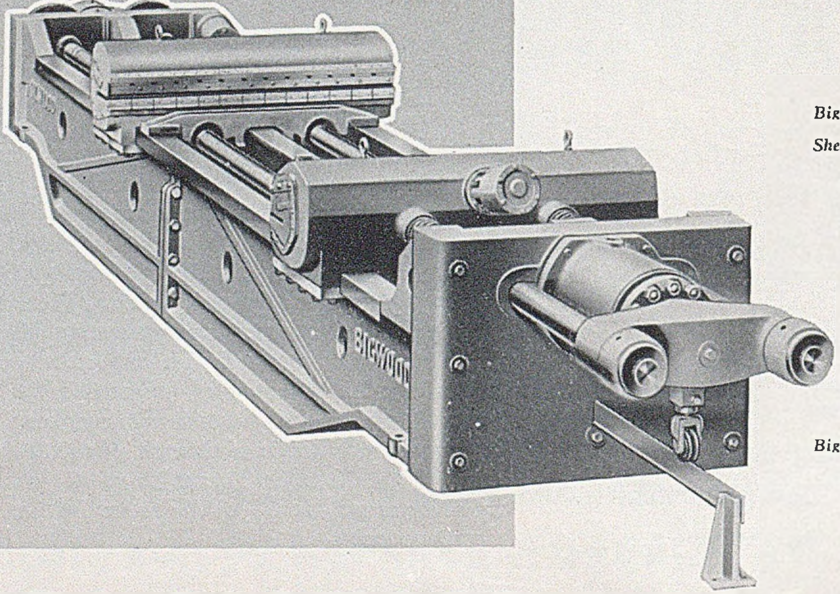


Bigwood Tube or Bar Straighteners are in use in light alloy work in many countries. (Illustrated is Type B—with 5 idle rolls.)

The long experience acquired by Bigwood's in the production of machinery for non-ferrous and light alloy work is at your disposal.



Bigwood 17-roll or 23-roll Precision Roller Sheet Leveller used with success on light alloys.



Bigwood 200-ton Sheet Stretcher with patented self-compensating grips for wide sheets.

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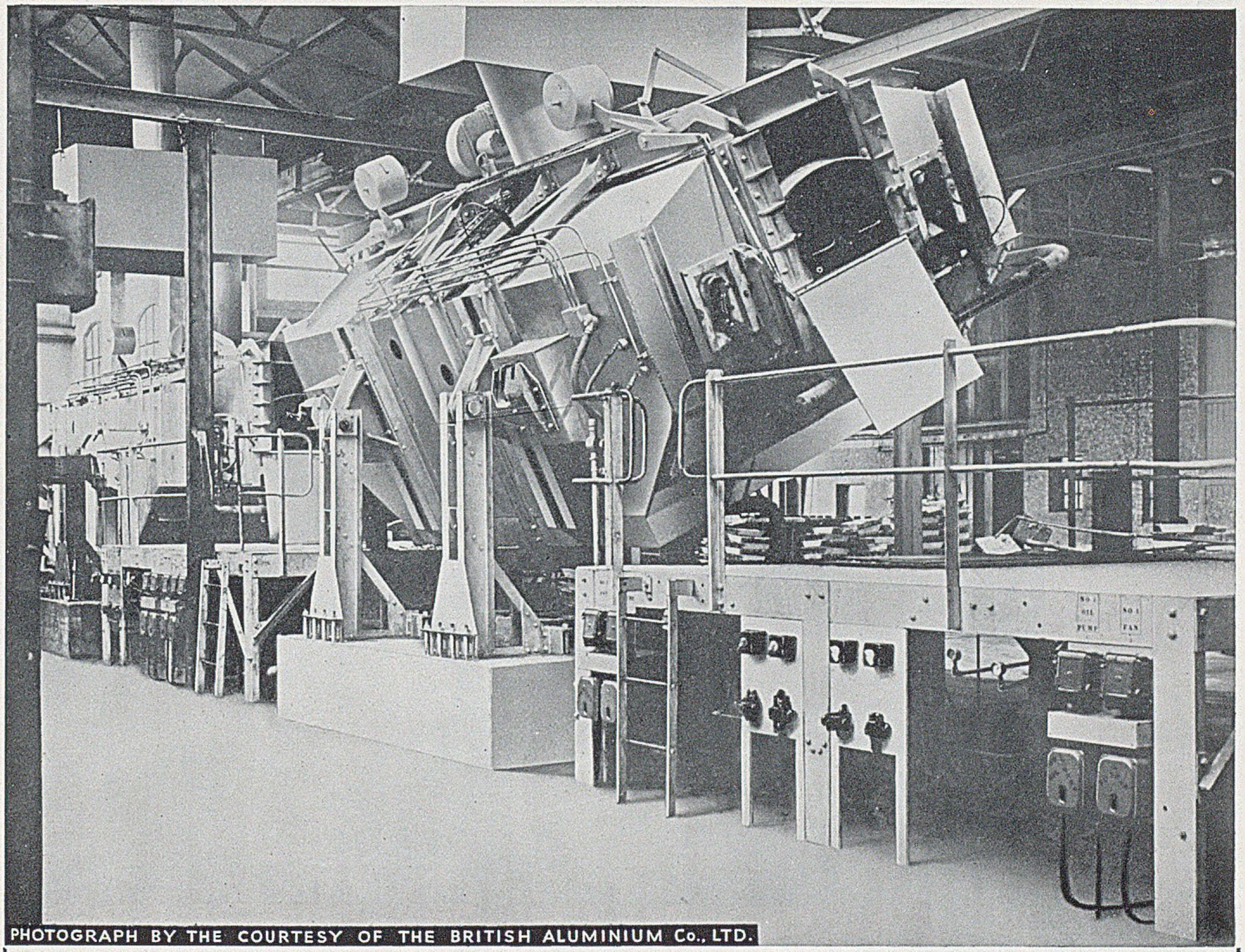
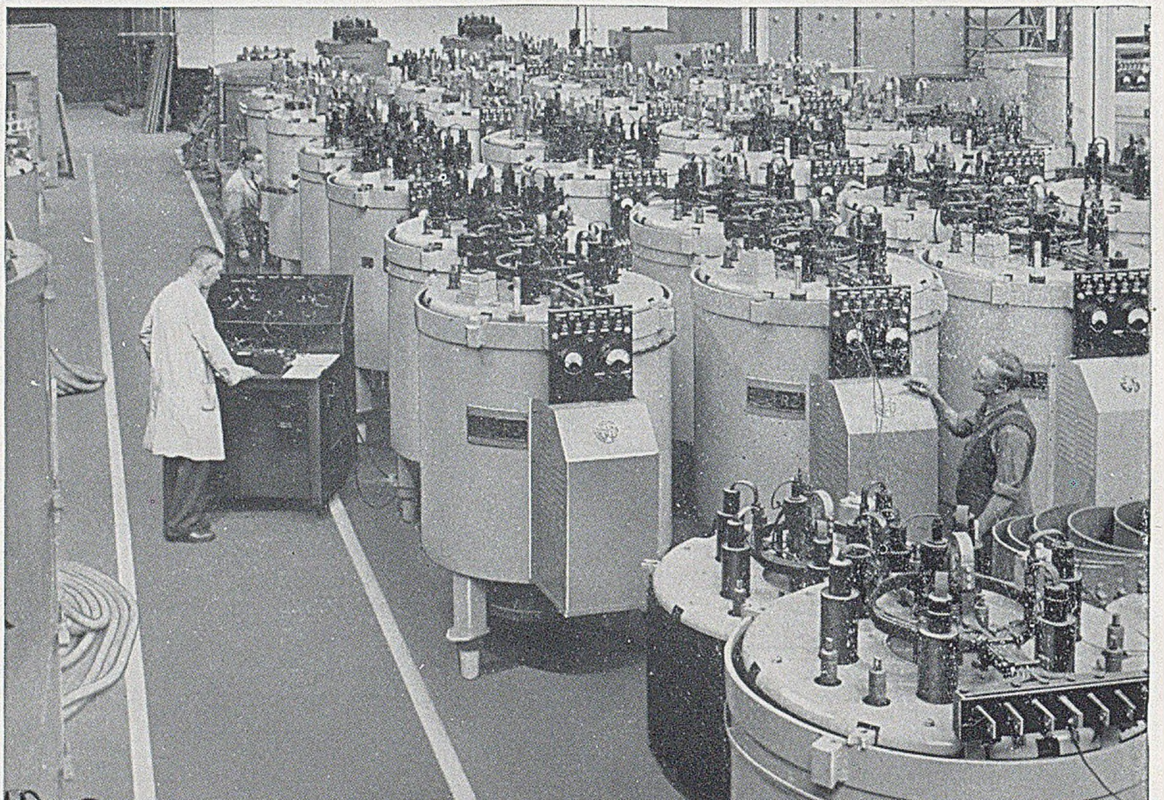
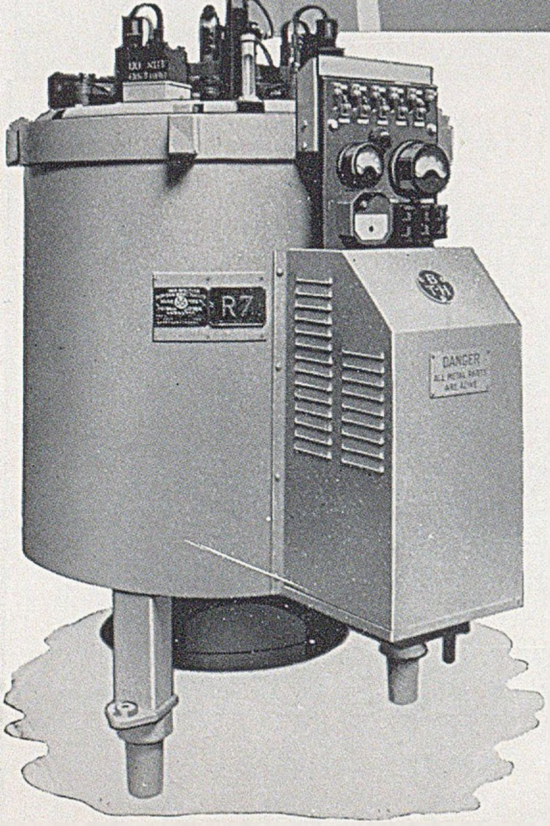


Rolling Mill Equipment





Pumpless Steel-tank Mercury-arc Rectifiers

Mercury-arc rectifiers have established themselves as an accepted type of electric current conversion apparatus for most applications.

The illustration shows a 21,000-kW., 800-volt, grid-controlled, mercury-arc rectifier equipment, comprising thirty pumpless, air-cooled, steel-tank, grid-controlled units giving a 60-phase rectification, being tested before despatch from the BTH Rugby Works.

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The Nimonic Alloys

These nickel-chromium alloys were developed primarily to meet requirements for materials which, in addition to good resistance to oxidation, have outstanding resistance to creep at the high temperatures encountered in gas turbines. There are, on the other hand, important applications where creep properties are not so critical and, accordingly, materials which have not undergone a creep test may safely be specified. In such cases, the alloys, although differently coded, are of similar compositions and are subjected to the same care in manufacture.

ALLOYS AVAILABLE

GROUP 1 CREEP-TESTED ALLOYS		GROUP 2 WROUGHT ALLOYS		GROUP 3 CASTING ALLOYS
Name.	Specification.	Name.	Specification.	Name.
Nimonic 80	D.T.D. 725	Nimonic C	—	Nimonic CC
Nimonic 80A	D.T.D. 736	—	—	—
Nimonic 90	D.T.D. 747	Nimonic B	—	Nimonic CB
		Nimonic 75	D.T.D. 703	—
		Nimonic F	D.T.D. 714	Nimonic CF
		Nimonic D	—	—

TYPICAL CREEP CHARACTERISTICS OF NIMONIC ALLOYS IN THE FULLY HEAT-TREATED CONDITION

Stress in tons/sq. in. required to produce a creep extension of 0.1 per cent at specified times and temperatures.

TIME & TEMPERATURE DEGREES C.	NIMONIC 80	NIMONIC 80A	NIMONIC 90
100 Hour Data.			
600	29.5	—	—
650	21.0	26.0	31.0
700	16.5	19.0	24.0
750	10.0	14.5	17.0
815	—	—	10.0
300 Hour Data.			
600	25.5	—	—
650	19.0	23.0	29.0
700	14.0	15.5	21.5
750	8.0	11.5	14.5
815	—	—	8.0
1,000 Hour Data.			
600	22.0	—	—
650	16.5	20.0	26.0
700	11.0	11.5	19.0
750	6.0	8.5	12.0
815	—	—	6.0
5,000 Hour Data.			
600	18.0	—	—
650	13.5	16.0	22.5
700	—	6.0	16.0
750	—	4.5	9.0
815	—	—	3.0

AVAILABLE FORMS

Nimonic 80, Nimonic 80A and Nimonic 90. Hot-rolled bar, black or machined finish; hot-rolled, cold-drawn, forged and extruded sections; forged discs and blocks; hot-rolled plate; cold-rolled sheet and strip; cold-drawn wire; hot pressings; cold-drawn tube; drop stampings.

Nimonic 75, Nimonic F, Nimonic D, Nimonic B and Nimonic C. Hot-rolled plate; cold-rolled sheet and strip; forged discs and blocks; hot-rolled bar, black or machined finish; hot-rolled, cold-rolled, cold-drawn, forged and extruded sections; cold-drawn wire; cold-drawn tube; extruded hollows; drop stampings, tubes and pressed components.

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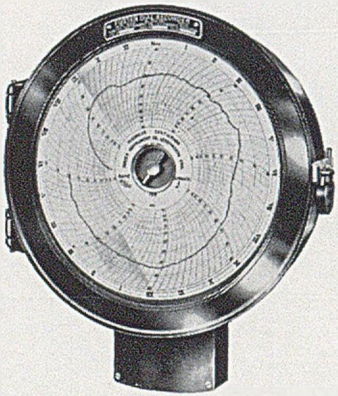

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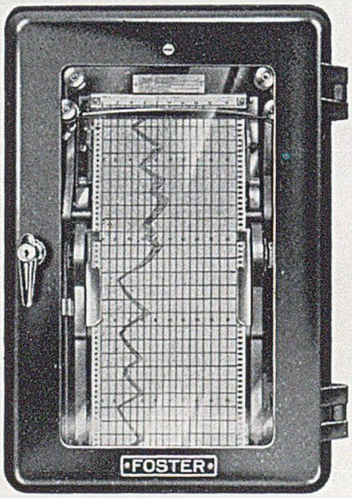

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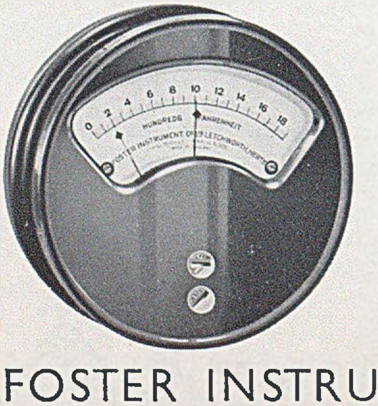

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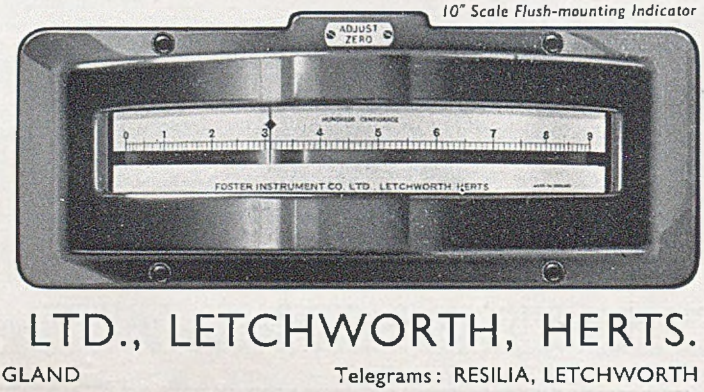

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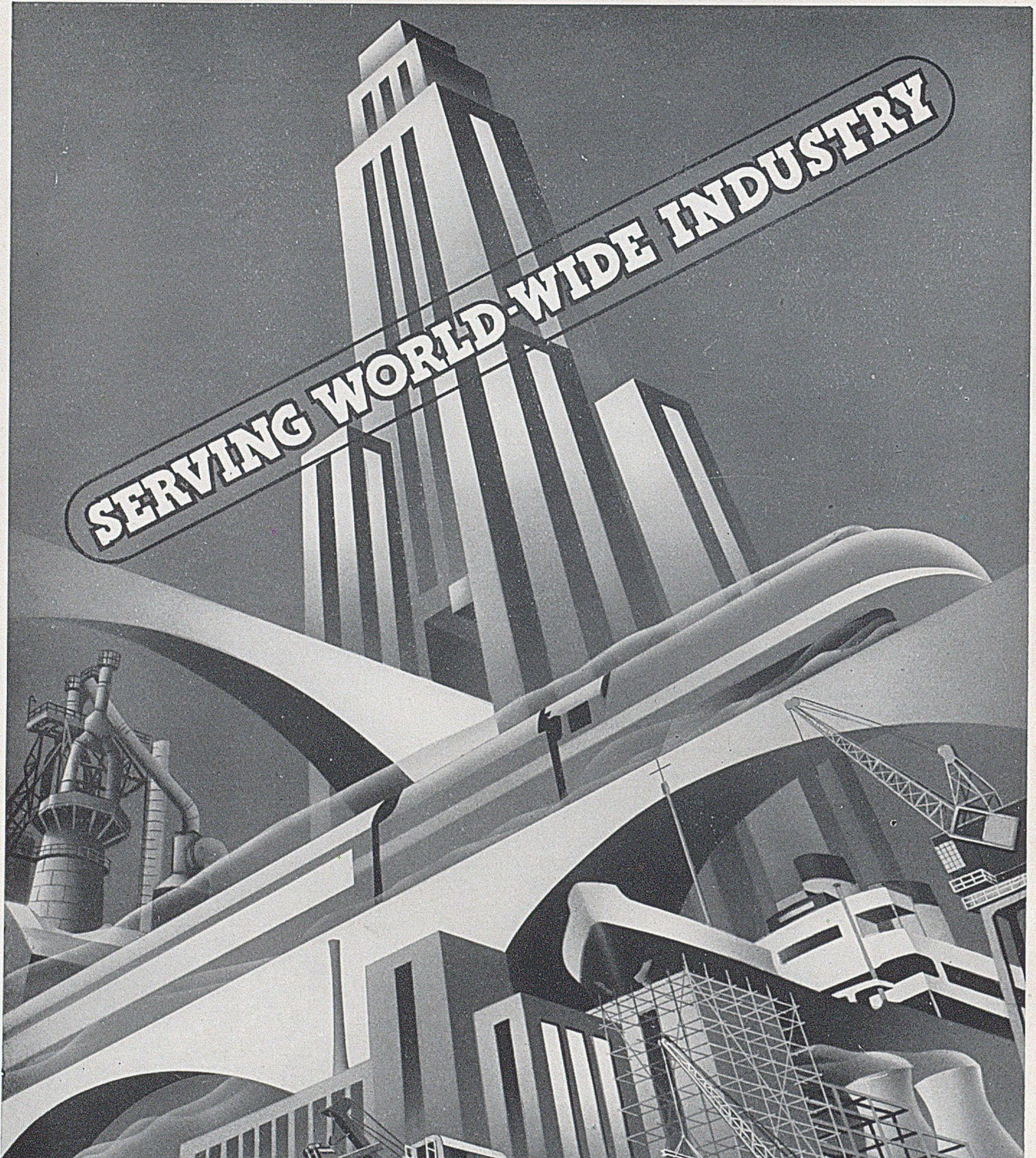
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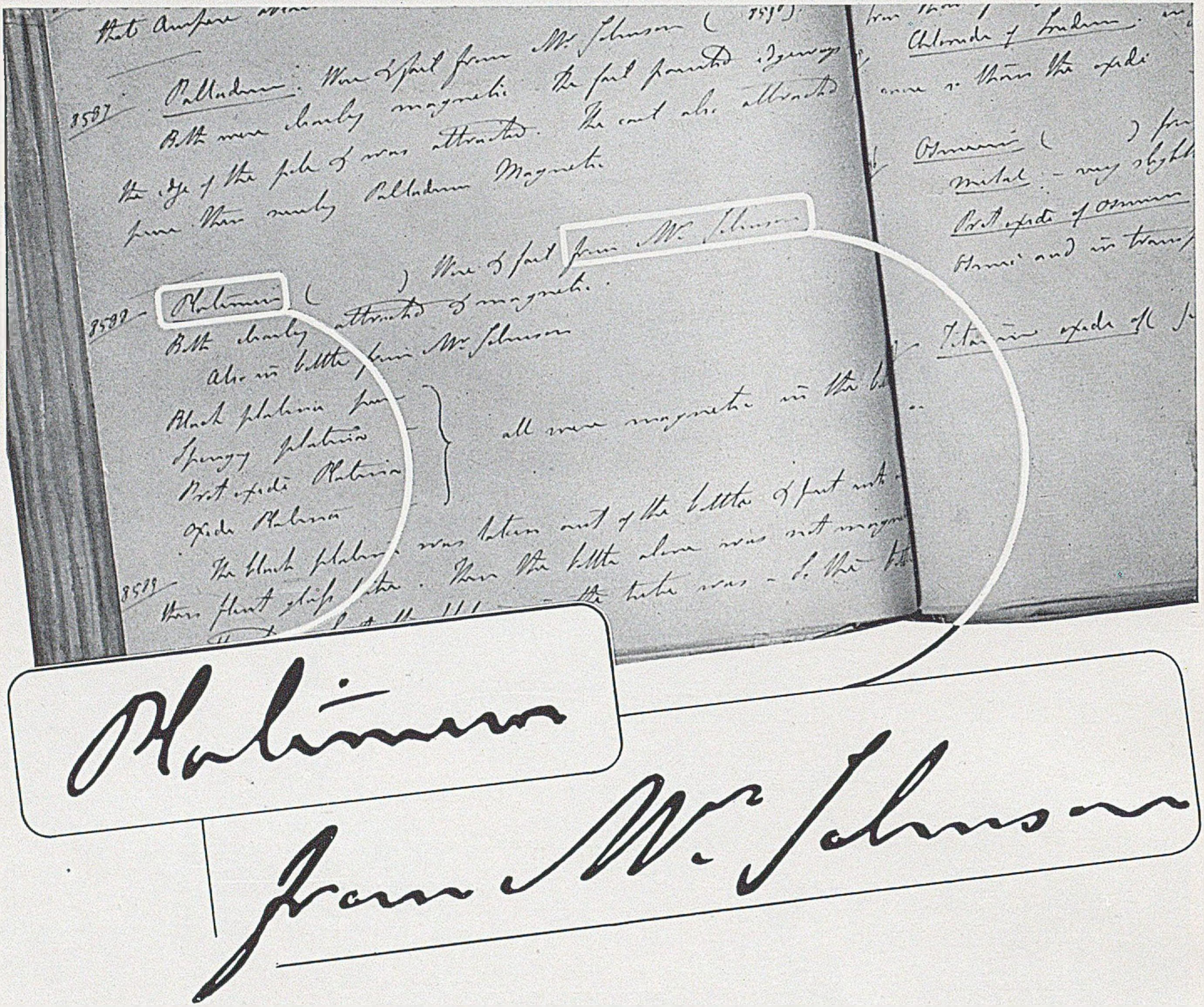
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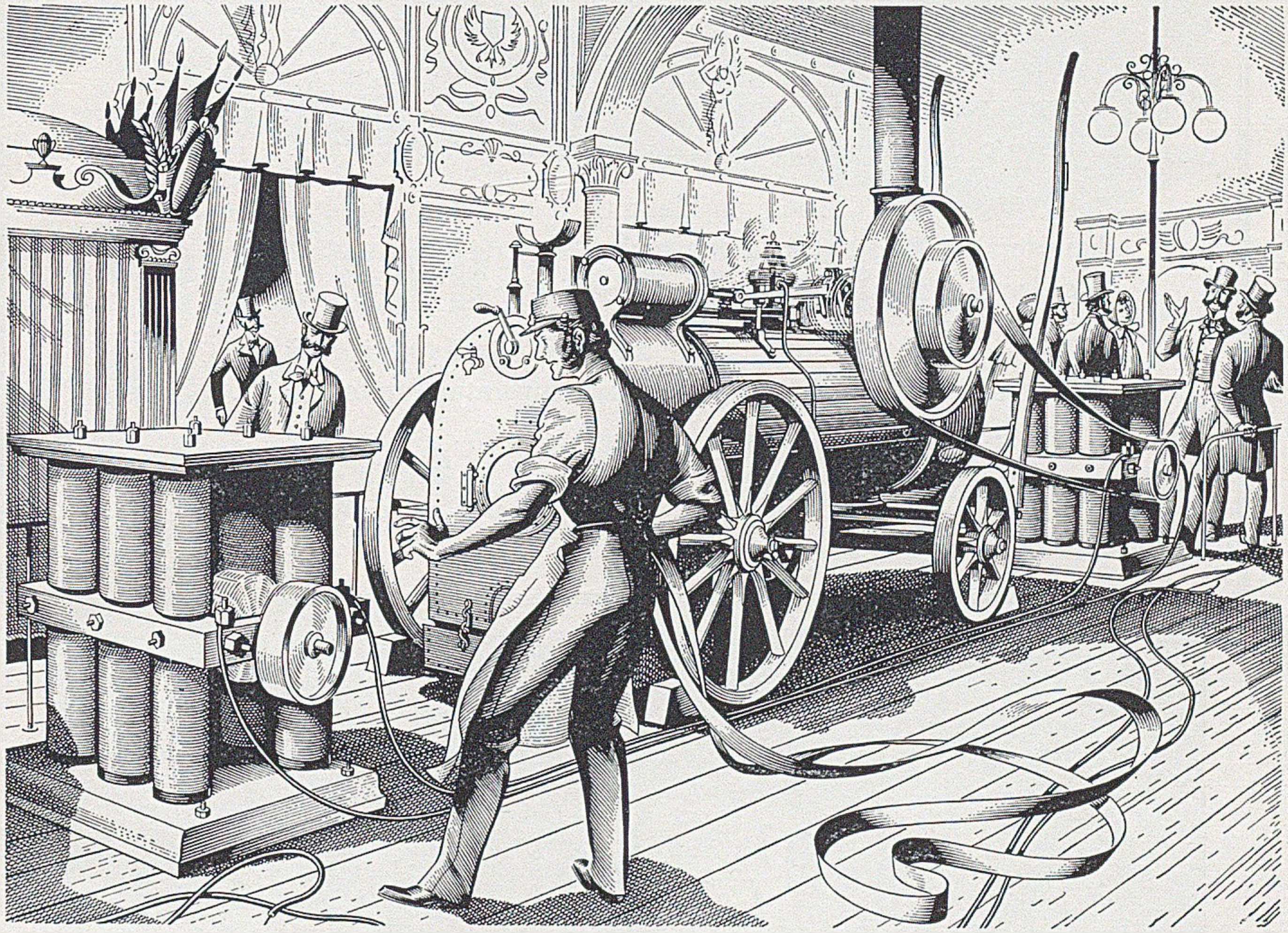

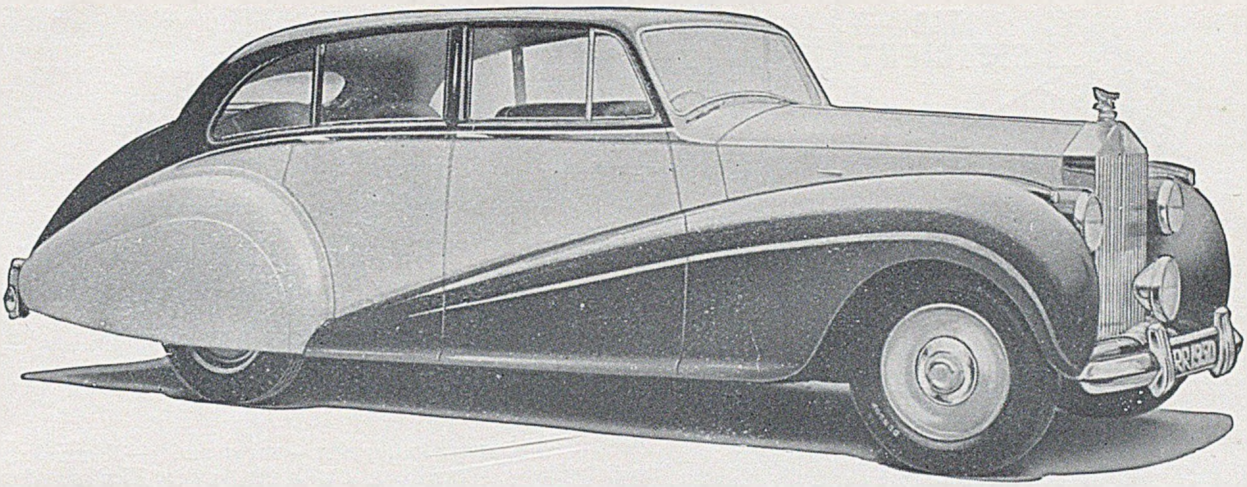

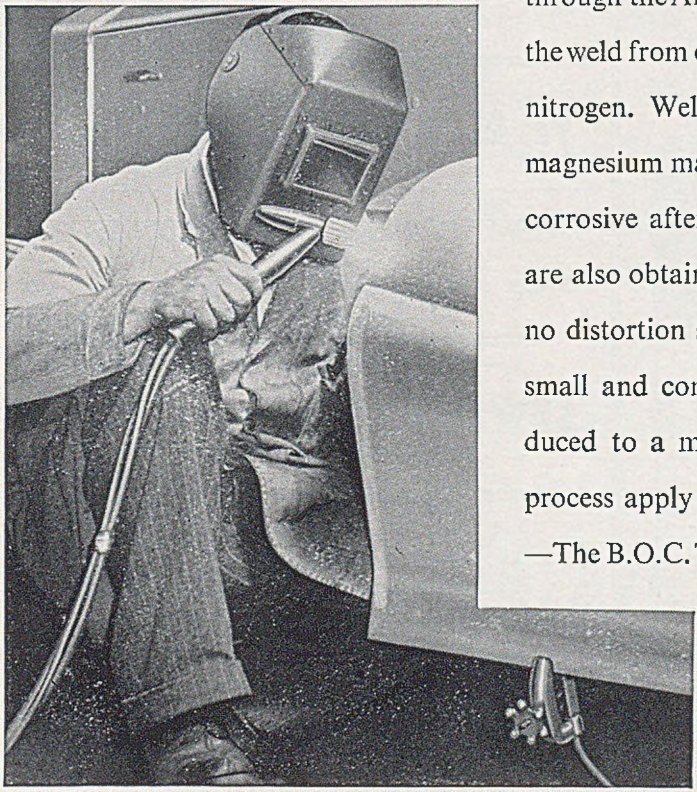

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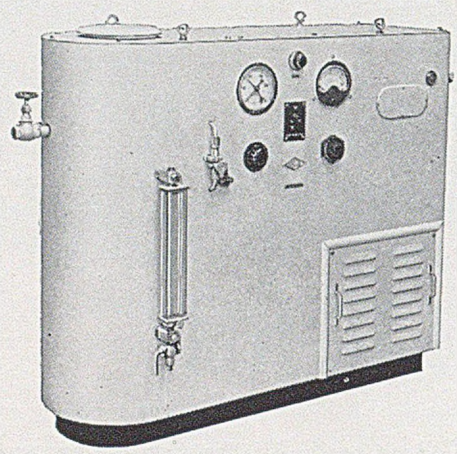

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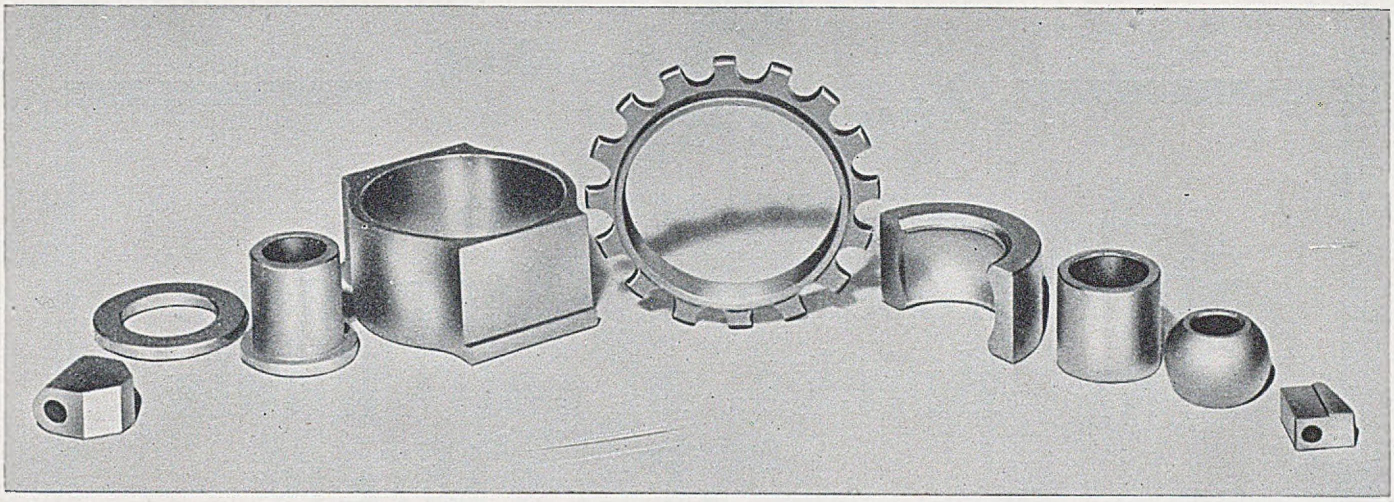

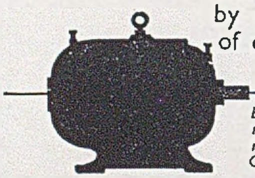

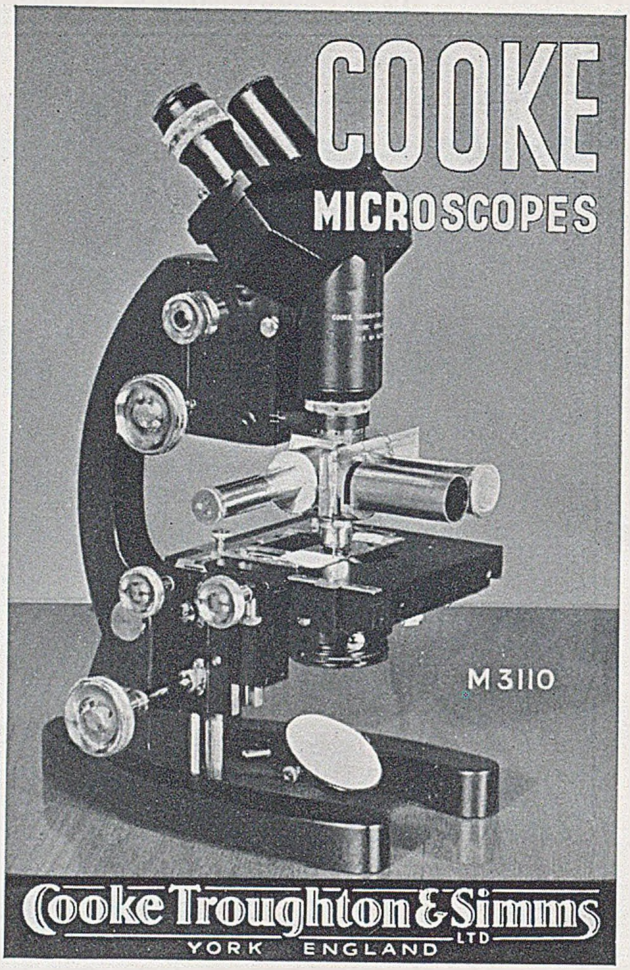

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BULLETIN

OF THE INSTITUTE OF METALS

VOLUME 1 SEPTEMBER 1951 PART 1

THE NEW JOURNAL

An editorial notice in the August issue said a brief farewell to the octavo *Journal*, which all of us have known since our first acquaintance with the Institute. The Report of Council for 1950 had prepared us for the change, and now we have in our hands the first number with the quarto page. The larger page is accompanied by a new design of paper cover for the monthly *Journal*. I well know the anxious thought that your Council has given to both these matters before deciding to make a change, and I sincerely hope that members will approve the action which has been taken.

The increase in size of page is, of course, the more important of the changes. In making it we have the benefit of the experience of many other important scientific and technical societies, including the Royal Society, the Institution of Electrical Engineers, the Institution of Mechanical Engineers, the Iron and Steel Institute, and the Physical Society. The periodicals of these societies confirm our expectation that the new format of our *Journal* will permit larger illustrations where desirable and a more dignified presentation of papers, the less frequent breaks at page endings being especially pleasing and making for easier reading.

The change in format has mainly, however, been decided upon for financial reasons. It will be known to all members that the cost of producing the *Journal* has risen in far greater proportion than the amount of subscriptions, and the problem before the Council has been to find ways of maintaining and, if possible, increasing the service rendered by the Institute without adding to the burden on individual members. The Council has been advised that an increase to a quarto page would bring financial advantages and, since these would coincide with improved presentation of papers, it was felt that the Council would have been failing in its duty if it had not grasped the opportunity thus presented.

Many will naturally regret the necessity for the disappearance of the familiar small-paged *Journal*, but I believe there will be a warm welcome from all members for the new *Bulletin* in which I have the honour to include this message. It is intended that the *Bulletin* shall incorporate the Institute news and announcements, the news of Local Sections and Associated Societies, news of kindred societies and of joint activities, which have hitherto appeared at the beginning of the monthly *Journal*. I hope and confidently expect that the *Bulletin* will also become a medium for the discussion, by correspondence, of all non-ferrous metallurgical topics other than mining and extraction. The pages of this new publication should reflect the views of members on subjects which, although they may not be appropriate for placing on record with the scientific papers contributed for discussion at General Meetings of the Institute, are nevertheless of importance for the progress of metallurgical science and practice. From time to time papers prepared for Local Section meetings are recommended as being of interest to the general membership of the Institute, even though it may not be claimed that they possess the features of original work or a new survey which are looked for in papers published in the annual proceedings of the Institute. The *Bulletin* will often be the appropriate vehicle for these communications.

Taste in covers varies: some will applaud, others will be less enthusiastic over the design which graces this first issue in the new format. We believe that, apart from its intrinsic merit, it is likely to have an appeal for readers who may not be familiar with the aims and activities of the Institute's work and whom it is desired to attract to read our news or proceedings and the advertisements. All of us are anxious that the cover should be worthy of the Institute, and this I believe has been achieved, but let us reflect that the paper cover of the monthly *Journal* is in any case an ephemeral thing which will disappear when the contents are bound in the cloth boards of the annual volumes.

Publication is the function which absorbs the greater part of the income of the Institute, and I thought it fitting when so important a change was to be made in the appearance and content of the *Journal* that your President should re-state the considerations which had led to this step being taken. I do not feel that my statement needs to be defensive or apologetic; rather do I commend our new *Journal* to you as evidence of vigorous health in our constitution, and as a production worthy of the high traditions of the Institute.

A. J. Murphy

INSTITUTE NEWS

Election to Fill a Casual Vacancy on the Council

In accordance with the powers granted in Article 28, the Council has nominated Mr. N. I. BOND-WILLIAMS, B.Sc., Chairman and Managing Director, Aston Chain and Hook Co., Ltd., Birmingham, to fill a vacancy for an Ordinary Member of Council, caused by the death of Mr. Harry Davies.

Mr. Bond-Williams will, as required by Article 28, present himself for re-election at the 1952 Annual General Meeting.

Election of Council for 1952–53

In accordance with the Articles of Association, the following are due to retire from the Council at the 1952 Annual General Meeting:

President:

Professor A. J. MURPHY, M.Sc.

Past-President:

Colonel Sir PAUL GUETERBOCK, K.C.B., D.S.O., M.C., T.D., D.L., J.P., M.A., A.D.C.

Vice-Presidents:

Major C. J. P. BALL, D.S.O., M.C.
C. J. SMITHELLS, M.C., D.Sc.

Honorary Treasurer:

W. A. C. NEWMAN, O.B.E., B.Sc., A.R.S.M., A.R.C.S., D.I.C.

Ordinary Members of Council:

D. F. CAMPBELL, M.A., A.R.S.M.
T. M. HERBERT, M.A.
H. W. G. HIGNETT, B.Sc.
A. R. POWELL

In addition, Mr. ALFRED BAER and Mr. BOND-WILLIAMS, who were appointed by the Council to fill casual vacancies during the year, also retire and present themselves for re-election, in accordance with Article 28.

Under Article 19, Professor A. J. MURPHY will fill the vacancy on the Council as Past-President.

In accordance with Article 22, the Council nominates the following members to fill the other vacancies:

As President:

C. J. SMITHELLS, M.C., D.Sc., Director of Research, The British Aluminium Company, Ltd., Gerrards Cross.

As Vice-Presidents:

G. L. BAILEY, M.Sc., Director, British Non-Ferrous Metals Research Association, London.

S. F. DOREY, C.B.E., D.Sc., Wh.Ex., F.R.S., Chief Engineer Surveyor, Lloyd's Register of Shipping, London.

As Honorary Treasurer:

E. H. JONES, Director and General Manager (Bristol), Capper Pass and Son, Ltd., Bristol.

As Ordinary Members of Council:

N. P. INGLIS, Ph.D., M.Eng., Research Director, Imperial Chemical Industries, Ltd., Metals Division, Birmingham.

IVOR JENKINS, D.Sc., Chief Metallurgist, Research Laboratories, General Electric Co., Ltd., Wembley.

A. G. RAMSAY, B.Sc., Ph.D., Director, The Mond Nickel Co., Ltd., and Works Manager, Clydach Refinery.

H. SUTTON, D.Sc., Director of Research and Development, Aircraft Materials, Ministry of Supply, London.

Major P. LITHERLAND TEED, A.R.S.M., Deputy Chief of Aeronautical Research and Development, Vickers-Armstrongs, Ltd., Weybridge.

W. J. THOMAS, Assistant Managing Director, The British Aluminium Co., Ltd., London.

Members are reminded that, in accordance with Article 22, any ten members may also at, or before, the business part of the Autumn General Meeting (16–18 September 1951), nominate in writing, with the written consent to act if elected of the person nominated, any duly qualified person other than one of those nominated by the Council to fill any vacancy on the Council, but each such nominator is debarred from nominating any other person for the same election. If two or more persons are nominated for any honorary office they (or such of them as are not Ordinary Members of Council who are not retiring at the next Annual General Meeting) will be deemed to have been nominated also for any vacancies among the Ordinary Members of Council. No person is eligible to fill any vacancy at such Annual General Meeting unless he has consented in writing to be nominated and has been nominated or deemed to be nominated for the same in compliance with this Article.

Senior Vice-President, 1952–53

The Council has elected Professor F. C. THOMPSON, D.Met., M.Sc., to serve as Senior Vice-President for the year 1952–53, and he will be their nominee for the Presidency in 1953–54.

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

The following papers have been invited for this Symposium, which is being arranged in connection with the 1952 Annual General Meeting:

(1) "Electric Furnaces for the Thermal Treatment of Non-Ferrous Metals", by an author selected by the Electric Furnace Section of the British Electrical and Allied Manufacturers' Association.

(2) "Gas-Fired Furnaces for the Thermal Treatment of Non-Ferrous Metals", by Mr. J. F. Waight (West Midlands Gas Board).

(3) "Batch Thermal Treatment of Light Alloys", by Mr. C. P. Paton (Northern Aluminium Co., Ltd.).

(4) "Batch and Continuous Annealing of Copper and Copper Alloys", by Mr. Edwin Davis and Mr. S. G. Temple (Imperial Chemical Industries, Ltd., Metals Division).

(5) "Flash-Annealing of Light Alloys", by Mr. R. T. Staples (T.I. Aluminium, Ltd.).

(6) "Bright Annealing of Nickel and Its Alloys", by Mr. H. J. Hartley (Henry Wiggin and Co., Ltd.).

(7) "Continuous Thermal Treatment of Light Metals", by M. M. Lamourdedieu (Société Centrale des Alliages Légers).

The papers will appear together in the January or February 1952 issue of the *Journal*.

Informal Discussion on "Tool and Die Materials for the Hot Working of Non-Ferrous Metals"

The special General Meeting on this subject, previously announced, will be held at the University of Birmingham on the afternoon of Thursday, 3 January 1952. The chair will be taken by Mr. Christopher Smith.

Election of Members

The following 30 Ordinary Members, 1 Junior Member, and 6 Student Members were elected on 18 July 1951:

As Ordinary Members

Allard, Marc., Ing. civil des Mines, Directeur, Institut de Recherches de la Sidérurgie, 185 rue President Roosevelt, Saint Germain-en-Laye (S. et O.), France.

Blyth, Howard Neville, B.A., Lecturer, Royal School of Mines, Prince Consort Road, London, S.W.7.

Bunton, John D., Chief Metallurgist, C.A.V. Ltd., Warple Way, Acton, London, W.3.

Chen, Neng-Kuan, Dr. Eng., Research Associate, Department of Mechanical Engineering, The Johns Hopkins University, Baltimore 18, Md., U.S.A.

Claxton, Cyril Charles, Joint Managing Director, Sheffield Twist Drill and Steel Company, Ltd., Sheffield.

Ellis, Dennis Thomas, Chief Metallurgist, Jaguar Cars, Ltd., Coventry.

English, Alan, Blast Furnace Assistant, Appleby-Frodingham Steel Co., Ltd., Scunthorpe, Lincs.

Fattah, Mohamed Ahmed Abdel, B.Sc., Engineer, Ministry of Defence, Cairo, Egypt.

Harris, Arthur Clement, Assoc. Met., Research Metallurgist, Simon-Carves, Ltd., Cheadle Heath, Stockport.

Haut, Frederick Joseph Georg, B.Sc., Constructional Engineer, Boxmoor Engineering Corporation, Ltd., Watford, Herts.

Jones, *Sir* Lewis, J.P., Secretary, The South Wales Siemens Steel Association, North Hill, 2 St. James' Crescent, Swansea.

Lewis, George Leonard, Sales Engineer, Vickers (Eastern), Ltd., Killick Building, Home Street, Bombay, India.

Littlewood, Frederick Richard, Chief Designer, British Sewing Machines, Ltd., Lombard Road, London, S.W.19.

Lombardi, *Dott.* Paolo, Technical Management Official, Società Metallurgica Italiana, Borgo Pinti 99, Florence, Italy.

O'Connor, Kenneth, Manager, Foundry Sheet and Strip Division, Dominion Foils (Canada) Ltd., Cap de la Madeleine, Quebec, Canada.

Orlando, *Dott.* Giuseppe, Vice-Chairman, Società Metallurgica Italiana, Borgo Pinti 99, Florence, Italy.

Peplow, Douglas Boraston, Metallurgist, British Electricity Authority, Southern Division, 111 High Street, Portsmouth.

Prasad, Rajendra, B.Sc., Assistant Professor of Metallurgy, College of Mining and Metallurgy, Benares Hindu University, Benares, India.

Pryor, Horace, Works Manager, Humber Ltd., Coventry.

Robinson, Eric A., M.C., Managing Director, The Superheater Company, Ltd., 53 Haymarket, London, S.W.1.

Rushworth, David, Fuel Economist, Albion Chambers, King Street, Nottingham.

Salah el-Din Nessim, Ahmad, B.Sc., Engineer, Ministry of Defence, Cairo, Egypt.

Sandberg, Alexander Christer Edward, B.Sc., Consulting Engineer, 40 Grosvenor Gardens, London, S.W.1.

Sargeant, Colin Wilfred, Works Manager, James Evans and Son, Ltd., Victoria Works, Bromford Lane, West Bromwich.

Scortecci, *Dott.* Massimo, General Secretary, Istituto Italiano della Saldatura, Via XX Settembre 8, Genoa, Italy.

Snedden, George T., Colvilles, Ltd., Crosshill Street, Motherwell, Lanarkshire.

Sparkes, Edward John, Chemist, Blackburn Generating Station, British Electricity Authority, Blackburn, Lancs.

Talini, *Dott.* Renzo, Chief of Chemical Laboratories, Società Metallurgica Italiana, Fornaci di Barga, Lucca, Italy.

Wood, H. Carrington, Wellman Smith Owen Engineering Corporation, Ltd., 25 Wilton Road, London, S.W.1.

Yates, Harry, Managing Director, Smith and McLean, Ltd., 179 West George Street, Glasgow, C.2.

As Junior Member

Mitchell, Neville M., Metallurgist, Royal Aircraft Establishment, Farnborough, Hants.

As Student Members

Braine, William Alan, Staveley Iron and Chemical Company, Ltd., Chesterfield.

McLennan, John Andrew, Trainee Metallurgist, Metal Manufactures, Ltd., Port Kembla, N.S.W., Australia.

Mirza, Mohammad Razi, B.A., Technical Apprentice, John I. Thornycroft and Co., Ltd., Woolston, Southampton.

Naylor, Graham L., Metallurgical Apprentice, Stewarts and Lloyds, Ltd., Bilston, Staffs.

Plackett, John Ronald, Student of Metallurgy, University of Birmingham.

Rolls, Roger, Research Assistant, Taylor Brothers and Co., Ltd., Trafford Park Steel Works, Manchester, 21.

PERSONALITIES

Mr. A. M. Baer

(Member of Council)

Alfred Baer was born in London in 1896 and educated at Harrow and Balliol College, Oxford. After his education had been interrupted by the 1914–18 War, in the course of which he served in Gallipoli and France, Mr. Baer took a B.A. degree and was called to the Bar of the Inner Temple. He entered industry when he joined Messrs. Henry Gardner and Co., Ltd., in 1920; he was appointed a director of the company in 1927. Throughout the last war he acted as Joint Controller of Non-Ferrous Metals. Last year he was appointed Vice-Chairman of The Consolidated Zinc Corporation, Ltd. In addition, Mr. Baer is Chairman of H. J. Enthoven and Sons, Ltd., a member of the Eastern Gas Board, and a director of The Zinc Corporation, Ltd., Imperial Smelting Corporation, Ltd., and New Broken Hill Consolidated, Ltd., among other companies.

[*Elliott and Fry, Ltd.*

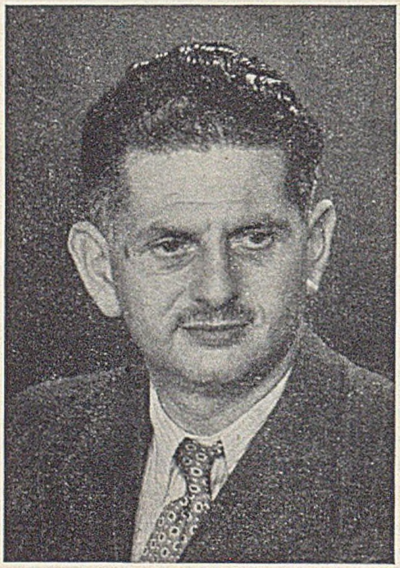
Mr. E. A. G. Liddiard

(Chairman of the London Local Section)

Edwin Andrew Guthrie Liddiard was born in 1903 and educated at Christ's Hospital. From 1922 to 1925 he was a laboratory assistant with Messrs. Cammell, Laird and Co., Ltd., Sheffield, and studied at the University in the evenings. The next three years were spent at Cambridge, where he took the Natural Science Tripos, Part I, and in his final year studied metallurgy under Heycock. From 1928 to 1932 Mr. Liddiard was a Research Metallurgist with I.C.I. (Fertilizers and Synthetic Products), Ltd., Billingham. He then joined the British Non-Ferrous Metals Research Association as Assistant Development Officer. Later he was appointed Assistant Research Superintendent (1937) and Research Manager of the Association (1943). When the Fulmer Research Institute was established in 1946, Mr. Liddiard was appointed Director, a position he still holds. He has taken an active part in Institute of Metals affairs and has for the last two years been a member of the Publication Committee.

Dr. H. M. Finniston

(Chairman of the Oxford Local Section)



Harold Montague Finniston was born in 1912 and educated in Glasgow, where he attended the University and the Royal Technical College. He graduated with first-class honours in metallurgy in 1933. After a period in industry with Messrs. Stewarts and Lloyds, Ltd., he returned to Glasgow in 1937 as Chief Research Officer of the Scottish Coke Research Committee. Having gained his Ph.D. degree in 1940, Dr. Finniston joined the Bragg Laboratory of the Naval Ordnance Inspection Department, Sheffield, where he was in charge of the Metallurgy Section until 1946, when he was seconded by the Royal Naval Scientific Service to the Department of Atomic Energy. He went to Canada to organize the metallurgical side of the atomic-energy work there, and on his return in the following year became Deputy Head of the Metallurgy Division of the Atomic Energy Research Establishment at Harwell. In 1948 he was appointed to his present post of Head of the Metallurgy Division, with the rank of Deputy Chief Scientific Officer.

Mr. John Arnott

(Chairman of the Scottish Local Section)

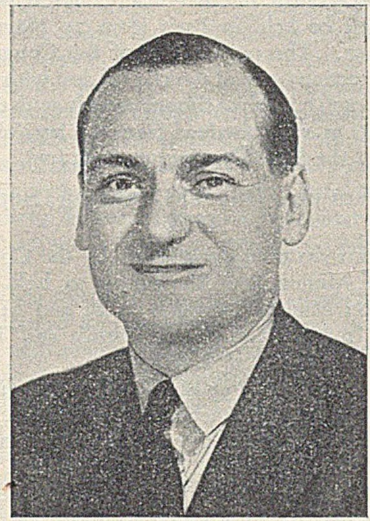
John Arnott was born in 1891 and educated at Ardrossan Academy. After leaving school, he became a premium student in the laboratory of Messrs. Tatlock and Thomson, city analysts of Glasgow, and a student in the Royal Technical College, Glasgow. He later spent some time in a small consulting laboratory where the work was mainly metallurgical. Joining the firm of G. and J. Weir, Ltd., Cathcart, in 1913, he has remained with them ever since, and is now their Chief Metallurgist. Mr. Arnott is the author of many papers and articles on non-ferrous metallurgical subjects. He is a Fellow of the Institution of Metallurgists, a Fellow of the Royal Institute of Chemistry, and a member of the Society of Chemical Industry, the Institute of British Foundrymen, the West of Scotland Iron and Steel Institute, and other bodies.

Mr. M. M. Hallett

(Chairman of the Sheffield Local Section)

Michael Mountjoy Hallett was born in 1911 and educated at King's School, Worcester, and University College, Cardiff, where he graduated in metallurgy in 1931 and took the M.Sc. degree in 1939. After graduation, he was awarded an industrial bursary by the Royal Commissioners for the Exhibition of 1851, and spent the next two years in the Research Laboratory of The Mond Nickel Co., Ltd., Birmingham. In 1933 he was appointed to the permanent staff of the same laboratory and worked on the scaling of low-alloy steels, on the development of heat-resisting alloys, and particularly on the various types of nickel cast iron. In 1943 he was appointed Chief Metallurgist of the Sheepbridge Stokes Centrifugal Castings Co., Ltd., and in 1949 became

Chief Metallurgist to the Sheepbridge Engineering, Ltd., Group. He has published papers on metallurgical aspects of alloy cast irons, and on corrosion-resisting alloys. He is a member of the Institute of Marine Engineers and of various other societies, both British and American. He serves on several committees of the British Cast Iron Research Association and the Institute of British Foundrymen.



PERSONAL NOTES

Mr. V. T. **Balasubramanian** has returned to India, where he will be in charge of the foundry of India Pistons, Ltd., Madras.

Dr. C. S. **Barrett**, of the Institute for the Study of Metals, Chicago, is visiting this country and will for a time be working in the Metallurgy Department at Birmingham University.

Mr. K. W. J. **Bowen** has been awarded the degree of Ph.D. of Cambridge University.

Mr. C. L. M. **Cottrell** has been appointed Assistant Chief Metallurgist (ferrous metals) to the British Welding Research Association.

Mr. C. J. **Dangerfield** has transferred from the Research Laboratories of The British Aluminium Co., Ltd., to the Company's works at Warrington.

Mr. H. E. **Dixon** has been appointed Chief Metallurgist (non-ferrous metals) to the British Welding Research Association.

Mr. J. **Dodd** has been awarded the degree of B.Sc. in Applied Science (Metallurgy) at Durham University.

Mr. A. G. **Duce** has been awarded the Ph.D. degree of Cambridge University.

Dr. **Hugh Ford** has been appointed Professor of Applied Mechanics in the University of London. The appointment, which is tenable at Imperial College, will have effect from 1 October 1951.

Mr. E. R. **Francis** has been appointed Technical Information Officer to The Mond Nickel Co., Ltd.

Dr. **Max Hansen** has been appointed Assistant Chairman of the Metals Research Department of the Armour Research Foundation of the Illinois Institute of Technology, Chicago. He was previously Supervisor of Physical Metallurgy Research of the Foundation.

Mr. A. R. **Harding** has been awarded the degree of Ph.D. of Birmingham University.

Mr. P. T. **Houldcroft** has been appointed Assistant Chief Metallurgist (non-ferrous metals) to the British Welding Research Association.

Mr. G. **Lucas** has been awarded the Ph.D. degree of Durham University.

Mr. G. H. **Noordhof** has been appointed Talks Producer in the B.B.C. Television Service and will be concerned mainly with scientific and medical programmes.

Dr. E. **Orowan** has been appointed George Westinghouse Professor of Mechanical Engineering at the Massachusetts Institute of Technology.

Mr. A. R. **Palmer** has left England to take up an appointment with the Colonial Development Corporation in Tanganyika.

Miss E. W. **Parker** has been appointed Assistant to the Manager of the Development and Research Department of The Mond Nickel Co., Ltd.

Mr. M. S. **Paterson** is returning from the Institute for the Study of Metals, Chicago, to Australia, where his address will be Aeronautical Research Laboratories, Department of Supply, Box 4331, G.P.O., Melbourne, C.1.

Dr. K. **Winterton** has been appointed Chief Metallurgist (ferrous metals) to the British Welding Research Association.

Mr. A. **Wipperman**, for a number of years on the staff of the Industrial Division of Johnson, Matthey and Co., Ltd., has recently left for Australia, where he will take up a senior appointment with the Company's Australian associate, Garrett, Davidson, and Matthey Pty., Ltd., 824 George Street, Sydney, N.S.W.

Deaths

The Editor regrets to announce the deaths of:

Mr. **Frederick William Dingwall** (Original Member), Chairman and Managing Director of Everitt and Co., Ltd., Liverpool.

Mr. **Alfred Holley Mundey** on 5 July at the age of 83. He was formerly Technical Director of Fry's Metal Foundries, Ltd., and served on the Council of the Institute in his capacity as Chairman of the London Local Section in 1927–28 and as an Ordinary Member of Council from 1930 to 1938.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

Official Visits by the President

The President, accompanied by the Secretary, will pay official visits to Local Sections and Associated Societies as detailed below:

Wed., 7 Nov. 1951.	Manchester Metallurgical Society.
Thur., 8 Nov. 1951.	London Local Section.
Tues., 15 Jan. 1952.	South Wales Local Section.

Leeds Metallurgical Society

Dr. R. Goodacre has succeeded Mr. F. Lindars as Secretary of the Society.

JOINT ACTIVITIES

Joint Committee on Metallurgical Education

Mr. D. R. O. Thomas has accepted an invitation to serve as Chairman of the Joint Committee in succession to Mr. E. G. Lawford.

Notch-Bar Testing of Materials in Relation to Welded Construction

The Joint Committee on Materials and Their Testing, in association with the Institute of Welding, announces a Symposium on " Recent Developments in Notch-Bar Testing of Materials and Their Relation to Welded Construction," to be held in London on Wednesday, 5 December 1951.

Papers are being invited from a number of British and foreign experts and will be preprinted and introduced by a rapporteur, with a view to discussion, which will occupy two sessions, one in the morning and one in the afternoon. Further particulars will be issued shortly.

NEWS OF KINDRED SOCIETIES

The Institution of Metallurgists

The following elections have recently been announced:

As Fellows

F. A. Forward (University of British Columbia), B. R. Nijhawan (National Metallurgical Laboratory, Jamshedpur, India), J. R. Rait (Hadfields, Ltd.), J. Taylor (Royal Technical College, Glasgow).

As Fellows from the Grade of Associate

J. W. Houghton (Park Gate Iron and Steel Co., Ltd.), H. Padget (Dorman Long and Co., Ltd.), T. Powell (British Electro-Metallurgical Co., Ltd.), O. R. Smith (Aluminium Laboratories, Ltd.).

As Associates

E. Calvert (Richard Johnson and Nephew, Ltd.), R. Clark (Thomas Firth and John Brown, Ltd.), L. J. Derham (Imperial Smelting Corporation, Ltd.), D. A. Driffill (Hadfields, Ltd.), J. A. Garner (Steel, Peech and Tozer), W. H. L. Hooper (Imperial Chemical Industries, Ltd., Metals Division), E. L. James (John Garrington and Sons, Ltd.), R. Mercer (Enfield Rolling Mills, Ltd.), C. G. Nicholson (Thomas Firth and John Brown, Ltd.), N. W. Thomson (B.S.A. Tools, Ltd.), W. O. Turton (John Baker and Bessemer, Ltd.), K. G. Wood (Cumberland Technical College).

As Associates from the Grade of Licentiate

I. C. H. Hughes (The Mond Nickel Co., Ltd.), I. P. McDowall (Stewarts and Lloyds, Ltd.), S. D. Pitt (Bundaberg Foundry Co., Ltd.), P. J. Rowling (Powers Samas Accounting Machines, Ltd.), H. J. Sharp (Hoover, Ltd.), H. Simms (Appleby-Frodingham Steel Co.), J. S. Thompson (Durham Chemicals, Ltd.), R. E. H. Williams (Powell Duffryn Technical Services, Ltd.), L. A. Willsmere (Ministry of Fuel and Power).

As Licentiates

J. S. Ambler (Eastern Smelting Co., Ltd.), E. H. Bridges (Davy and United Roll Foundry, Ltd.), L. J. Holton (Acloque and Co.), A. W. Meadows (Dudley Drop Forging Co., Ltd.), G. W. Miles (Edgar Allen and Co., Ltd.), A. S. Narayanan (Tata Iron and Steel Co., Ltd.), K. A. Oliver (C. Akrill Co., Ltd.), D. A. Ryder (Joseph Lucas (Gas Turbines), Ltd.), P. Tharmalingam (Government of Ceylon), P. J. Wakeman (Hale and Hale, Ltd.).

International Congress on Industrial Chemistry

The Twenty-Fourth International Congress on Industrial Chemistry will be held in Paris from 25 November to 1 December 1951. The proceedings of the Congress will be divided in 19 sections, one of which will deal with metallurgy. An exhibition will take place simultaneously with the congress. Full particulars may be obtained from the Société de Chimie Industrielle, 28 rue Saint-Dominique, Paris (7e).

Symposium on Mineral Dressing, September 1952

Arrangements are now being made by the Council of the Institution of Mining and Metallurgy to hold a two-day Symposium on Mineral Dressing on 23 and 24 September 1952, at the Imperial College of Science and Technology, South Kensington, London, S.W.7.

The purpose of the Symposium will be to discuss recent developments (*a*) in fundamental concepts and experimental methods, (*b*) in practice at selected plants of particular interest, and (*c*) in plant design and machine manufacture.

Several papers are already in preparation, and others are being considered by prospective authors. Anyone who wishes to submit a paper is invited to send a synopsis of its contents and an indication of its probable length to the Secretary of the Institution, Salisbury House, London, E.C.2.

All papers accepted for the Symposium will be sent as preprints to those who propose to attend the two-day discussion, and a volume containing the papers and a report of the discussion will be published after the meeting.

OTHER NEWS

Research on Creep and Fatigue Properties of Metals at Elevated Temperatures

Shortly after the end of the war, the Light Alloys Sub-Committee of the Ministry of Aircraft Production Metallurgical Research Committee reviewed the position in the field of research on creep and fatigue properties of metals at elevated temperatures. Under the chairmanship of Professor L. Aitchison, a policy was framed for the guidance of workers employed on investigations for the Government Services on creep. A memorandum incorporating the Sub-Committee's recommendations on this subject was drafted in 1947. At the same time thought was given to corresponding investigations on fatigue.

When, in 1948, the Ministry's Research Committee was superseded by the Inter-Service Metallurgical Research Council, consideration of this subject was continued by the Non-Ferrous Metals Committee of that Council under the chairmanship of Mr. (now Professor) A. J. Murphy. In due course a further memorandum was produced, incorporating that Committee's recommendations on research on the fatigue of metals at elevated temperatures.

These memoranda were primarily intended to help in the formulation of Government research policy. No extensive series of researches was directly initiated to implement these various recommendations, which were regarded as providing a co-ordinating policy within which investigations in this field could, if attention were paid to the broad aspects suggested in the memoranda, be directed towards the provision of information of the greatest general benefit. Research into the mechanism of creep and fatigue was not regarded as coming within the scope of these recommendations.

Within the somewhat limited range of laboratories concerned, this policy has met with success. Investigators studying some particular application of metals or alloys at high temperatures have been pleased to broaden their work to bring it within the scope of the general recommendations made. For this reason the Non-Ferrous Metals Committee of the Inter-Service Metallurgical Research Council has recommended the issue of a general statement summarizing the views expressed and recommendations made in the earlier memoranda, with the object of giving an opportunity to all investigators working in this field to modify and direct their researches in such a way as to contribute the maximum amount of useful information on the behaviour of metals and alloys when subject to creep and fatigue conditions at high temperatures.

The approach to this matter by the Committees concerned has been a desire for the greatest amount of fundamental information on the behaviour of metals under these conditions. It is inevitable that many laboratories are primarily concerned with providing data on the creep and fatigue resistance of specific alloys. The values thus obtained are available to engineers, and may contribute in some measure to the general understanding of the phenomena concerned. It may happen, however, that those investigators who have made tests with a view to providing data are not in a position to extend them sufficiently to provide information of fundamental scientific value. There are, within the Committee's knowledge, certain researches in this field which could be so extended with little additional time and effort and whose value, therefore, in their contribution to our general knowledge, could be greatly enhanced.

This note is not published with the intention of attracting a number of new workers into this field. Admittedly, an enormous amount of further knowledge on these subjects is needed, but, in view of the heavy demands for research workers on other aspects of metallurgy, an intensive attack on creep and fatigue might absorb an effort that would be out of proportion to the probable immediate benefits. Publication, however, will have served its purpose if it encourages those who of necessity or by desire are working in this field to try to fit their investigations into the framework which is here described.

Factors Recommended for Consideration

It is suggested that the most important general approach in this matter should be an orderly investigation of the influence of variables in metallographic structure and in composition on creep and fatigue resistance at elevated temperatures. The following factors are considered to be important:

A. *Factors Relating to the Matrix :*

(i) Grain-size.

(ii) Orientation.

(iii) The effect of previous mechanical strain and of the manner in which this is effected.

(iv) Effect of solute metals present in alloying proportions in solid solution.

(v) Effect of small amounts of dissolved impurity.

(vi) Crystal structure of the matrix crystals (i.e. face-centred or body-centred cubic, hexagonal, &c.).

B. *Factors Relating to the Second, " Dispersed " Phase :*

(i) Distribution of particles (e.g. grain-boundary distribution or uniform distribution).

(ii) Particle size.

(iii) Amount of second phase present.

(iv) Nature of second phase present (whether " intermetallic compound " or " electron compound ", &c.—influence of crystal structure and other properties of second phase).

C. *Temperature-Dependent Factors :*

Relaxation towards equilibrium of non-equilibrium structures at the elevated service temperatures, recrystallization, grain growth, absorption or globularization of precipitates, &c., and the influence of stress on such structural changes.

Materials Recommended for Study

As will be obvious from the fact that this approach to the subject was suggested by the Light Alloys Sub-Committee, aluminium alloys were primarily in mind at the outset. It was, however, felt throughout all these discussions that the work should not be confined to aluminium alloys, but that a number of different solute and solvent metals must be considered if information of general application is to result. The Committee has accordingly made the following more specific recommendations.

Bearing in mind the large number of variables referred to above, it is clearly necessary to choose for examination systems in which there is a reasonable prospect of studying these factors separately and in which, therefore, such factors as alloy constitution are already well known. It is furthermore considered that the work should be carried out using pure materials so that the effects of impurities may later be investigated separately, and that in the initial stages of the work the materials should be in structural equilibrium at the temperature of the test before the start of the test. The information resulting as to the behaviour of the material under such controlled conditions will provide a background against which subsequent work on the behaviour of non-equilibrium structures is less difficult to interpret. The importance of a full and detailed description of the specimens tested and of their method of manufacture, including surface finish, cannot be over-emphasized. In some investigations already in progress or contemplated, it may only be possible to make a specific and accurate measurement of grain-size, composition, and any tendencies to preferred orientation in the specimens used and to record all available details of their history. The provision of such information gives a greatly enhanced value to the results of what might otherwise be an *ad hoc* creep or fatigue test on an ill-defined specimen.

(a) *Factors Relating to the Matrix*

For studies of the effect of factors associated with the matrix, the alloys chosen must remain reasonably homogeneous over a range of compositions and temperatures. In this section of the work investigations on pure metals and on single crystals had rather less encouragement than the investigation on alloy systems, except in so far as the study of the pure metal is necessary to compare its behaviour with that of a solid solution based thereon. While, in Section A, item (iv) is perhaps receiving the greatest emphasis in work now in progress, the other items should clearly receive simultaneous consideration, since the behaviour of a polycrystalline aggregate will be profoundly affected by grain-size, orientation, and other such matters.

Alloy systems particularly mentioned as worthy of consideration in this part of the work are the aluminium–zinc and aluminium–magnesium alloys and copper-base alloys in which the valency of the solute can readily be varied (the lattice distortions produced by a large number of elements in copper are accurately known, and can therefore be directly compared with mechanical properties). Magnesium alloys, such as those with cadmium or aluminium, provide materials based on a hexagonal metal, and iron, which can exist in face-centred cubic and body-centred cubic forms, provides an opportunity of correlating properties with changes in crystal structure.

(b) *Factors Relating to the Second Phase*

Systems in which the second phase is completely insoluble in the solid state are difficult to find, but some work is in progress in co-operating laboratories on the aluminium–iron and lead–copper systems, where this condition is approached.

The effects of the mode of occurrence, distribution, and particle size of a dispersed phase could be studied in the system aluminium–copper, about the precipitation processes in which much is already known. In this case a definite intermetallic compound of limited homogeneity range is formed. Research should preferably be planned to include the effects of different types of dispersed phase. Many systems of the "aluminium plus intermetallic compound" type are available, while in the system aluminium–magnesium there is an intermediate phase of relatively wide homogeneity range. It may be necessary at some stage of the work to extend investigations to the effect of second phases of the "electron compound" type, in order to complete the picture.

(c) *Temperature-Dependent Factors*

The effect of these variables could be examined using the aluminium–copper system or perhaps, better, in some system such as aluminium–magnesium, where the variation of solubility with temperature is more pronounced.

General Observations

It is not intended in this note to do more than indicate an approach to the problem and a general line of thought. Metallurgists will no doubt think of alloy systems other than those mentioned above, a study of which could profitably form a part of the general investigation here outlined. It is desired to emphasize only that a programme of this general type might be expected to give information with regard to the effects of several factors treated, as far as possible, separately. The success or otherwise of the correlation would tend to reveal whether factors other than those enumerated should be considered. More complex alloys, more nearly related to alloys at present in service, could then be examined and developed in the light of the knowledge gained.

It cannot be too strongly emphasized that there is no desire or intention in the minds of the Committee to direct researches on creep and fatigue of metals into any unduly narrow channel. The Committee's object is only to call the attention of all interested parties to the desirability of carrying out work in this field on such materials and in such a way, wherever possible, as to make the most effective contribution to our structure of knowledge on the subjects.

The Non-Ferrous Metals Committee would welcome information as to the particular researches of investigators working in this field. Those interested should communicate with the Secretary, Non-Ferrous Metals Committee, Inter-Service Metallurgical Research Council, Metallurgy Department, Royal Aircraft Establishment, Farnborough, Hants.

Post-Graduate Courses in Structures and Materials

A post-graduate course in Theory of Structures and Strength of Materials, extending over one year (October to June), will begin at the Department of Engineering, Cambridge University, on 8 October 1951.

The latest advances made in the understanding of the behaviour of metallic structures under static, repeated, or fatigue loading, will be the principal subject of the course, with particular emphasis on welding as a method of fabrication. No attempt will be made to teach conventional methods of design, but present-day practice and possible future developments will form the subject of critical study.

The object of the course is not to train research workers, but to help engineers to apply the latest advances in knowledge. The course will include lectures, colloquia, and laboratory work, and each student will be encouraged to make a detailed study of some problem of particular interest to him.

The course is open to University graduates with industrial experience, and also to students who have suitable equivalent qualifications.

Applications for admission should be sent to the Secretary, Faculty Board of Engineering, Engineering Laboratory, Cambridge, from whom further particulars may be obtained.

British Non-Ferrous Metals Federation

The following officers of the Federation have been appointed for the year 1951–52: *President*, Mr. W. J. Terry; *Past-Presidents*, Dr. Horace W. Clarke and Mr. W. H. Henman; *Vice-Presidents*, Mr. W. F. Brazener, Mr. H. C. Gibbins, and Mr. H. E. Jackson; *Treasurer*, Mr. A. L. Johnson; *Chairman of the Executive Committee*, Mr. W. J. Terry.

Members of Council elected for the year are:

Brass and Copper Tube Association: Mr. L. Endall, Mr. H. C. Gibbins, Mr. S. G. S. Heinrich, and Mr. W. R. D. Macdonald.

Cold-Rolled Brass and Copper Association: Mr. W. F. Brazener, Dr. Horace W. Clarke, Mr. A. Dando, and The Hon. John Grimston.

Extruded Brass and Copper Alloy Association: Mr. J. C. Colquhoun, Mr. G. Meredith, Mr. E. J. Overton, and Mr. T. A. M. Roberts.

High-Conductivity Copper Association: Mr. A. L. Johnson, Mr. M. A. T. Johnson, Mr. W. J. Terry, and Mr. F. Waine.

Manufactured Copper Association: Mr. C. E. Prosser and Mr. W. F. Slater.

Brass Wire Association: Mr. C. F. Aston and Mr. A. S. Hollings.

Zinc Rollers' Association: Mr. S. C. Hunn and Mr. E. N. Rowbotham.

Nickel Silver Association: Mr. F. Geo. Allen and Mr. W. H. Henman.

Non-Electrical Copper Association: Mr. G. E. Jewell and Mr. P. Shakespeare.

Condenser Plate Association: Mr. S. G. Homfray and Mr. H. E. Jackson.

Brazed Brass Tube Association: Mr. H. J. Bailey and Mr. G. A. D. Smith.

APPOINTMENTS VACANT

THE GRAIN REFINEMENT OF ALUMINIUM ALLOY CASTINGS BY ADDITIONS OF TITANIUM AND BORON *

By A. CIBULA,† M.A., A.I.M., MEMBER

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

SYNOPSIS

Previous work showed that the grain refinement produced by the addition of titanium or boron to aluminium casting alloys is primarily caused by nucleating particles in the melts; the nuclei in titanium-containing alloys were found to be titanium carbide crystals, though only a small proportion of the added titanium was present in this form. The main objects of the present work were to identify the nuclei in other fine-grained aluminium alloys and to find ways of increasing the proportion of the refining elements present as nucleating compounds.

By centrifuging the particles from molten alloys containing boron but no titanium and observing the change in grain-size produced, evidence was obtained that the nuclei in these alloys are aluminium boride crystals. The minimum boron addition for adequate refinement of these alloys therefore depends mainly on the solubility of aluminium boride in molten aluminium.

Attempts to increase the concentration of titanium carbide in alloys containing titanium achieved no useful results, owing to difficulties in forming or dispersing the carbide as fine particles. The addition of boron instead of carbon was more effective in producing refinement, nucleating particles of titanium boride being formed at very low concentrations of titanium and boron; moreover, boron was more easily added than carbon, as aluminium–boron master alloys could be used. The boride formed was isomorphous with pure TiB_2, but had slightly different lattice dimensions and a range of composition ($a_0 = 3{\cdot}010\text{–}3{\cdot}016$ Å., $c_0 = 3{\cdot}235\text{–}3{\cdot}240$ Å.).

The grain refinement of some commercial casting alloys by simultaneous additions of titanium and boron was studied in detail; grain coarsening due to high casting temperatures, repeated remelting, or gravity segregation during solidification was less than in alloys containing much larger percentages of titanium alone. Alternative methods of adding titanium and boron and the refinement produced by borides of transition metals other than titanium were investigated.

As the boron additions required when titanium was present were small, the mould reaction previously encountered in alloys of high boron content was largely avoided.

I.—INTRODUCTION

It has been shown in previous work [1] that the small grain-size produced by the addition of a grain-refining element to aluminium casting alloys is mainly caused by fine particles of a specific compound of this element, which are either introduced in the added hardener or formed by reaction with a constituent of the melt. This compound has certain properties which enable the particles to act as nuclei when the casting starts to solidify, and thus to produce a large number of grains.

The resulting refinement of the macrostructure greatly improves the properties of cast solid-solution-type alloys,[2, 4, 5, 6, 7] and one of the objects of the present work was to produce even finer grain-sizes, particularly when high casting temperatures have to be used. The refining element is used most effectively —and economically—when as much of it as possible is present as small particles of the nucleating compound, and the aim of most of the experiments described below was therefore to find ways of achieving this.

In some of these experiments it was first necessary to identify the nuclei. The earlier work [1] on grain refinement by the addition of boron, for example, did not establish the identity of the nuclei, although it was suggested that they were crystals of aluminium boride (AlB_2); further experiments, described in Section III, have therefore been made to confirm this suggestion.

The nuclei in alloys refined by the addition of titanium and tantalum were found [1] to be interstitial carbides of these elements, and indirect evidence indicated that the nuclei in alloys containing other transition metals were the similar carbides formed by these elements; attempts have therefore been made to increase the carbon content of some titanium hardener alloys and to make direct additions of titanium carbide to molten aluminium alloys. The results are summarized in Section IV.

The work described in Section III confirmed that crystals of aluminium boride could act as solidification nuclei. This suggested that the isomorphous borides formed by several of the transition metals might also

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* Manuscript received 3 February 1951. The work described in this paper was made available to members of the B.N.F.M.R.A. in a confidential research report issued in December 1950.

† Research Investigator, British Non-Ferrous Metals Research Association, London.

be suitable, particularly as these borides are similar to the interstitial carbides; it had already been reported to the Association * that simultaneous additions of titanium and boron, which might be expected to form titanium boride in the melt, refined the grain-size of aluminium alloys more effectively than the addition of either element alone. The grain-refining effects of additions of titanium and boron together were therefore studied in some detail, including the identification of nuclei and the examination of the effects of remelting, increase in casting temperature, and gravity segregation during solidification. The grain-refining effects of additions of boron together with several other transition elements were examined briefly. The results are given in Section V.

II.—EXPERIMENTAL TECHNIQUE

1. Materials

The compositions of the two grades of aluminium, the commercial casting alloys, and the hardener alloys, and the methods of preparation of the hardener alloys, are described in Table I.

Super-purity aluminium was used as the basis metal in all the experiments in which it was important to avoid the presence of impurities, including the preparation of the hardener alloys which were required. Most of the experiments on the refinement of alloys of commercial composition and purity were carried out with aluminium–4·8 or 4·9% copper alloy (B.S. 1490, LM 11), made from commercial-purity aluminium and the hardener alloys described in Table I. A few other experiments were made with the commercial casting alloys noted in the table.

The fluxes used in the experiments described in Section IV, 3, were made from chemicals of Analar grade.

2. Melting and Casting

The high-purity materials were melted in Salamander crucibles which were coated with pure alumina cement and, except in the preparation of hardeners, were heated in an electric furnace to minimize gas pick-up. These melts were stirred with pure alumina (ΔR) rods and were cast without being degassed. Bars of ¾ in. dia. for grain-size measurement were cast in green Mansfield sand, and 1-in.-dia. bars, from which specimens to be centrifuged were machined, were made in cast-iron moulds.

The melts of commercial-grade alloys, usually 3 kg. in weight, were heated in uncoated crucibles in a gas-fired furnace. They were degassed with chlorine as soon as the temperature reached 660°–720° C., the flow of gas being gradually increased to raise the temperature of the melt up to, but not above, the selected casting temperature, making use of the heat generated by the strongly exothermic reaction between the chlorine and the melt. A constant time of degassing was used in each series of alloys poured at a constant temperature for comparison of grain-sizes. The reduced-pressure gas-test was used to check that the melts were substantially gas-free. With the exceptions noted below, the commercial-purity alloys were cast into inclined D.T.D. bar moulds in green synthetic sand and were poured in 8–12 sec. per bar.

Melt temperatures were measured to ±5° C. with a Chromel/Alumel thermocouple.

3. Examination of Castings

Sections for grain-size determination were taken from the junction of the D.T.D. bars and the feeder-heads (before heat-treatment) and from the middle of the ¾-in.-dia. bars. The average grain diameters were measured with a rule, at a convenient magnification.

As previous work [4] had shown that boron promotes mould reaction in the aluminium–4·9% copper alloy, the distribution of porosity in the polished sections of the test-bars was noted and the overall percentage of voids in each bar calculated from its density, measured before heat-treatment. The density of the void-free alloy containing 4·90% copper was taken as 2·800 g./c.c. in the as-cast condition.

Some of the D.T.D. bars in the aluminium–copper alloy were solution-treated at 525°–535° C. for 16 hr. and aged at 140° C. for 16 hr. in a forced-air-circulation furnace, and the ultimate tensile strength and elongation were determined on 0·564-in.-dia. test-pieces machined from the bars.

4. Centrifuging Experiments

The use of the centrifuge developed by Baker for the identification of constituent particles in molten metals has been described previously.[1, 3, 8]

Cylindrical 12-g. specimens were melted in a closed graphite crucible which lay on its side in a horizontal tube-furnace mounted on the arm of the centrifuge. The melt was then rotated about a vertical axis with a centrifugal acceleration of 140 g. and was finally allowed to solidify with the centrifuge either rotating or stationary, as desired. The temperature of the metal could be measured to ±10° C. or better by means of a thermocouple passing into the lid of the crucible with the weld close to the surface of the specimen.

III.—THE IDENTIFICATION OF NUCLEI IN ALLOYS REFINED BY THE ADDITION OF BORON

1. Previous Work

In earlier work [1] on the addition of boron to pure aluminium, it was observed that 0·01% boron was just sufficient to diminish undercooling before solidification and slightly to reduce the diameter of the

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* This information was originally given in 1948 by Dr. Strauss of Messrs. Foundry Services, Ltd., and has since been confirmed by other members of the British Non-Ferrous Metals Research Association.

TABLE I.—*Composition and Preparation of Alloys Used.*

Alloy	Mark	Alloy Content, %	Impurities, %	Method of Preparation	Section of Report Describing Experiments in Which the Alloy is Used
			Aluminium Metal		
S.p. Al	...	<0·01 total	0·001 Ti	...	III; IV, 3; V, 2.
C.p. Al	...	<0·3 „	0·11 Si; 0·16 Fe; 0·002 Ti	...	IV, 2 and 4; V, 3 and 4.
			Hardener Alloys		
Al–Cu	NOU 501	50 Cu	...	Cathode Cu (99·99%) dissolved in molten s.p. Al.	III; IV; V.
Al–Ti	NKY 25	1·56 Ti	0·21 Si; 0·01 Fe; traces Mg, Mn, Cu, Zn.	Potassium titanofluoride stirred into molten s.p. Al in graphite crucible at 1100° C.	III; V, 3.
	NKY 21	1·81 Ti	...	As NKY 25, but prepared at 1300°–1400° C.	IV, 2.
	NKY 22	1·20 Ti	...	As NKY 25, but prepared at 1600° C. (approx.).	
	OJF 1	2·66 Ti	...	Commercial hardener, as received.	
	OJF 2	2·22 Ti	...	Commercial hardener (OJF 1) remelted to 1100° C. under cover of KF + graphite.	
Al–B	NLC 7	0·32 B	0·08 Si; 0·01 Fe; <0·002 Ti.	Potassium borofluoride stirred into molten s.p. Al at 1100° C. in graphite crucible.	III.
	NLC 10	0·50 B	0·16 Si; <0·01 Fe; <0·01 Ti; traces Mg, Cu, Zn.		V, 2 and 4.
Al–Ti–B	NKY 27	1·3 Ti; 0·7–1·5 B	...	C.p. Al and commercial Al–2·7% Ti hardener melted to 1100° C. and commercial Al–25% B alloy added through cryolite flux. Cast at c. 1400° C.	V, 2.
	NKY 28	0·26 Ti; 0·16 B	...	As NKY 27 but commercial Al–10% Ti hardener added after boron at c. 1500° C., and alloy poured immediately at c. 1100° C.	
Al–V	OGQ 1	4·9 V	0·7 Fe; 0·23 Si; 0·06 Cu; 0·01 Ti; traces Mg, Mn, Ni, Zn.	Commercial Al–50% V alloy dissolved in molten Al under cryolite at 1100° C.	V, 4.
Al–Zr	NSK 10	1·75 Zr	...	Commercial Al–24% Zr alloy dissolved in molten Al at 1100° C.	
Al–Nb	NLA 6	0·21 Nb	...	Commercial Al–17% Nb + Ta alloy dissolved in molten Al at 1150° C.	
Al–Cr	NSH 3	6·95 Cr	...	Electrolytic Cr dissolved in molten Al at 1000° C.	
			Commercial Casting Alloys		
B.S. 1490, LM 11	...	4·8 Cu; 0·13 Ti	...	...	V, 3.
B.S. 1490, LM 10	...	10·6 Mg	0·001 Ti	...	
B.S. 1490, LM 4	...	3·8 Cu; 4·0 Si; 0·6 Fe; 0·6 Mn; 0·08 Ti	...	...	

C.p. Al: commercial-purity aluminium; S.p. Al: super-purity aluminium.

columnar grains of the cast alloys, indicating that a very small number of nuclei were present at this boron content; larger additions produced more marked refinement and prevented any detectable undercooling. These effects approximately coincided with the appearance of aluminium boride (AlB_2) crystals which were observed in alloys containing 0·02% or more boron, in agreement with recent data on the aluminium–boron system.[9]

The aluminium atoms in the boride lie on a simple hexagonal lattice,[10] in the close-packed planes of which the interatomic distance (3·00 Å.) is similar to the corresponding distance in solid aluminium (2·86 Å.). This similarity and the above observations suggested that aluminium boride crystals were responsible for the observed grain refinement, nucleation of the aluminium-rich solid solution being accomplished by the deposition of the close-packed planes of aluminium atoms on to corresponding planes in the boride lattice.

It was also observed [1,4] that little or no grain coarsening occurred in alloys containing boron when the melts were superheated to temperatures up to 960° C. before casting, whereas alloys containing titanium showed considerable grain coarsening after superheating. This observation also agreed with the suggestion that aluminium boride crystals function as nuclei. In the alloys containing titanium, the nuclei are particles of titanium carbide, both elements of which are minor constituents of the alloy; when the melt is superheated, the nuclei dissolve, but they do not easily re-precipitate on cooling, presumably because both the carbon and the titanium are greatly diluted and the nucleation of titanium carbide crystals is therefore difficult. The aluminium boride crystals must also dissolve when the melt is heated to the aluminium–boron liquidus temperature, but, as aluminium is the major constituent of the alloy, the formation of aluminium boride crystals should be less liable to suppression on cooling,* and they should therefore re-precipitate when the casting cools and thereafter nucleate the aluminium-rich dendrites; the loss of refinement after superheating should therefore be less than in alloys containing titanium. (Nevertheless, some supersaturation occurs even with aluminium boride, as described in the Appendix.)

The absence of grain coarsening in alloys containing boron would be a useful property in commercial casting alloys, but, unfortunately, the addition of 0·03–0·05% boron required for grain refinement was also found to increase gas pick-up on melting and to produce mould reaction in sand castings of aluminium–copper [4] and aluminium–magnesium–beryllium alloys.[11] The following experiments were therefore made to confirm the identity of the nuclei in alloys containing boron, so that it could be decided whether the amount of boron required for refinement might be reduced and the undesirable effects of boron additions thus minimized.

2. PRIMARY PARTICLES IN AN ALUMINIUM–BORON ALLOY

In order to identify particles other than aluminium boride crystals, which might form the nuclei in alloys containing boron, attempts were made to concentrate the primary particles in a molten aluminium–boron alloy by centrifuging, and to identify the segregate by X-ray examination; a similar technique had been used successfully with alloys containing titanium.[1]

A specimen of high-purity aluminium–0·04% boron alloy was centrifuged at 680° C. for 15 min.; a section through the surface containing the segregated particles is shown in Fig. 1 (Plate I). All the lines in the diffraction pattern obtained by examination of the segregate with glancing-angle X-rays were identified as those of aluminium or aluminium boride (AlB_2). Numerous crystals of this compound, often hexagonal in shape, could be seen in the section through the segregate, but the only other constituent which could be distinguished was occasional plates very similar to those identified as aluminium carbide in other specimens. The latter compound has been observed in several boron-free alloys prepared from hardeners made under an active flux in graphite crucibles, and is unlikely to have any significant connection with the grain-refining effects of boron.

The above observations are consistent with the view that the solidification nuclei in the specimen were the crystals of aluminium boride; however, the possibility cannot be excluded that the nuclei were crystals of another compound which were very similar to the boride in appearance and too few to be detected by X-ray examination. The relation between grain-size and the presence of aluminium boride particles was studied further in the experiments described next.

The grain-refining effects of boron in aluminium are most marked when another element such as copper is present which produces concentration gradients during solidification and, by hindering crystal growth, allows the maximum number of nuclei to become effective in forming new crystallites. The experiments described in the remainder of this section were therefore carried out with alloys containing large percentages of copper to obtain the maximum range of grain-sizes; in the above experiments with binary aluminium–boron alloys, the detection of constituents was facilitated by the absence of $CuAl_2$, but the increase in grain-size caused by centrifuging was small.

3. THE REMOVAL OF ALUMINIUM BORIDE CRYSTALS FROM ALUMINIUM–COPPER–BORON ALLOYS BY CENTRIFUGING

(a) *The Influence of Temperature*

The removal of the nuclei from a molten alloy by centrifuging should result in a coarse grain-size after solidification. If the nuclei are aluminium boride crystals, it should be possible to produce grain coarsening only by centrifuging at a temperature well below the aluminium boride liquidus temperature, for otherwise most or all of the boride would be in solution and could not be segregated at the bottom of the crucible. On the other hand, if the nuclei are not aluminium boride but are some other compound, grain coarsening would result from centrifuging even above the aluminium boride liquidus temperature if the other compound were not in solution at that temperature. Experiments were therefore made to determine whether the removal of aluminium boride crystals by centrifuging an aluminium–5% copper–0·04% boron alloy at a low temperature resulted in grain coarsening and whether grain coarsening was

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* The chance of a molecule of AlB_2 forming in the melt is proportional to the square of the concentration of boron atoms (the concentration of aluminium atoms being practically unity); similarly, the chance of a molecule of TiC forming is proportional to the product of the concentrations of titanium and carbon atoms. The latter product is much smaller than the former at the usual concentrations of these elements and the compound TiC is therefore probably more likely to supersaturate. This simple comparison of the tendency of the two compounds to supersaturate is only very approximate as the formation of the smallest stable crystal nuclei is dependent on several other factors (which cannot be calculated at present).

produced even at higher temperatures at which no segregation of aluminium boride occurred.

Two specimens of this alloy were centrifuged for 15 min., at 655° C. (only 10° C. above the aluminium–5% copper liquidus temperature) or 760° C., and were allowed to solidify with the machine stationary; two other specimens were merely heated in the stationary centrifuge for 15 min., one at each temperature, to obtain the grain-sizes of the non-centrifuged alloy for comparison.

As anticipated, a segregate of aluminium boride crystals (shown in Fig. 9, Plate II) was obtained only in the specimen centrifuged at the lower temperature; in the other centrifuged specimen the crystals were observed over the entire cross-section where they had re-precipitated only after centrifuging had stopped. Corresponding effects were observed in the macrostructures of the specimens: an increase in grain-size occurred after centrifuging at 655° C. (as may be seen by comparison of Figs. 5 and 6, Plate I), whereas the two specimens treated at 760° C. had identical grain-sizes (slightly larger than that shown in Fig. 6, Plate I).

The largest grain-sizes that could be obtained in specimens cooled from each of these two temperatures, if it were possible completely to remove the boron-containing nuclei in each case, should be the same as the grain-sizes of boron-free aluminium–5% copper specimens melted under the same conditions. The macrostructures of two specimens of the boron-free alloy, heated at 655° and 760° C., respectively, are shown in Figs. 7 and 8 (Plates I and II). Comparison of Figs. 7 and 5 (Plate I) shows that the grain coarsening which occurred in the specimen centrifuged at 655° C. was almost the maximum that could be produced, in contrast with the great difference in grain-size between the boron-free specimen heated at 760° C. (Fig. 8, Plate II), and the boron-containing alloy centrifuged at the same temperature.

These results confirm that grain coarsening occurs after centrifuging an aluminium–copper–boron alloy only if most of the aluminium boride crystals are thus removed from the molten alloy.

(b) *The Influence of Alloy Composition*

The densities of the aluminium boride crystals and the primary dendrites of solid solution in the aluminium–5% copper–0·04% boron alloy (3·2 and 2·56 g./c.c., respectively), are greater than that of the liquid metal (approximately 2·48 g./c.c.) during the first stages of solidification. When a specimen of this alloy is melted and is then allowed to solidify under centrifugal force, the boride crystals segregate to the base of the crucible (i.e. away from the axis of rotation) as the temperature falls and dendrites of solid solution then start to grow into the liquid metal from the same surface. Only the first dendrites are formed close to AlB_2 crystals, and, whether these act as nuclei or not, the resulting macrostructure should be coarse grained. This is confirmed by the specimen shown in Fig. 10 (Plate II).

If the copper content is increased to 20%, however, the density of the primary solid-solution dendrites (2·62 g./c.c.) becomes lower than that of the copper-rich liquid (2·8–3·2 g./c.c.) throughout solidification, whereas the aluminium boride crystals are still denser than the liquid metal. If this alloy had solidified under centrifugal force, the first dendrites of solid solution formed at the face containing the segregated aluminium boride crystals would move towards the opposite face of the specimen (i.e. towards the axis of rotation), and this should be repeated with successive dendrites, all of which could thus be formed in the presence of aluminium boride crystals; consequently, if the latter function as nuclei the resulting grain-size should be much finer than in a boron-free alloy of the same copper content, in spite of the initial segregation of the boride and in contrast with the 5% copper alloy.

To confirm this conclusion, a specimen of aluminium–20% copper–0·04% boron alloy was centrifuged for 12 min. at 620° C. and solidified under centrifugal force; a boron-free alloy was similarly treated.

The macro-section of the boron-free specimen consisted mainly of large columnar dendrites, as shown in Fig. 11 (Plate II) (from the centre of the specimen) and Fig. 12 (Plate II) (from the base). The latter shows a dense band of dendrite-free eutectic at the base of the specimen.

The boron-containing alloy, however, consisted almost entirely of small equiaxial dendrites, as shown in Fig. 13 (Plate II) (from the centre of the specimen); the only large dendrites grew from the sides of the specimen near the base (they are visible at the right of Fig. 14, Plate II). A thin, discontinuous layer of segregated aluminium boride crystals was observed at the base of the specimen, but other crystals were scattered over the remainder of the specimen in the primary dendrites. Small dendrites, which had become trapped in the band of solid eutectic, were observed growing from the patches of boride crystals, Fig. 14 (Plate II), but none were observed at the corresponding face of the boron-free alloy, suggesting that the formation of primary dendrites was facilitated by the boride crystals and proceeded to a later stage of solidification.

4. Further Discussion

The results obtained above are all in agreement with the suggestion that aluminium boride crystals are the nuclei in alloys containing boron. The only other explanation is that the nuclei are another compound of boron with a constituent of high-purity aluminium, which dissolve in molten aluminium over approximately the same range of temperature as aluminium boride and re-precipitate on cooling, but no evidence of the existence of such a compound has been obtained by visual or X-ray examination. For this reason and others which were discussed above, the second explanation is unlikely, and it seems reasonable to conclude that the nucleating particles in aluminium–boron alloys are aluminium boride crystals.

The maximum proportion of the added boron that can be present as aluminium boride is determined by the aluminium–boron equilibrium diagram; it is, therefore, not possible to prevent undesirable mould reaction effects by reducing the amount of boron unless a third element can be found which, when present in small amounts, forms a similar boride at lower boron contents. This possibility is discussed further in Section V.

IV.—THE EFFICACY OF GRAIN REFINEMENT BY ADDITIONS OF TITANIUM

1. Previous Work

The results of previous experiments showed that the nuclei in fine-grained alloys containing titanium are particles of titanium carbide.[1] The amount of titanium present as titanium carbide was estimated to be only 0·001–0·01%, i.e. only 1–10% of the added titanium, in an alloy containing 0·1% titanium; it has also been estimated that of the remaining 0·09% titanium only about 0·003% is required in solution in the molten aluminium and in equilibrium with undissolved carbide, if excess carbon is present. This suggests that, if it were possible to prepare titanium hardener alloys containing greater proportions of carbon, smaller additions of titanium, down to 0·01% or less, might be sufficient to produce maximum grain refinement, provided that the alloy contained amounts of other alloying elements sufficient to produce the necessary concentration gradients round growing crystals of the aluminium-rich solid solution.

Micro-examination of the alloys described above showed that some of the titanium carbide crystals were as large as 5 μ and had dendritic structures; this indicated that a larger carbon content in a titanium hardener alloy might increase the average size of the carbide particles rather than increase their number, and the amount of titanium required for refinement might therefore not be much reduced. Nevertheless, attempts were made to prepare titanium alloys of high carbon contents, and experiments on the direct addition of titanium carbide to aluminium alloy melts were also carried out.

In other work,[4] it was found that castings prepared from commercial alloy ingot already containing titanium had much larger grain-sizes than castings made from melts of the same composition which had been prepared from a laboratory titanium hardener. A possible reason for the smaller number of nuclei in the commercial alloy was that the titanium hardener used in its preparation had a lower carbon content than the laboratory hardener, which had deliberately been made under conditions which facilitated carbon pick-up. An alternative explanation was that the coarse grain-sizes of the commercial alloy were the result of remelting, flux treatments, flushing with an inert gas, or holding the molten alloy for long periods at a low temperature during manufacture, all of which have been shown [1, 4, 13] to produce a reduction in the number of nuclei. To determine which was the correct explanation, the effectiveness of a commercial aluminium–titanium hardener alloy was compared with that of the laboratory hardener previously used, with the result described below.

2. The Addition of Carbon to Titanium Hardener Alloys

(a) *Commercial Titanium Hardener*

A commercial titanium hardener containing 2·66% titanium was treated with graphite under a cryolite flux at approximately 1100° C. The effectiveness of the treated and as-received alloys was compared with that of the laboratory hardener previously used [4] by measuring the grain-sizes of test-bars in an aluminium–4·8% copper–0·12% titanium alloy prepared from each hardener and cast at a series of

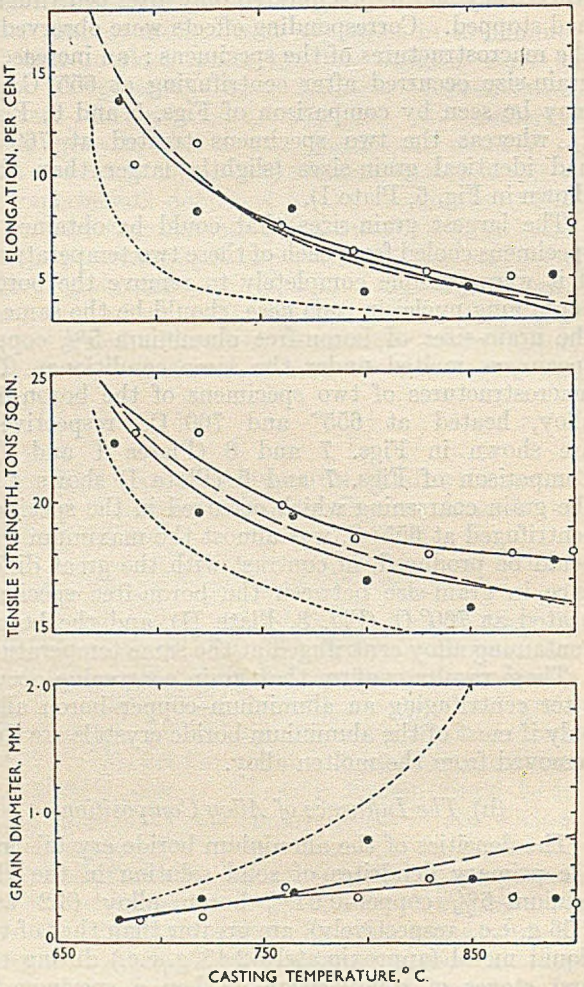

Fig. 15.—Comparison of Grain-Sizes and Tensile Properties of Test-Bars in Al–4·8% Cu–0·12% Ti Alloy, prepared from (●——●) as-received and (○——○) carbon-treated commercial titanium hardeners, (— — —) a laboratory titanium hardener, and (- - - - -) commercial B.S. 1490, LM 11 alloy ingot.

temperatures between 680° and 900° C. The increase in grain-size with pouring temperature, caused by solution of the nuclei, should be least in castings containing the greatest proportion of titanium in the form of carbide, i.e. in the alloys made from the most effective titanium hardener alloy.

The results of grain-size, ultimate tensile strength, and elongation measurements are shown in Fig. 15; the results previously obtained with the laboratory titanium hardener and those obtained with commercial ingot already containing titanium have been included for comparison.

The results obtained from the carbon-treated and the as-received commercial hardener and the laboratory-made hardener were much the same, but the castings made with each of the three hardeners had considerably finer grain-sizes, with correspondingly higher tensile properties, than castings made from the commercial B.S. 1490, LM 11 alloy ingot. The scatter in the results obtained with the as-received commercial hardener was large; this was probably caused by segregation in the hardener which resulted in a variation in titanium content in the castings.

The commercial hardener was therefore as effective as the laboratory-made hardener and, assuming the former to be typical of commercial hardeners, the coarse grain-sizes of commercial aluminium–4·5% copper ingot already containing titanium are not caused by the use of an ineffective hardener in its preparation; the loss of refinement is probably, therefore, the result of the manufacturing process of the ingot.

The latter conclusion is supported by the even larger grain-sizes of castings made from some batches of secondary alloys; for example, test-bars made from a batch of B.S. 1490, LM 4 alloy containing 0·08% titanium had a grain-size of 2·5 mm. when poured at 700° C., compared with approximately 0·25 mm. which can be obtained by adding the same amount of titanium in a hardener alloy.

(b) *Aluminium–Titanium Hardener Alloys Prepared at High Temperatures*

The solution of the nuclei caused by superheating aluminium–titanium alloys indicates that the solubility of carbon in these alloys increases with temperature. Titanium hardener alloys were therefore prepared at higher temperatures than are normally used, in an attempt to saturate the alloys with carbon at these temperatures and obtain higher carbide contents.

The hardeners were made from potassium titanofluoride at approximately 1350° and 1600° C., respectively, graphite being added to the flux. To prevent the formation of larger titanium carbide crystals, the alloy prepared at the highest temperature was granulated in water instead of being cast into ingot moulds. The flux cover was very fluid at 1100° C., vigorously attacking the crucible and pieces of graphite, but at 1600° C. it formed a solid crust which apparently merely coated the graphite, and the advantage of the high temperature of preparation was thus partly lost.

These hardeners were compared with one made at 1100° C. by casting series of test-bars as described above. The results obtained with the hardener made at 1350° C. showed no improvement; considerably inferior results were obtained with the hardener prepared at 1600° C.

It is possible that the use of more stable fluxes which remain fluid at high temperatures might facilitate the introduction of carbon and yield more successful results, but the prospect was not sufficiently promising to warrant further work along these lines.

3. DIRECT ADDITIONS OF TITANIUM CARBIDE

Titanium carbide powder of particle size 1 μ (the finest available commercially) was mixed with 120-mesh aluminium powder and pressed into pellets containing 10% by weight of carbide. The pellets were then stirred into 600-g. melts of high-purity aluminium–5% copper alloy, from which bars for grain-size measurements were cast. A high casting temperature of 850° C. was used to produce a wide range of grain-sizes for ease of comparison.

No addition could be made without a flux, as the carbide merely oxidized and mixed with dross. Thin flux covers, of the compositions noted in Table II, were used in subsequent experiments, the results of which are included in the table.

The analysis shows that only 10% or less of the added titanium actually entered the melt and, although these small amounts were sufficient to

TABLE II.—*Additions of Titanium Carbide to Aluminium–5% Copper Alloy Cast at 850° C.*

Mark	Method of Addition	Flux	Titanium Addition		Grain Dia., mm.
			Nominal, %	By Analysis, %	
[Previous work (1-in.-dia. bars)]	Al–Ti hardeners	None	0·12	0·12	0·5–0·7
	Commercial ingot	None	...	0·12	1·9
NOU 150A	Titanium carbide pellets	None	None	...	3·5
B, D		46% KF, 54% AlF_3 (m.p. 570° C.)	None	...	1·5; 2·5
C, E			0·12	<0·01; 0·015	1·0; 0·9
F			0·20	0·010	0·75
G			0·30	0·025	0·5
H		52% NaF, 48% KF (m.p. 700° C.)	None	...	1·5
J			0·12	0·01	1·2
K			0·30	0·025	0·7

produce a marked reduction in grain-size, equal to that obtained by adding 0·12% titanium as a hardener alloy, the efficiency in the use of the added titanium was less than when a hardener is used.

In other experiments, attempts were made to reduce the amount of flux required and thus the proportion of carbide trapped in the flux. The carbide powder was stirred into small pellets of molten flux which, when solid, were plunged into molten aluminium–copper alloy. The results showed no significant improvement.

4. Further Discussion

Although the results obtained above with titanium carbide additions confirmed that it is possible to produce grain refinement with much smaller titanium additions than are normally used, no method of practical value was found. The results showed that the obstacles in increasing the carbon content of aluminium–titanium alloys are largely caused by the difficulty of achieving intimate contact and wetting between carbon or titanium carbide and molten aluminium, either due to interference by oxide films or to inherently unsuitable angles of wetting. It has been suggested that one way of avoiding the difficulty would be by pre-wetting titanium carbide powder by sintering with nickel or cobalt powder, but the high melting point of these metals would be inconvenient with aluminium alloys and bridging between carbide particles might prevent their complete dispersion.

The introduction of carbon into molten aluminium–titanium alloys is also limited by the low solubility of carbon in the melt, for any excess of carbide would tend to remain where it was formed, in contact with the source of carbon, instead of dispersing in the melt, unless the carbide could be precipitated in the liquid metal.

In the work described in the next section on the use of titanium boride instead of titanium carbide, the difficulties described above were overcome by using separate aluminium–titanium and aluminium–boron hardener alloys; by this means it was possible to precipitate the boride particles in the melt and control the excess of either constituent. This could not be done with titanium carbide additions because carbon cannot be added already alloyed with aluminium.

V.—GRAIN REFINEMENT BY SIMULTANEOUS ADDITIONS OF TITANIUM AND BORON

1. The Transition-Metal Borides

Several transition metals form diborides which are isomorphous with aluminium boride (AlB_2) and have very similar lattice dimensions[14, 15, 16] as shown in Table III. These compounds have strongly metallic properties and a typical interstitial structure,* and are similar to the interstitial compounds formed by carbon, nitrogen, and hydrogen with certain of the transition metals.

Table III.—*Lattice Parameters of Some Diborides with AlB_2-Type Hexagonal Structures.*[14, 15, 16]

Compound	a_0, Å.	c_0, Å.	c/a
AlB_2	3·00	3·25	1·08(3)
TiB_2	3·028	3·228	1·064
ZrB_2	3·170	3·533	1·114
NbB_2	3·086	3·306	1·071
TaB_2	3·088	3·318	1·074
VB_2	2·998	3·057	1·020
CrB_2	2·969	3·066	1·03

The similarity of the diborides to aluminium boride and to the interstitial carbides, all of which probably form nuclei in aluminium alloy melts, suggests that the diborides might function in the same way. This would be of practical value, however, only if the concentrations of boron required to precipitate the transition metal diborides in molten aluminium were less than the concentration necessary to form aluminium boride, for otherwise the latter would form in preference to the former. Unfortunately, no data have been published on the aluminium–transition metal–boron ternary systems or on the free energies or heats of formation of the borides, which would enable a comparison of the respective concentrations to be made. However, the information that the addition of boron and titanium together produced greater refinement than either element alone, suggested that crystals of titanium diboride did, in fact, form in the melt and act as nuclei during solidification.

If this were so, the addition of boron to alloys containing titanium would increase the proportion of titanium present in an effective form (either carbide or boride) and make it possible to reduce the amount of the metal required for refinement. The effects would thus be similar to those aimed at in the experiments on the addition of carbon described in the previous section, but the addition of boron is easier to carry out, by using, for example, an aluminium–boron hardener alloy. The objectionable effects of boron additions—the increase in the gas pick-up during melting and the promotion of mould reaction after casting—would also be reduced by the presence of titanium if smaller boron additions were required; this point is examined in more detail below.

It has been assumed that if the formation of a transition metal boride can occur in molten alu-

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* The borides differ, however, from the true interstitial compounds, in which the non-metallic atoms are isolated from each other in the interstices of a lattice of metal atoms, in that the boron atoms themselves form a network with fairly strong bonds.[16] For this reason the metal atoms in several of the diborides are more widely spaced than in the pure metals, and several transition metals can thus form interstitial diborides even though the interstices in the metal lattices are not quite large enough to accommodate the isolated boron atom; the formation of an interstitial carbide, on the other hand, is closely limited by the relative sizes of the carbon atom and the transition metal atom, in accordance with Hägg's rule.

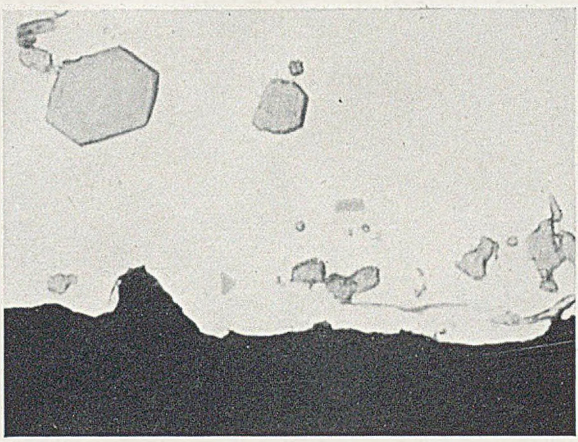

FIG. 1.—Segregated AlB_2 Crystals in Centrifuged Al–0·04% B Alloy. × 1000.

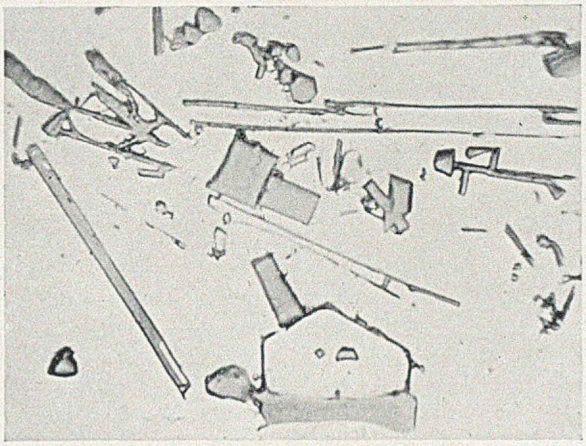

FIG. 2.—Segregated Primary Crystals in Centrifuged Al–0·04% B–0·12% Ti Alloy. × 2000.

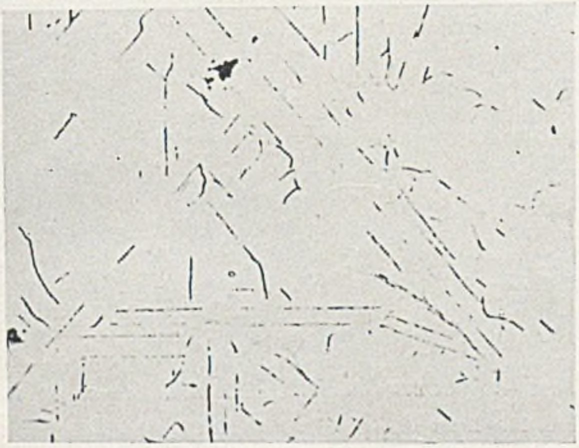

FIG. 3.—Precipitated Constituent in Interior of Centrifuged Al–0·04% B Alloy, after Ageing for 12 months at Room Temperature. × 150.

FIG. 4.—Same as Fig. 3. × 1000.

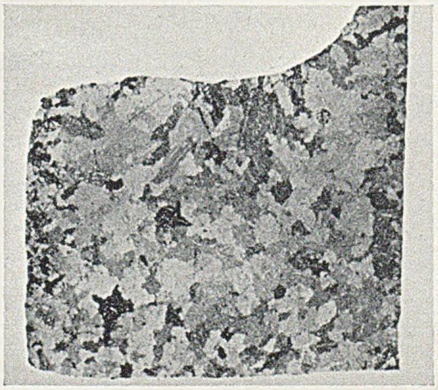

FIG. 5.—Al–5% Cu–0·04% B Alloy, Centrifuged for 15 min. at 655° C. × 3.

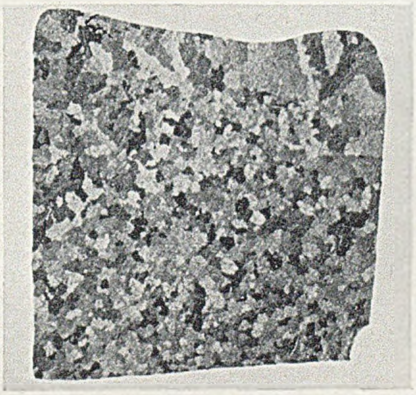

FIG. 6.—Al–5% Cu–0·04% B Alloy, Heated for 15 min. at 655° C. in Stationary Centrifuge. × 3.

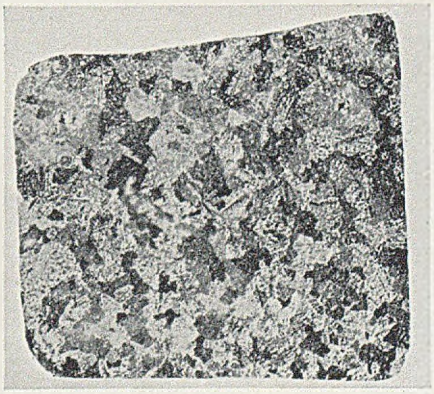

FIG. 7.—Al–5% Cu Alloy (boron-free), Heated for 15 min. at 655° C. in Stationary Centrifuge. × 3.

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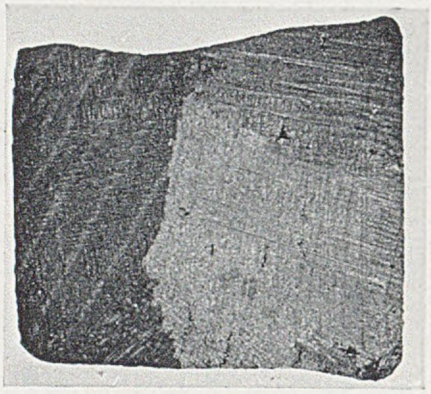

FIG. 8.—Al–5% Cu Alloy (boron-free), Heated for 15 min. at 760° C. in Stationary Centrifuge. × 3.

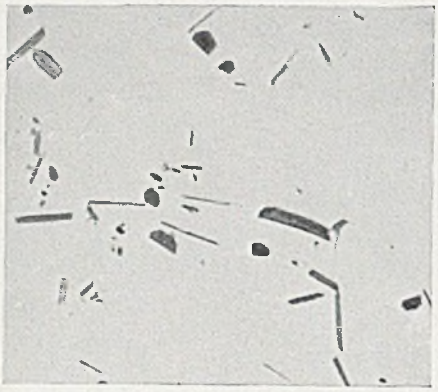

FIG. 9.—Segregate of AlB_2 Crystals at Right-Hand Edge of Specimen shown in Fig. 5. × 1000.

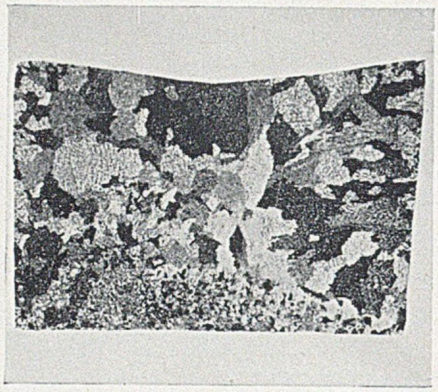

FIG. 10.—Al–5% Cu–0·04% B Alloy, Heated for 12 min. at 660° C. and Solidified under Centrifugal Force. × 3.

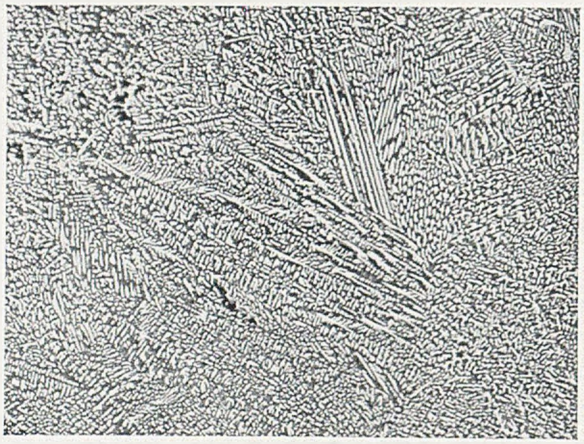

FIG. 11.—Large Dendrites in Interior of Al–20% Cu Alloy (boron-free), Solidified under Centrifugal Force. Etched in 0·5% HF. × 15.

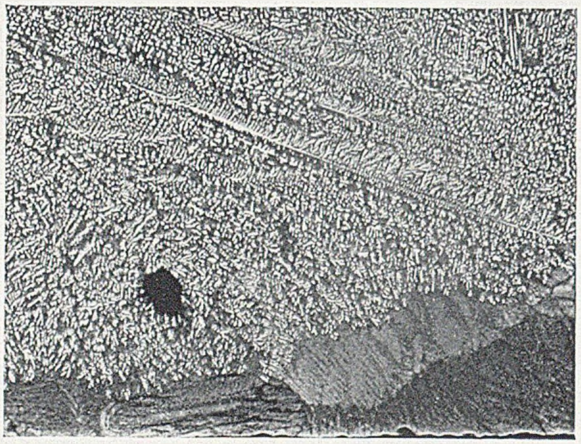

FIG. 12.—Large Dendrites near Base of Specimen Shown in Fig. 11. Etched in 0·5% HF. × 15.

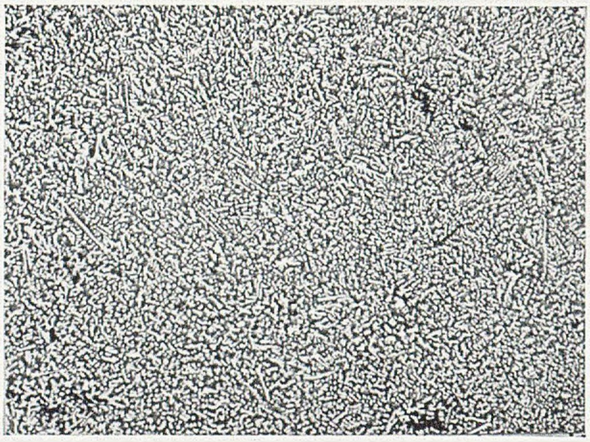

FIG. 13.—Small Dendrites in Interior of Al–20% Cu–0·04% B Alloy, Solidified under Centrifugal Force. Etched in 0·5% HF. × 15.

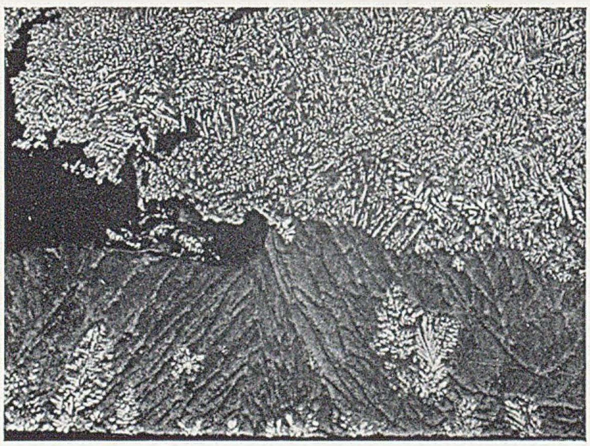

FIG. 14.—Small Dendrites Forming at Base of Specimen Shown in Fig. 13, from Segregated AlB_2 Crystals. Etched in 0·5% HF. × 15.

minium, the diborides described in Table III will be precipitated. Some of the transition metals, however, form other borides more complex than those described in the table, and the possibility that one of these complex borides may be formed, perhaps only over certain concentration ranges, must be remembered.

2. NUCLEI IN ALUMINIUM–TITANIUM–BORON ALLOYS

(a) *Addition of Boron to Coarse-Grained Aluminium–0·14% Titanium Alloy*

In order to study the grain-refining effects of additions of boron to an aluminium–titanium alloy, nuclei of titanium carbide which were already present were first removed by bubbling nitrogen through the molten alloy under a cover of magnesium chloride/potassium chloride flux, as described previously.[1] A 4-kg. batch of coarse-grained aluminium–0·14% titanium alloy was thus prepared.

Melts of 300 g. of this alloy were then heated to 760°–770° C., at which temperature additions of the aluminium–0·5% boron hardener were made, and the grain-size bar was immediately cast at 760° C. Super-purity aluminium was substituted for the aluminium–titanium alloy in one melt to confirm the absence of refinement by 0·01% boron (the largest addition used) when only 0·001% titanium was present; 0·2% copper was also added to this melt to provide small concentration gradients during solidification similar to those caused by the presence of titanium in the other alloys.

It was also desirable to confirm that the additions of the boron hardener did not re-introduce significant quantities of carbon into the titanium alloy. Unfortunately the carbon content of the hardener was not known, but an indication of the importance of this factor was obtained by making an addition of commercial-purity aluminium equal in weight to the largest addition of boron hardener alloy.

TABLE IV.—*Effect of Boron Additions on Grain-Size of Aluminium–Titanium Alloys Cast at 760° C.*

Mark, NKZ	Titanium Content of Melt, %	Additions, %	Grain Dia., mm.
136	0·14	None	Columnar grains
137		0·0001 B	0·8
135		0·001 B	0·23
131		0·01 B	0·23
132		2 c.p. Al	1·5 + columnar grains
133	approx. 0·001	0·01 B, 0·2 Cu	1 + columnar grains

The grain-sizes of the test-bars given in Table IV show clearly the marked effect of very small additions of boron, and confirm that the concentration of boron required to precipitate titanium boride particles is less than is required to form aluminium boride (0·0001% compared with approximately 0·01%).

The precise concentration of boron required probably varies with the titanium content, according to a reaction such as the following:

$$\underset{\text{aluminium)}}{\text{Ti (solution in}} + \underset{\text{aluminium)}}{\text{2B (solution in}} \rightleftharpoons \text{TiB}_2 \text{ (solid)}$$

in which the presence of excess titanium reduces the concentration of boron in solution in equilibrium with titanium boride.

The boron addition required for maximum refinement (approximately 0·0001–0·001%) is of the same order as the estimated carbon content of a fine-grained aluminium–0·1% titanium alloy.[1]

(b) *Primary Particles in an Aluminium–0·12% Titanium–0·04% Boron Alloy*

An aluminium–0·12% titanium–0·04% boron alloy was centrifuged at 680° C. for 15 min. to concentrate the primary particles. The surface at which the particles had segregated was then examined with glancing-angle X-rays and sectioned for micro-examination; scrapings from the surface were examined in an X-ray powder-camera for the determination of lattice parameters.

The diffraction pattern obtained was of an AlB_2-type structure, but the lattice parameters deduced from the pattern were not those of TiB_2 or AlB_2 but were of intermediate dimensions. Furthermore, the lines in the diffraction pattern were diffuse, whereas those produced by aluminium boride crystals in the aluminium–0·04% boron alloy described above were very sharp. As the size of the crystals was not small enough to account for this difference, it apparently indicated a range of composition with a corresponding range in lattice dimensions. The lattice parameters obtained in these experiments are given in Table V, together with the parameters of the two pure borides. The results suggest that the boride which is formed in the aluminium–titanium–boron alloy is titanium boride with either excess titanium or considerable proportions of aluminium boride in solution; the latter explanation would not be altogether unexpected, in view of the close similarity of lattice structures and dimensions of the two borides and the fact that the crystals are formed in the presence of a great excess of aluminium.

TABLE V.—*Lattice Dimensions of the Hexagonal AlB_2-Type Structures Identified in Aluminium–Titanium–Boron and Aluminium–Boron Alloys.*

Lattice Dimensions, Å.	AlB_2	TiB_2	Crystals in Al–0·04% B Alloy	Crystals in Al–0·04% B–0·12% Ti Alloy
a_0	3·00	3·028	3·000	3·010–3·016 approx.
c_0	3·25	3·228	3·243	3·235–3·240 ,,
c/a	1·08(3)	1·066	1·081	1·074

The range of composition of the boride crystals indicates that they did not attain equilibrium with the melt. This may have been caused by coring produced during the growth of the crystals, or by a coating of titanium boride formed on the original crystals of aluminium boride introduced in the hardener alloy; the hollow, shell-like structures seen in the micro-section through the segregated particles (Fig. 2, Plate I) may be evidence of such a coating.

3. Grain Refinement of Some Commercial Casting Alloys

(a) *Grain-Sizes*

To obtain information on the practical value of the above results, a number of test-bars were cast in commercial alloys refined with titanium and boron which were added as the hardener alloys in Table I.

Three series of castings were made in aluminium–4·9% copper alloy containing 0, 0·01, and 0·05% boron respectively; the titanium content in each series varied from approximately 0·002% (the inherent content of the commercial-purity aluminium) to 0·2%. Two melts of each composition were prepared, one being cast at 680° and the other at 800° C.

The grain-sizes of the test-bars are shown in Fig. 16. The bars cast at 800° C. show the differences in grain-size most clearly and indicate the effectiveness of the boron additions in reducing grain coarsening caused by the use of this high casting temperature. Although an addition of 0·01% boron was inadequate to produce refinement when the titanium content was only 0·002%, in the presence of 0·01% or more titanium this small boron addition was just as effective as much larger amounts. The addition of 0·02% titanium and 0·01% boron together had a refining effect somewhat greater than that of 0·05% boron or of 0·20% titanium added singly.

The bars cast at 680° C. had a much smaller range of grain-sizes, though the slight differences observed were in the same order as in the bars cast at 800° C. Measurements [17] of the temperatures of the metal entering the test-bar moulds during casting showed that, even at the beginning of pouring, the temperature of the metal was at the liquidus temperature of the alloy; the particles of solid metal formed in the pouring stream as it cooled would subsequently act as nuclei and thus account for the small maximum grain-size in the castings poured at a low temperature. The loss of heat and the amount of solid formed during pouring would vary with the melting and casting conditions, and would be greatest with small melts such as were used in the present work.

Two combinations of titanium and boron contents were selected from the above results—0·03% titanium + 0·01% boron, and 0·01% titanium + 0·003% boron—and the increase in grain-size with casting temperature of aluminium–4·9% copper alloys refined with these additions was investigated more fully by casting test-bars at various temperatures between 680° and 900° C.

The results (shown in Fig. 17) confirm that the increase in grain-size with casting temperature is considerably less than in alloys refined with the much larger titanium additions normally made [4] (shown by the dotted line in Fig. 17). For castings made from melts which need not be heated above 720° C. the smaller titanium and boron additions should be adequate.

The grain-sizes of bars cast from three melts in aluminium–10% magnesium–0·02% beryllium alloy refined by the addition of 0·03% titanium + 0·01% boron are given in Table VI; the grain-sizes are similar to those obtained with aluminium–copper alloys.

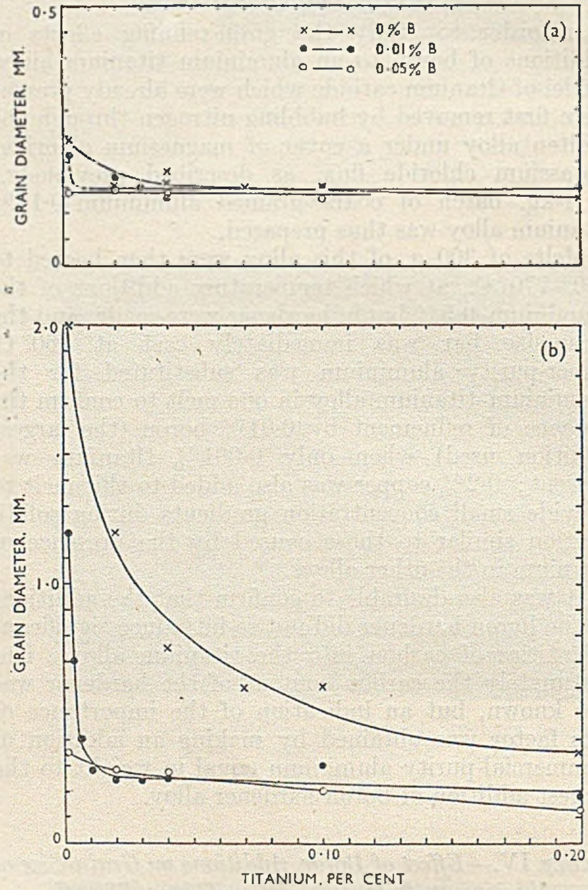

Fig. 16.—Variation of Grain-Size with Ti Content of Test-Bars in Al–4·9% Cu Alloy Containing 0, 0·01, or 0·05% B, poured at (*a*) 680° C. or (*b*) 800° C.

The addition of 0·01% boron to a secondary alloy (B.S. 1490, LM 4) already containing 0·08% titanium also produced a marked reduction in grain-size in test-bars poured at 700° C., from 2·5 mm. in the boron-free alloy to 0·25 mm. Many commercial casting alloys, although containing 0·1% or more titanium, are nevertheless coarse grained, because of the absence of nucleating particles, and the addition of small quantities of boron alone to these alloys should produce satisfactory grain refinement.

The above results show that the addition of 0·01% titanium and 0·003% boron should adequately refine the grain-size of aluminium alloy castings poured at temperatures below 720° C., but larger additions—up to 0·03% titanium and 0·01% boron—are required when higher casting temperatures are used. In attempts to prepare a hardener alloy containing both titanium and boron (described below, p. 13), severe gravity-segregation of titanium boride crystals occurred in the melt, suggesting that excessive additions of titanium and boron would cause similar segregation in melts of commercial aluminium casting alloys. A member of the Association has encountered segregation of titanium boride in melts held at 700° C., and containing only 0·005% of each element. It is clear, therefore, that additions of titanium and boron must be made with caution, especially where the melt is liable to stand undisturbed for long periods.

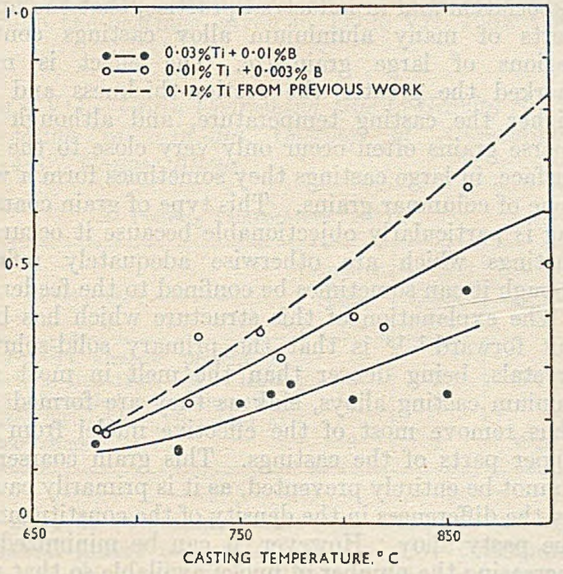

Fig. 17.—Variation of Grain-Size with Casting Temperature of Al–4·9% Cu Test-Bars Containing Ti and B.

Table VI.—*Grain-Sizes of Test-Bars in Aluminium–10% Magnesium–0·02% Beryllium Alloy Refined with* 0·03% *Titanium and* 0·01% *Boron.*

Casting Temp., ° C.	Grain Dia., mm.
700	0·30
750	0·33
800	0·37

(b) *Gas Pick-Up and Mould Reaction*

The increase in gas pick-up during melting and the promotion of mould reaction after casting, which were caused by large boron additions in previous work,[4] were confirmed in these experiments. The melts containing 0·05% boron required the longest periods of degassing with chlorine before the reduced-pressure test indicated freedom from dissolved gas, and the percentages of voids in the test-bars poured from these melts were high when the titanium content was below 0·1%. These effects were most marked in the alloys cast at the highest temperatures, as shown in Fig. 18.

Fig. 19 illustrates the sub-surface pinhole formation typical of mould reaction caused by 0·05% boron, in an aluminium–copper alloy of low titanium content cast at 680° C. The pinholes in an alloy cast at 800° C. were less regular, and much more porosity occurred in the interior of the test-bars, as would be expected. The high boron content produced other noticeable surface effects—an increase in dross formation, and a green tinge on the top surface of the feeder heads in alloys cast at the lower temperature—which are no doubt connected with the effects described above.

There were no significant differences in the percentages of voids in test-bars containing 0·01% or less boron cast at 680° C., but when the pouring temperature was above 750° C. even 0·01% boron produced a significant, though small, increase in porosity in alloys of low titanium content, as shown in Fig. 18.

When the titanium content was 0·1% or greater, no mould-reaction effects were seen even with the highest boron additions, the percentages of voids in the test-bars being the same as in the boron-free bars. This was probably the result of the reduction in the concentration of boron in solution in equilibrium with excess titanium, as suggested in Section V, 2 (*a*), for it is unlikely that boron present as titanium boride crystals, instead of in solution, would promote reaction of the melt with moisture.

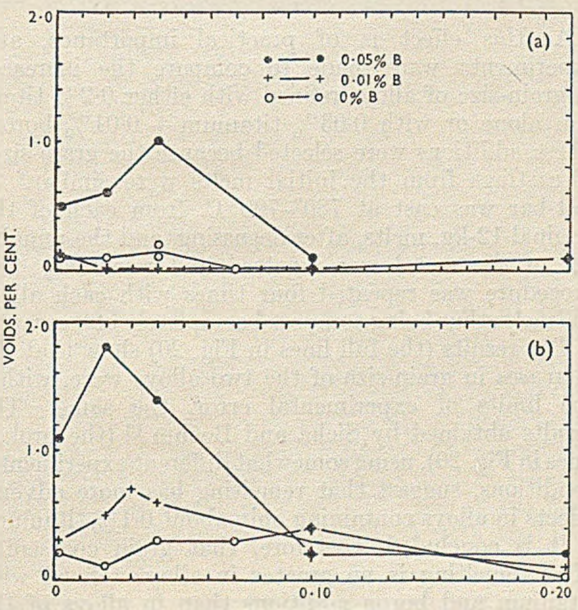

Fig. 18.—Variation of Percentage of Voids with Ti Content of Test Bars in Al–4·9% Cu Alloy Containing 0, 0·01, or 0·05% B, poured at (*a*) 680° C. or (*b*) 800° C.

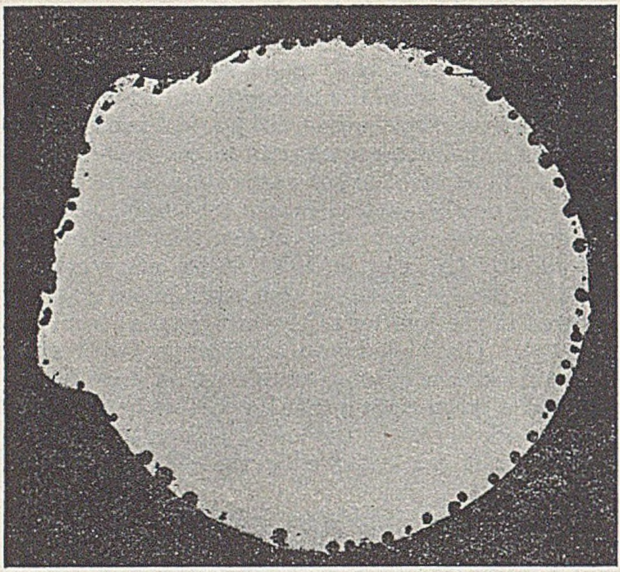

FIG. 19.—Sub-Surface Pinholes Caused by Mould Reaction in Test-Bar of Al–4·9% Cu–0·05% B Alloy Cast at 680° C. × 2·5.

(c) *Persistence of Refinement on Remelting*

An increase in grain-size in aluminium casting alloys may result not only from the melt being superheated but also from repeated melting,[13] the effect being greater the higher the melt temperatures. The reason for the grain coarsening is probably the increasing solubility of the nucleus compound in the melt as the temperature rises above the melting point; the smallest nuclei would tend to dissolve completely and, when the melt is cooled again, they would probably re-precipitate on to existing, undissolved nuclei, thus progressively reducing the total number of nuclei.

As this effect is of practical importance, and experiments were made to compare the increases in grain-size of alloys refined with either 0·2% titanium alone or with 0·03% titanium + 0·01% boron. These additions were selected because the grain-sizes of castings from the initial melts were similar. A test-bar was cast at 780°–790° C. from each of the original 12-kg. melts, after degassing, and the remainder of the metal was ingotted and remelted; this procedure was repeated four times with each alloy, the melts thus being composed completely of "scrap".

The results (the full lines in Fig. 20) show that the increases in grain-size of the two alloys were, within the limits of experimental error, the same. The results obtained by Sicha and Boehm [13] (the broken lines in Fig. 20), using somewhat different experimental conditions, suggest that remelting has more adverse effects in alloys containing only about 0·1% titanium.

It is concluded, therefore, that grain coarsening after remelting is no greater in alloys refined with titanium and boron additions than in alloys of the same initial grain-size refined with titanium alone. The grain coarsening after an addition of 0·03% titanium + 0·01% boron is less than in alloys containing the normal titanium additions of 0·10–0·15%.

(d) *Gravity Segregation During Solidification*

It has been observed, in previous work by the Association and in industrial practice, that the upper parts of many aluminium alloy castings contain regions of large grain-size. The effect is more marked the greater the section-thickness and the higher the casting temperature, and although the coarse grains often occur only very close to the top surface, in large castings they sometimes form a wide zone of columnar grains. This type of grain coarsening is particularly objectionable because it occurs in castings which are otherwise adequately refined, though it can sometimes be confined to the feeders.

The explanation of this structure which has been put forward [1, 18] is that the primary solid-solution crystals, being denser than the melt in most aluminium casting alloys, sink as they are formed and thus remove most of the effective nuclei from the upper parts of the castings. This grain coarsening cannot be entirely prevented, as it is primarily caused by the differences in the density of the constituents of the pasty alloy. However, it can be minimized by increasing the number of nuclei available so that new grains can form as the original crystals are removed by sedimentation; furthermore, the grains are finer and therefore sink through the melt more slowly.

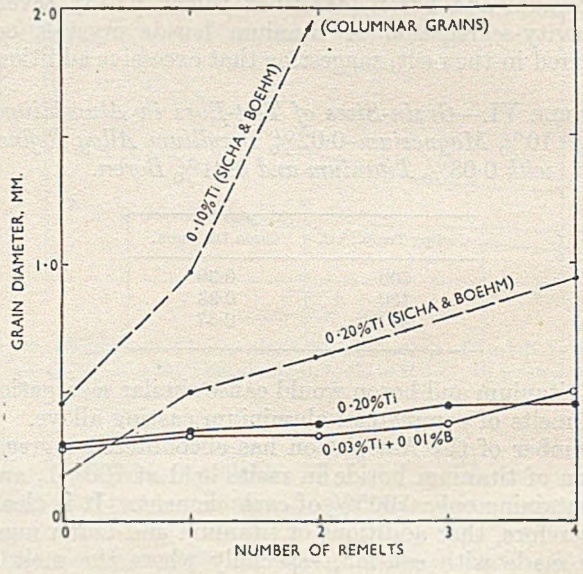

FIG. 20.—Increase of Grain-Size after Remelting in Al–4·9% Cu Alloy Test-Bars Refined with 0·20% Ti or 0·03% Ti + 0·01% B, Cast at 780°–790° C.

Grain coarsening caused by sedimentation in alloys containing titanium should be reduced by additions of boron, for this should increase the number of nuclei. To illustrate this effect of boron additions, horizontal cylinders of 5 in. dia. and 7 in. long were cast in aluminium–4·8% copper alloy containing (1) 0·13% titanium already present in the ingots of B.S. 1490, LM 11 alloy; (2) 0·13% titanium added as a hardener alloy; or (3) 0·03% titanium + 0·01%

boron added as hardener alloys. Each melt was poured at 725° C. after degassing.

Measurements of the grain-sizes in vertical sections through the castings, given in Table VII, show that the zone of columnar grains was smallest and the equiaxial grains were finest in the alloy containing both titanium and boron. These results confirm that the additions of titanium and boron together, in the above amounts (3), reduce but do not eliminate grain coarsening caused by gravity segregation during solidification.

TABLE VII.—*Grain Coarsening Due to Gravity Segregation in 5-in.-dia. Castings of Aluminium–4·8% Copper Alloys.*

Pouring Temperature 725° C.

Grain-Refining Additions	Depth of Columnar Zone, mm.	Width of Columnar Grains, mm.	Dia. of Equi-axial Grains, mm.
0·13% Ti already present in comm. B.S. 1490, LM 11 ingot	18	3	0·7
0·13% Ti added as hardener alloy	17	3	0·4
0·03% Ti + 0·01% B added as hardeners	9	1	0·25

4. ALTERNATIVE METHODS OF ADDING TITANIUM AND BORON

(a) *Aluminium–Titanium–Boron Hardeners*

As it would be of some advantage if the additions of titanium and boron could be made in a single hardener alloy, two attempts were made to prepare an alloy containing 1·5% titanium and 0·5% boron.

In the first experiment, an aluminium–1·5% titanium alloy was heated to 1600° C. and the required amount of aluminium–25% boron alloy was stirred into the melt through a flux of cryolite. The final alloy contained only 0·1% boron which was heavily segregated in the last-poured ingots, the titanium content being the same in all ingots.

Test-bars containing 0·03% titanium, made from this hardener and cast at temperatures up to 870° C., had grain-sizes only slightly coarser than are obtained by adding 0·12% titanium as a normal titanium hardener alloy. The addition of boron had therefore increased the effectiveness of the 1·5% titanium alloy as a grain-refiner, but the grain-sizes produced with this hardener were not as fine as can be obtained using separate additions of a boron hardener alloy.

The second titanium–boron hardener alloy was made by rapidly stirring an aluminium–10% titanium alloy into an aluminium–0·5% boron melt at 1600° C. and casting immediately. The first-poured ingot had titanium and boron contents of only 0·26 and 0·16%, respectively, the boride crystals having segregated into the final ingot.

These results confirm that the solubility of titanium boride in molten aluminium is very low, and it is doubtful whether it would be practicable to prepare a single hardener alloy containing large percentages of both titanium and boron without marked segregation.

(b) *Reduction of Salts of Titanium and Boron*

Titanium and boron may also be introduced into aluminium alloy melts by the reduction of salts such as potassium titanofluoride (K_2TiF_6) and potassium borofluoride (KBF_4), and by using mixtures of these salts the additions may be made simultaneously. A proprietary mixture of two similar salts is available.

Several experiments were carried out in which quantities of the proprietary material were plunged into 3-kg. melts of aluminium–4·9% copper alloy at 750° C. After the dross had been removed, the temperature of the metal was raised to 800° C. by bubbling chlorine through the melt and grain-size bars were cast; the resulting grain-sizes were compared with those obtained using hardener alloys. One melt of aluminium–10% magnesium alloy was similarly treated. The results are given in the first three parts of Table VIII.

The grain-sizes obtained, though fine, were not as small as those obtainable by using hardener alloys, even when excess of the salt was added. The reason for this difference may have been that the boride crystals produced by the addition of mixed salts were larger than those formed from hardener alloys, for analyses showed that the transfer of titanium and boron was generally efficient by both methods. Only with the 10% magnesium alloy was the reduction of salts inefficient, which is in agreement with earlier work.[3] The best result was obtained with the commercial 4·8% copper alloy already containing 0·13% titanium.

Other results (in the last part of the table) obtained with a simple mixture of potassium titanofluoride and potassium borofluoride showed that the temperature of addition affected the resulting grain-size; the mixture melted completely only at 750°–770° C. and when this occurred the resulting grain-size was as fine as that obtained by using hardener alloys, but at lower temperatures the addition of the mixture of salts was less effective, apparently because of loss of boron. The results obtained with the proprietary mixture would possibly have been better if a higher temperature had been used, but it would usually not be desirable to exceed 750° C. in commercial melting practice because of the increase in gas pick-up. The size of the melt may also affect the results obtained with the proprietary mixture.

5. THE EFFECT OF OTHER BORIDES

Zirconium, vanadium, niobium, and chromium form diborides similar to titanium boride, as described

Table VIII.—*Addition of Titanium and Boron to Aluminium Alloy Melts by Reduction of Salts.*

Mark, NOU	Alloy	Addition of Titanium and Boron						Grain-Size, mm.
		Source	Temperature of Addition, ° C.	Ti Content, %		B Content, %		
				Nominal	Analysis	Nominal	Analysis	
287	Al–4·9% Cu–0·001% Ti	Commercial preparation of Ti and B salts.	750	0·03	0·03	0·005	0·01	0·5
289				0·12	0·08	0·02	0·02	0·5
300	Al–4·9% Cu–0·001% Ti	Hardener alloys.	With charge	0·03	0·03	0·01	0·01	0·25
288	Al–10% Mg	Commercial preparation of salts.	750	0·03	<0·01	0·005	<0·005	1·0
307A	Al–4·8% Cu–0·13% Ti (comm. ingot)	No addition.	...	...	...	...	...	1·1
307B		Commercial preparation.	750	0·06	...	0·01	0·005	0·4
311	Al–4·9% Cu–0·001% Ti	Potassium titanofluoride + potassium borofluoride mixture.	710–730	0·03	0·03	0·01	<0·005	0·45
312			750–770	0·03	0·04	0·01	0·005	0·27

in Table III. Series of aluminium–4·9% copper alloys were therefore made to which were added 0·01% boron and varying amounts of each of the above elements, to compare the effects of these additions with that produced by using titanium. All the additions were made as hardener alloys, and the melts were cast at 800° C.

The results of the grain-size measurements are plotted in Fig. 21, together with those obtained from bars containing titanium with 0·01% boron. None of the four alternative transition metals was more effective than titanium in reducing the grain-size at low concentration, if the additions are compared on the basis of either weight per cent. or atomic per cent. Both vanadium and niobium produced marked grain refinement, but none occurred with zirconium or chromium additions, presumably because the concentrations of the latter elements were too low to precipitate their respective diborides, or because other (unsuitable) borides were formed.

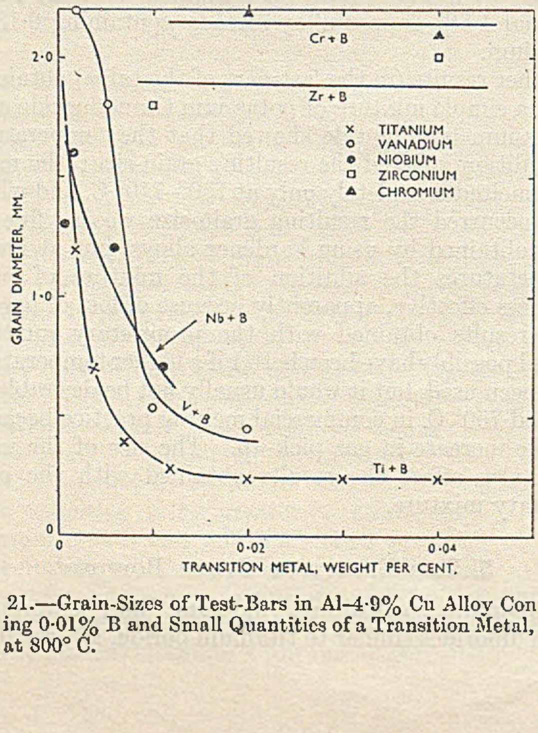

Fig. 21.—Grain-Sizes of Test-Bars in Al–4·9% Cu Alloy Containing 0·01% B and Small Quantities of a Transition Metal, Cast at 800° C.

VI.—SUMMARY AND CONCLUSIONS

(1) Substantial evidence has been obtained which shows that the grain refinement of aluminium alloys by the addition of boron is caused by nucleation by aluminium boride crystals in the melt. This conclusion was supported by other results obtained by using the isomorphous borides of certain transition metals. The metal atoms in these borides lie on simple hexagonal lattices, in the close-packed planes of which the interatomic distances are similar to the corresponding interatomic distance in solid aluminium.

(2) When boron is added to aluminium alloy melts containing titanium, marked grain refinement is produced which is thought to be caused by the formation of a boride with lattice dimensions similar to those of aluminium boride, which is probably titanium boride containing aluminium boride in solid solution. When the titanium content is over about 0·005%, this mixed boride is precipitated at very much smaller boron concentrations than are required to form pure aluminium boride in titanium-free alloys, and the boron addition necessary for maximum grain refinement is correspondingly less. For example, grain refinement of an aluminium–0·14% titanium alloy was produced by as little as 0·0001–0·001% boron—a quantity very similar to the estimated amount of carbon required to refine an alloy of similar titanium content.

(3) Two combinations of titanium and boron contents have been selected from these results for practical use—0·01% titanium + 0·003% boron for alloys cast at low temperatures, and 0·03% titanium + 0·01% boron for alloys which have to be heated above 720° C. before casting. Even the highest

additions represent a considerable saving in the use of titanium, and the grain refinement obtained is little affected by overheating the melt. Mould reaction and the loss of refinement caused by remelting in alloys containing the above additions are slight and not likely to be of any practical importance, and grain coarsening due to gravity segregation during solidification is reduced.

(4) Excessive additions of titanium and boron cause segregation of titanium boride crystals in the melt. As the loss of refinement on remelting is small, it should be unnecessary to make additions of titanium and boron to remelted scrap metal already containing these elements. Even with this precaution, however, segregation of titanium boride may still be encountered in quiescent melts unless the metal is stirred vigorously just before casting.

(5) Titanium and boron have also been added to aluminium alloy melts by the simultaneous reduction of molten salts of these elements, but the grain-sizes produced by this method are affected by the temperature of addition and the composition of the melt, and the method may be completely successful in practice only if the magnesium content is small and the temperature of addition high.

(6) Experiments on the addition of boron with transition metals other than titanium indicate that low concentrations of vanadium and niobium also produce marked refinement in the presence of boron.

(7) Very little success was obtained in attempts to increase the carbon content of alloys containing titanium; the addition of carbon presented much greater experimental difficulties than the addition of boron, and is unlikely to prove as successful as the latter.

ACKNOWLEDGEMENTS

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APPENDIX

THE SUPERSATURATION OF ALUMINIUM BORIDE IN ALUMINIUM

When the centrifuged aluminium–0·04% boron specimen, which was discussed in Section III, 2, was examined again after twelve months, a new constituent was observed over the whole cross-section, which was quite different in shape from the segregated aluminium boride crystals. The distribution of the new constituent suggested that it had precipitated from the solid metal on crystallographic planes and at the boundaries of dendrite arms as indicated in Fig. 3 (Plate I).* At a high magnification, the precipitate was similar in appearance to aluminium boride (AlB_2), but was quite different in shape, as shown in Fig. 4 (Plate I). It was largely situated in bands around what seemed to be sections of dendrite arms (as can be seen, for example, at the right of Fig. 3 (Plate I), so that the dendrites were revealed in a similar way to that usually produced by coring.

These observations can be interpreted in the following manner. The formation of primary aluminium boride crystals may be partly suppressed at high rates of cooling (40° C./min.) and the dissolved boron concentrates in the residual liquid as the supersaturated melt solidifies, producing coring. Some of the aluminium boride may form at the boundaries of dendrite arms, where its concentration is highest, at the moment when they solidify, but most of the compound precipitates from the solid metal on ageing.

This interpretation explains several observations which have been made in earlier work. For example, castings in high-purity aluminium–boron alloys, with a few exceptions, have columnar macrostructures, which can be explained by the absence of concentration gradients large enough to prevent the growth of the columnar grains in the solidifying alloys; some castings, however, have been observed with central zones, varying in size from casting to casting, of very fine equiaxial grains at the boundaries of which was a discontinuous constituent.[1] This phase now appears to have been aluminium boride which was formed in the way described above, and it is therefore possible that the concentration gradients which led to its formation caused the restriction of the growth of the dendrites which resulted in the fine equiaxial structures. The grain-boundary precipitate was observed in greatest amounts in castings made from melts which had been heated above 860° C., but the reason for the variation from casting to casting cannot be given.

The lamellar pattern of the precipitate between the arms of dendrites in the centrifuged specimen is apparently identical with the eutectic-like structures which have been observed in previous work.[3] The observation may also explain the wide diversity of equilibrium diagrams which have been proposed for the binary aluminium–boron system,[12] one of which showed a eutectic point at more than 15% boron and 565° C., and another a eutectic at less than 0·05% boron and 658° C., with primary aluminium boride forming only at higher boron contents in both cases.

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* During the period of twelve months the metallographic specimen had been mounted twice, the temperature being raised to approximately 130° C. each time.

REFERENCES

1. A. Cibula, *J. Inst. Metals*, 1949–50, **76**, 321.
2. W. A. Baker, *B.N.F.M.R.A. Research Rep.*, **A.550**, 1940.
3. M. D. Eborall, *J. Inst. Metals*, 1949–50, **76**, 295.
4. A. Cibula and R. W. Ruddle, *J. Inst. Metals*, 1949–50, **76**, 361.
5. D. C. G. Lees, *J. Inst. Metals*, 1946, **72**, 343.
6. A. L. Mincher, *Metal Ind.*, 1950, **76**, 435.
7. A. L. Mincher, *B.N.F.M.R.A. Research Rep.* **A.860**, 1949.
8. W. A. Baker and M. D. Smith, *B.N.F.M.R.A. Research Rep.* **A.694**, 1945.
9. W. L. Fink and L. A. Willey, "Metals Handbook", p. 1155. Cleveland, O.: **1948** (American Society for Metals).
10. W. Hofmann and W. Jäniche, *Z. Metallkunde*, 1936, **28**, 1.
11. N. B. Rutherford, B.N.F.M.R.A. unpublished work.
12. L. Mondolfo, "Metallography of Aluminium Alloys". New York: **1943** (John Wiley and Sons).
13. W. E. Sicha and R. C. Boehm, *Trans. Amer. Found. Soc.*, 1948, **56**, 398.
14. R. Kiessling, *Acta Chem. Scand.*, 1949, **3**, 90.
15. R. Kiessling, *Acta Chem. Scand.*, 1949, **3**, 595.
16. J. T. Norton, H. Blumenthal, and S. J. Sindeband, *Trans. Amer. Inst. Min. Met. Eng.*, 1950, **188**, 1060.
17. R. A. Skinner, B.N.F.M.R.A. unpublished work.
18. R. W. Ruddle, *J. Inst. Metals*, 1950, **77**, 1.

SOME OBSERVATIONS ON THE OCCURRENCE OF STRETCHER-STRAIN MARKINGS IN AN ALUMINIUM–MAGNESIUM ALLOY *

1320

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SYNOPSIS

Detailed observations have been made of the appearance of and dimensional distortion associated with surface markings developed by the progressive stretching of aluminium–3% magnesium alloy sheet in different conditions of cold working and annealing. In material of 0·025 mm. grain-size, markings develop with a very small strain and are at first normal to the tension axis but subsequently become random in direction, reaching maximum intensity at about 1%, and decaying within a 2% extension. These random markings consist of a series of kinks, and there is no thinning of the sheet. Parallel, intersecting bands or shallow grooves at a definite angle to the direction of stretching first appear at about 2% extension and increase progressively in intensity up to the point of fracture. Parallel bands, which are caused by local thinning or necking, are of much less intensity than the random markings, which are the main cause of defects in pressing operations.

When the grain-size exceeds 0·05 mm., random markings do not occur in stretching, nor are they obtained in partially annealed or temper-rolled sheet irrespective of grain-size. Parallel bands are found in all these materials. When the grain-size is increased substantially, the well-known "orange-peel" effect develops on stretching and completely masks any other effect which might be present, the degree of roughening from this cause in sheet of 1·0 mm. grain-size being comparable with that produced by random markings in fine-grained sheet.

I.—PREVIOUS WORK

STRETCHER-STRAIN markings are of particular significance in the manipulation of soft mild-steel sheet, and both practical and theoretical aspects of their occurrence have been the subject of numerous investigations. The physical appearance of strain markings and the conditions under which they occur have been discussed in some detail by Jevons,[1] who states that the uneven surface associated with stretcher-strains develops within an elongation of about 4%, beyond which the surface again becomes smooth and even. The standard method of preventing the occurrence of stretcher-strains in steel is by a light rolling reduction or passage through roller levellers, although the condition of immunity to strain marking is retained for only a short time.

Strain markings have been reported on several other metals, such as 70 : 30 brass, Duralumin,[2] and commercial aluminium,[3] but Jevons points out that in non-ferrous metals the effect is quite different from that experienced in steel, and occurs with much greater deformation. The intensity is also less, and all these metals are regularly used for deep and shallow pressings without any seriously detrimental stretcher-strain defects. Knight and Murray,[4] in a study of an aluminium–magnesium alloy, have shown that strain markings occurring with light stretching have some resemblance to those observed on mild steel. The markings were found to be particularly intense in fine-grained sheet, and in this connection it is of interest to observe that in an investigation by one of the present authors,[5] covering a number of alloys of zinc, strain lines were obtained only in a zinc–magnesium alloy, which has the characteristic of recrystallizing with a very fine grain-size.

In recent scientific studies of stretcher-strain phenomena, particular attention has been paid to the associated discontinuities in stress/strain curves, rather than to the strain markings themselves. McReynolds[3] has reviewed work in this field, and has also carried out a detailed investigation of the behaviour in straining of commercial-purity aluminium sheet, but it is clear that the relationship between the steps or discontinuities and strain markings is by no means clearly defined. For example, it has been reported that both mild steel[6, 7] and 70 : 30 brass[8] show stepped curves only at elevated temperatures, although pronounced markings develop in each case on stretching at room temperature. The extent to which discontinuities in the stress/strain curves are detected depends to some extent upon the elastic recovery in the straining mechanism, and studies of this type cannot be expected to provide information on the nature or range of occurrence of the strain markings themselves.

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II.—PRESENT INVESTIGATION

1. Scope

The experiments described were concerned with the appearance, magnitude, and conditions of occurrence of strain markings in an aluminium–3% magnesium alloy, which, from published literature and a preliminary survey made by the authors, was indicated as being probably the type of aluminium alloy most liable to such effects.

Various possible methods of measuring irregularities on a sheet surface were considered. For example, strained specimens were sectioned with the object of examining the surface under the microscope. However, even when tapered sections were cut at a narrow angle to the surface, and the precaution was taken of copper plating, in order to retain the sharpness of the surface, before mounting in methyl methacrylate, the magnitude of the effect was for the most part too small for an accurate assessment to be made by ordinary microscopical means. Mechanical methods proved more suitable and the experiments described were carried out with the Talysurf instrument, in which the displacement of a stylus traversing the specimen surface is magnified electronically and recorded on a paper chart.

The observation of Knight and Murray[4] that strain markings are most intense on annealed fine-grained material, was confirmed, and the stretching of fine-grained sheet up to the point of fracture was investigated in some detail. Additional observations were concerned with the study of markings in annealed sheet of different grain-sizes, and in temper-rolled and partially annealed sheet.

2. Preparation of Material

All the sheet material required was rolled from a single ingot of 5 in. thickness, the composition of which, as determined by chemical analysis on the 0·25-in.-thick, hot-rolled stock, was: magnesium 3·03, manganese 0·46, iron 0·23, silicon 0·14, copper 0·03, titanium 0·01%, remainder aluminium. An earlier investigation[9] had shown that the final grain-size of this type of alloy depends mainly on the ultimate cold-rolling reduction and, to a smaller extent, on the rate of heating during subsequent annealing. Consequently, in order to obtain the largest possible range of grain-sizes in the present experiments, the sequence of cold-rolling and intermediate annealing operations was so arranged that final rolling reductions ranged from 5 to 80%, intermediate annealing being carried out at 400° C.

The cold-rolled samples were annealed by three alternative procedures, decided by reference to the curves reproduced in Fig. 1, which were plotted from hardness tests on samples reduced 10, 25, and 80% in thickness by cold rolling, and annealed for 2 hr. at suitable temperature intervals.

After final rolling, rectangular specimens 8 × 2 in. were cut from the flattened sheet and the following treatment applied: (1) a rapid anneal in a salt bath at 500° C., designed to give the finest series of grain-sizes; (2) an anneal at 400° C. with slow heating in an air furnace, providing a series of coarser grain-sizes; or (3) an anneal at 325° C. in an air furnace in order to obtain the partially softened non-recrystallized condition on some of the specimens, and to increase still further the range of grain-sizes available.

A series of temper-rolled specimens was prepared from annealed 18-gauge sheet of fine grain-size, obtained by annealing at 400° C. after a final reduction of 80% in thickness, and therefore in its soft condition susceptible to stretcher-strain markings. Temper rolling was carried out in the same direction as previous cold rolling, and reduction in thickness ranged from 1 to 10%, control being effected by measurement of elongation on a length of 4 ft.

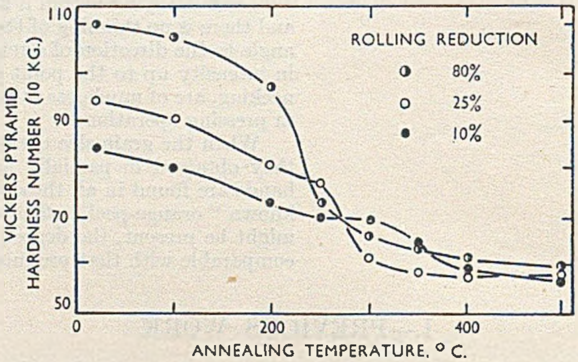

Fig. 1.—Annealing Curves of Aluminium–Magnesium Alloy Sheet.

Immediately before submitting them to stretching tests, all the 8 × 2-in. samples, whether annealed or temper-rolled, were polished by hand on both surfaces with metal polish, and then brightened by electrolytic treatment in a mixture of phosphoric acid and alcohol. The bright surface gave an almost level trace on the Talysurf apparatus and enabled markings to be seen or photographed with equal facility.

Surfaces were etched for grain-size measurement by immersing small areas at the ends of strips in the phosphoric acid mixture, and applying a lower voltage than that at which polishing takes place, for a period of some 10–20 sec. Grain-sizes were estimated by comparison, at a magnification of ×100, with a standard series of photomicrographs covering a range of mean grain diameters of 0·017–0·12 mm. For larger grain-sizes, the ×100 image was measured directly.

All strips annealed in a salt bath recrystallized. On the other hand, with specimens annealed in an air furnace at 400° or 325° C., recrystallization occurred only after rolling reductions greater than 5 and 10%, respectively. The grain-size of recrystallized material varied from 0·025 to 1·0 mm., the values obtained being included in Table III.

In the non-recrystallized specimens grain-sizes were apparently unchanged by the final rolling and annealing procedures. The X-ray diffraction patterns were, however, characteristic of what is now regarded as the recovered condition, in which the individual grains seen under the microscope are apparently agglomerates of smaller crystals, differing in orientation by only a few minutes of a degree.

3. STRETCHING EXPERIMENTS

The 8 × 2-in. specimens were stretched in a hydraulic tensile-testing machine, the operation being interrupted at progressively increasing extensions for Talysurf traces to be made on the rolled surfaces. Extensions of up to 1% were measured with a Lamb extensometer of the mirror-and-scale type, and those between 1 and 2% with a direct-reading dial instrument, the gauge-length being 2 in. in both cases. Extensions greater than 2% were measured with spring-loaded dividers.

(a) *Fine-Grained Specimens*

The whole range of surface effects broadly termed stretcher-strain markings can be most readily described by reference to the stretching of a sample prepared by annealing after a cold reduction of 80%, when at 400° C. the grain-size developed is 0·025 mm. Two distinct types of strain markings were observed. The first of these, which has been termed “ random markings ”, reaches a maximum intensity with about a 1% extension, thereafter diminishing rapidly in intensity with further stretching. The strain lines are often slightly curved and vary considerably in direction and in closeness of spacing, and the pattern appears to be identical on the two surfaces of the sheet. At an extension of about 2%, when random markings have virtually disappeared, the second type of marking begins to appear and consists of a cross-hatching of closely spaced parallel bands inclined at about 57° on either side of the line of stretching. These bands increase progressively in intensity, until necking and fracture eventually occur on one of them. The random markings at about 1% extension are by far the most significant and prominent of the two types of strain effect.

The nature of random markings, the progressive development of which is illustrated in Figs. 5, 6, and 7 (Plate IV), can best be understood by describing experiments carried out on a typical area of a stretched specimen, such as that illustrated in Fig. 2 (Plate III). The traces made by the Talysurf stylus on the two opposite faces of the sheet along the line *XX* are shown in Fig. 3 (Plate III), and in Figs. 2 and 3, the letters *A*1, *B*1 to *A*5, *B*5 refer to the successive areas over which the stylus passes. It will be noted that there is a sharp break or change in direction corresponding to the boundary line between light and dark areas. In effect, the surface can be regarded as consisting of planes inclined at variable small angles, which calculation shows to be generally smaller than 1′, to the general level of undistorted surface. Since, moreover, the traces are closely similar on the two sides of the specimen, it would seem that in effect the strip is concertina-ed, the two faces being parallel with no perceptible thinning at any point. A further series of surface traces along parallel lines 1–9 is recorded in Fig. 4 (Plate III). It will be noted that where the tongue-shaped marking is narrow the inclination is steeper than where the marking is wide, and it is clear, therefore, that there is a slight degree of curvature of the individual segments. The form of the surface can be more clearly understood by reference to Fig. 12 (Plate VI), which shows a model of a strained specimen with markings similar to those illustrated in Fig. 2. The surface contour is on the same magnification as Figs. 3 and 4, but thickness is reduced relative to the other dimensions.

The first random markings were detected at very small extensions. A typical specimen in which the overall plastic strain is 0·015% is illustrated in Fig. 5 (Plate IV). The surface within the gauge-length itself is free from markings, and the points over which markings occur may therefore be assumed to have suffered an extension exceeding 0·015%. This is confirmed by continuing the stretching, for with the first intrusion of strain markings into the gauge-length (Fig. 6, Plate IV), the extension has increased to 0·03%, and calculations from measurements of surfaces partially affected by strain markings indicates that the development of random markings is accompanied by an extension of at least 0·5%. The spread of strain marking over the whole specimen is illustrated in Fig. 7 (Plate IV), in which the measured plastic strain over the gauge length is 1%. Smooth and uniform stretching of a specimen up to 1% extension is difficult to achieve, because of the sudden stress-relief that accompanies the formation of new areas of random strain markings, but with careful control it is possible to watch the movement of the strain lines or tongue-like formations, which can proceed either by a smooth and regular movement or by more rapid shooting and darting. When stretching first begins the lines are almost normal to the direction of strain, but as extension proceeds they tend to curve at the ends, and also to fork and branch, so that by the time they approach maximum intensity at about 1% extension, they take a great variety of forms, and their direction appears to be quite random. There is, however, some reason to believe that if uniform straining in one direction could be achieved, the lines would remain normal to that direction, the forking and branching being probably associated with lack of homogeneity in the material due to slight variations in gauge and temper.

The degree of roughening or magnitude of the strain-marking effect on each individual specimen was taken as the maximum vertical distance between any pair of adjacent inflections in the Talysurf trace; for example, in Fig. 5 the distance measured is *PQ*. Mean values of such measurement obtained at intervals in the progressive stretching of a number of fine-grained specimens are plotted in Fig. 13 (Plate VI), curve *A* (the broken line) representing strains over which the surface was only partly affected by markings. The strain markings are probably of maximum intensity as soon as they appear, when their depth is about 0·003 in., although an exact evaluation is difficult when the surface is only partially affected. As stretching continues the bands multiply and groups of markings merge; accordingly, striations become less clearly defined, the measured roughness values diminishing rapidly so that at 2% elongation random marking has almost disappeared (Fig. 8, Plate IV) and surface irregularities are of the order of 0·0001 in. The sharp fall in roughness values between 1 and 2% extension will be observed in Fig. 13. Parallel bands, traversing specimens at an angle of about 57° to the direction of stretching, are just apparent at 2% extension, and roughness at this stage results partly from these bands and partly from residual random markings. As stretching is continued, parallel bands increase in depth and Fig. 9 (Plate V) illustrates the surface appearance of a specimen with a permanent extension of 5%, where the bands are closely spaced and quite conspicuous.

Features of parallel bands are their straightness, and manner of propagation, groups moving in a series of short jumps, seldom sweeping a distance greater than about 1 in. At 2% strain the magnitude of roughness associated with the bands is probably much less than 0·0001 in., but as stretching continues, the magnitude increases progressively, as indicated in Fig. 13, the maximum depth at an extension of 10% being 0·0003 in., when the operations were discontinued. Surface-contour records taken at this stage, Fig. 10 (Plate V), show that every depression in the system of parallel bands has an associated depression of similar magnitude in the opposite surface, causing in effect, a series of necks, in one of which, when stretching is continued, fracture eventually occurs.

(b) *Specimens of Differing Grain-Size*

Similar though less-detailed observations were made on the full range of annealed specimens, the appearance of which, after stretching to 1, 5, and 10% permanent extension, is recorded in Table I. Plots of the severity of strain marking, using the basis of measurement already described in connection with the sheets of fine-grained specimens, are shown in Fig. 13, for a representative series of grain-size values.

Behaviour in stretching was related entirely to grain-size. Specimens prepared by different rolling and annealing schedules, but with equal grain-size values, developed the same types and intensities of strain marking.

Random markings occurred with maximum intensity at about 1% extension in specimens with grain-sizes of from 0·025 to 0·042 mm. and were similar in appearance and of the same order of intensity over this range, as was also the decay of marking as the extension was increased from 1 to 2%. A specimen with the next largest grain-size of 0·05 mm., when stretched up to an extension of 1%, was, however, entirely free from random markings.

Clearly marked parallel bands developed in all specimens within the grain-size range of 0·025–0·06 mm., and the intensity of marking for any given degree of deformation appeared to be, within the limits of experimental error, independent of grain-size, as can be seen from the three curves relating to specimens within this range of grain-sizes plotted in Fig. 13. With grain-sizes of 0·05 mm. and over, parallel bands were not masked by random markings, and in some instances were detected visually at 1% strain, but they were too shallow at this stage—less than 0·0001 in.—to measure with the Talysurf instrument. With increasing strain, parallel bands became progressively deeper, the merging of bands being the mechanism by which this occurred up to the point at which necking began at an extension of about 20%.

The upper grain-size limit for the appearance of parallel bands is governed by the occurrence of the well-known "orange-peel" effect, which results from non-homogeneous deformation within individual grains in material of large grain-size. In stretching sheet of 0·25 mm. grain-size the orange-peel effect is superimposed on parallel bands, but the roughness value at between 1 and 10% extension is only slightly higher than in fine-grained material. With further increase in grain-size the orange-peel markings are of much greater significance, and with a 1-mm. grain-size roughening is severe and completely masks parallel bands at all extensions (see Fig. 11, Plate V).

Table I.—*Stretching Characteristics of Annealed Sheet.*

Rolling Reduction, %	Grain-Size, mm.	Type of Marking		
		1% Extension	5% Extension	10% Extension
Annealed 30 min. in a salt bath at 500° C.				
5	1·0	Severe orange-peel effect.	Severe orange-peel effect.	Severe orange-peel effect.
10	0·15	Slight orange-peel effect.	Orange-peel effect and parallel bands.	Orange-peel effect and parallel bands.
20	0·060	No markings.	Parallel bands.	Parallel bands.
25	0·057			
30	0·042	Severe random markings.		
40	0·040			
80	0·025			
Annealed 2 hr. in an air furnace at 400° C.				
5	0·025 °	Slight parallel bands.	Parallel bands.	Parallel bands.
10	0·25	Slight orange-peel effect.	Orange-peel effect.	Orange-peel effect.
20	0·08	No markings.	Parallel bands.	Parallel bands.
25	0·054			
30	0·042	Severe random markings.		
40	0·038			
80	0·025			
Annealed 6 hr. in an air furnace at 325° C.				
5	0·025 °	Slight parallel bands.	Parallel bands.	Parallel bands.
10	0·025 °			
20	0·057 †	No markings.		
25	0·057			
30	0·05			
40	0·042	Severe random markings.		
80	0·026			

° Not recrystallized. † Partially recrystallized.

PLATE III.

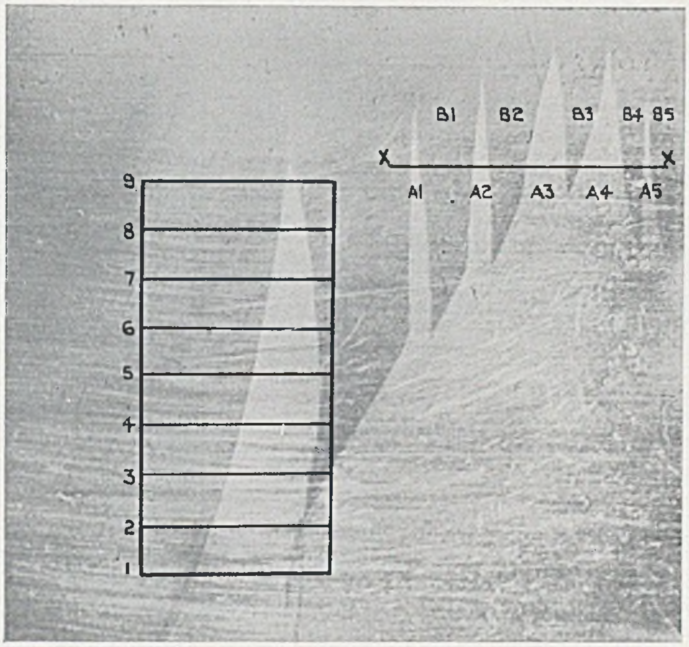

FIG. 2.—Typical Random Markings in Sheet of 0·025 mm. Grain-Size. × 2.

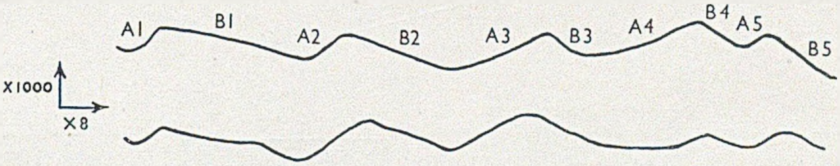

FIG. 3.—Talysurf Traces on Top and Bottom Surfaces Along Line *XX* in Fig. 2.

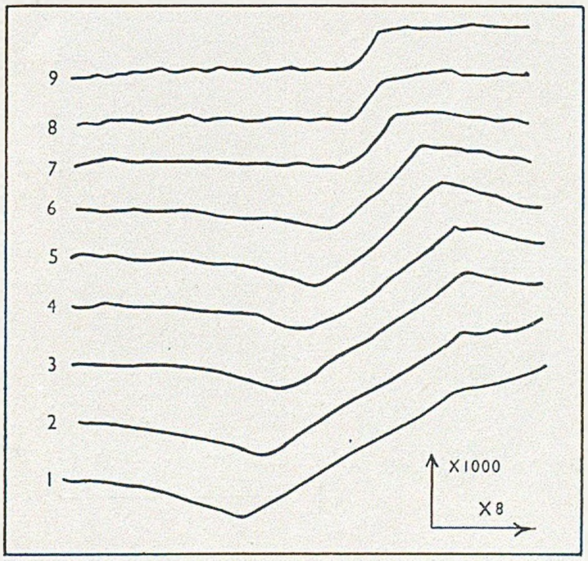

FIG. 4.—Talysurf Traces Along Lines 1 to 9 in Fig. 2.

[*To face p.* 20.

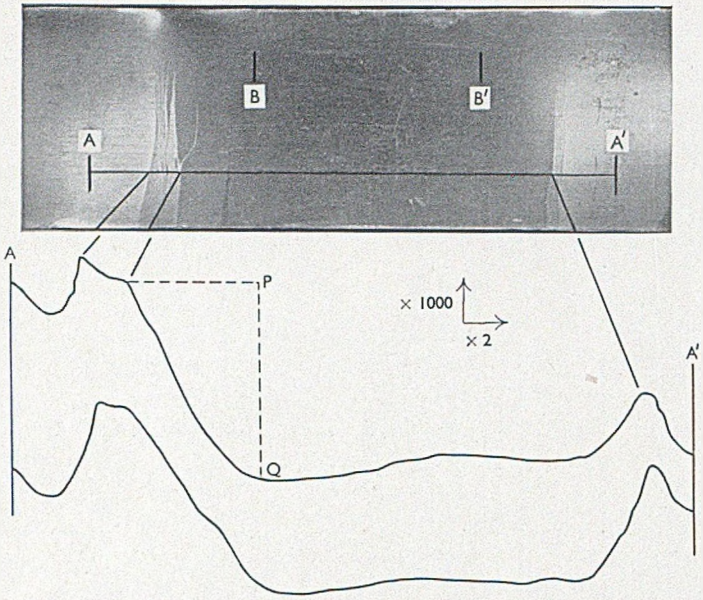

Fig. 5
(0·015%)

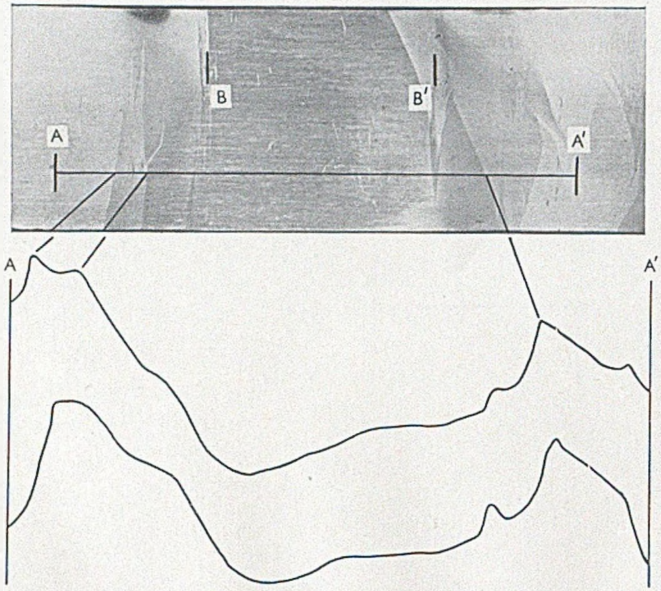

Fig. 6
(0·03%)

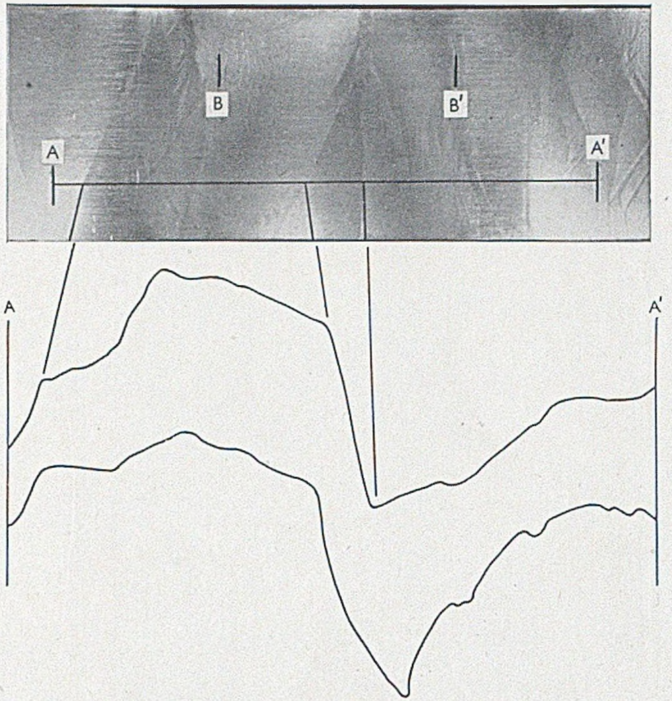

Fig. 7
(1·0%)

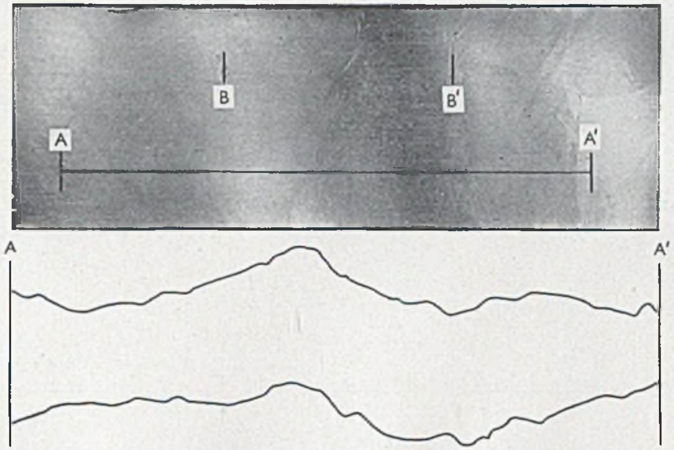

Fig. 8
(2·0%)

Plate IV.

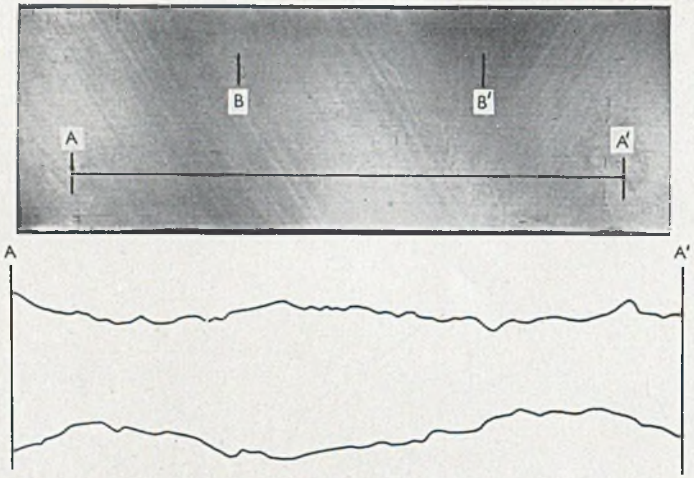

Fig. 9
(5·0%)

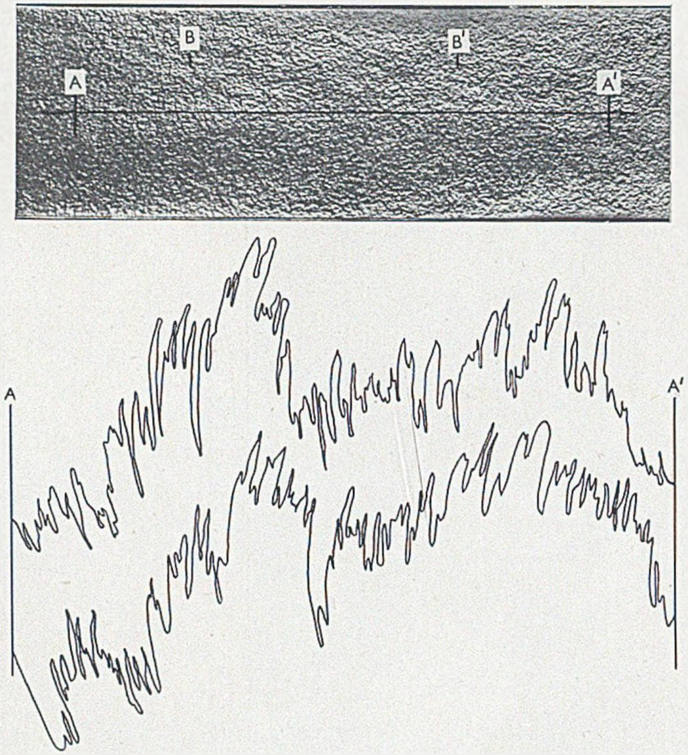

Fig. 11
(10%)

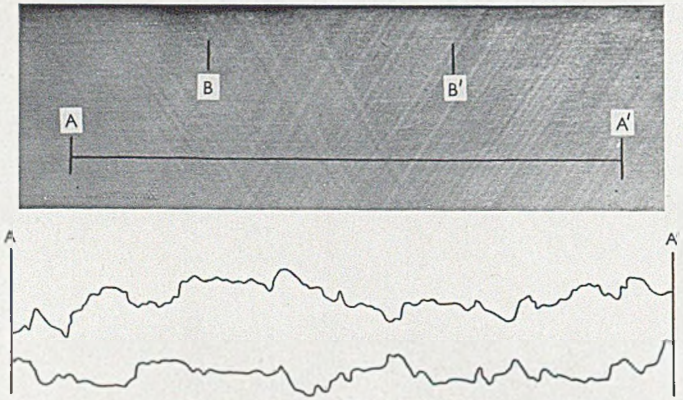

Fig. 10
(10%)

Figs. 5–10.—Strain Markings Obtained by the Progressive Stretching of an Aluminium–Magnesium Alloy Sheet of 0·025 mm. Grain-Size, with Talysurf Traces of Upper and Lower Surfaces from *A* to *A′*.
Fig. 11.—Markings on Aluminium–Magnesium Alloy Sheet of 1·0 mm. Grain-Size.

Permanent extensions over the 2-in. gauge length *BB′* are as indicated in parentheses.

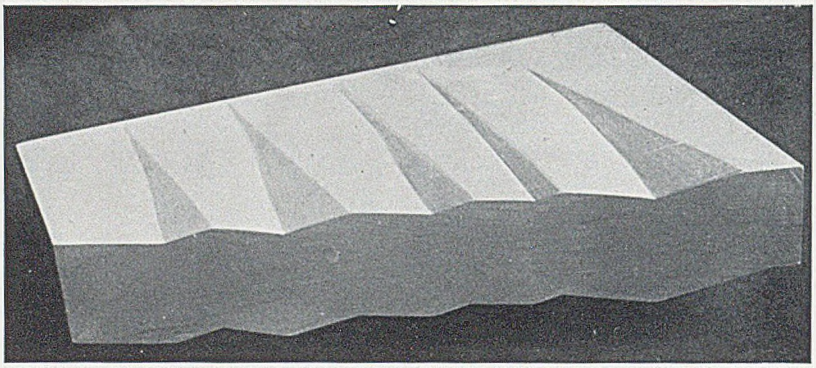

FIG. 12.—Solid Model of a Group of Random Markings.

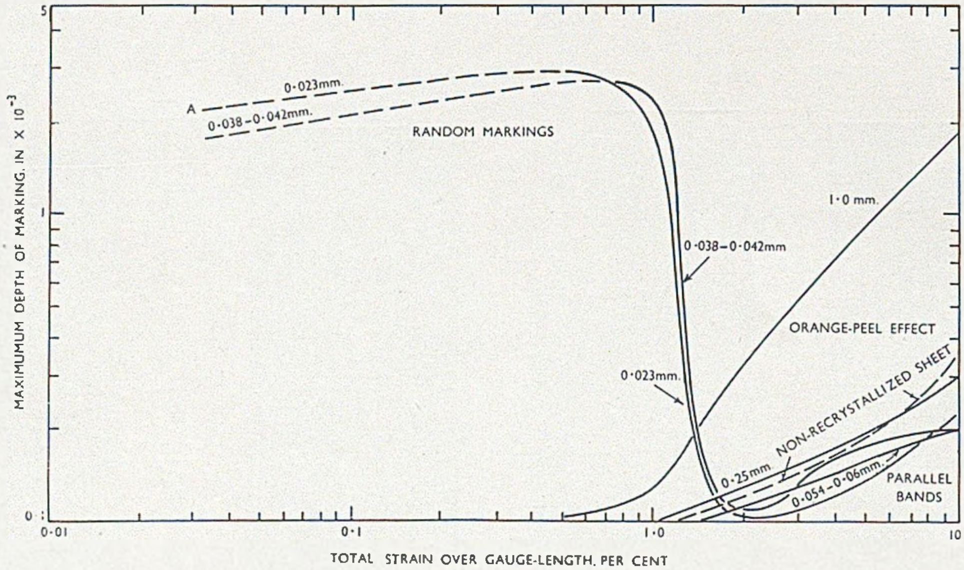

FIG. 13.—Effect of Strain on the Depth and Type of Markings on Annealed Sheets of Mean Grain-Size Ranging from 0·023 to 1·0 mm.

TABLE II.—*Stretching Characteristics of Temper-Rolled Sheet of Grain-Size, Before Rolling,* 0·025 *mm.*

Rolling Reduction, %	0·5% Extension		1% Extension		2% Extension		4% Extension	
	Maximum Talysurf Deviation, 10^{-3} in.	Type of Marking	Maximum Talysurf Deviation 10^{-3} in.	Type of Marking	Maximum Talysurf Deviation, 10^{-3} in.	Type of Marking	Maximum Talysurf Deviation, 10^{-3} in.	Type of Marking
1	0·1	Parallel bands	0·1	Parallel bands	0·15	Parallel bands	0·18	Parallel bands
2	0·1		0·1		0·15		0·18	
4	0·15		0·15		0·2		0·22	
10	0·18		0·2		0·22		0·3	

The degree of roughening resulting from orange-peel effect with a 1-mm. grain-size at 10% stretching is indicated in Fig. 1 (Plate VI), and it will be noted that the measured irregularities are of about the same order as those associated with random markings in fine-grained sheet with 1% stretching. Orange-peel effects, like parallel bands, are of relatively slight significance when the amount of stretching is small.

The non-recrystallized specimens, all three of which had about the same apparent grain-size of 0·025 mm., were entirely free from the random strain markings which are obtained in the fully recrystallized metal of similar grain-size. Parallel bands, however, began to appear at about 1% strain and deepened progressively with further stretching in precisely the same manner as in recrystallized sheet, a typical surface-roughness curve being shown as a broken line in Fig. 13.

(c) *Temper-Rolled Specimens*

As already stated, sheet for temper-rolling experiments was annealed at 400° C. after a final rolling reduction of 80%, and, having a grain-size of about 0·025 mm., was intensely susceptible to random markings.

Detailed observations were made and results are recorded in Table II on specimens subjected to rolling reductions of 1, 2, 4, and 10% in thickness; and in order to ascertain whether the behaviour in stretching is affected by strain-ageing effects, as it is in mild steel, the first series of experiments, carried out within 1 hr. of rolling (and in one or two instances within a few minutes of rolling) were repeated after an interval of three months.

None of the temper-rolled specimens showed any signs of random markings with light straining, as indeed might be expected by analogy with steel. Parallel bands, however, could be observed in every instance with only 0·5% strain and were, in all but the most lightly rolled specimens, of measurable depth with this extension, so that these bands were clearly of greater significance than in annealed sheet. The surface appearance and the maximum degree of irregularity of the various temper-rolled specimens with applied strains of 0·5, 1·0, 2·0, and 4·0% are recorded in Table II. It will be noted that the depth of the bands increases progressively with stretching, as in annealed sheet, while for a constant deformation, depth also increases slightly with temper-rolling reduction; but even with a 10% temper rolling, the depth of parallel bands for any given extension is only about twice that obtaining in annealed sheet, and the total surface irregularity associated with parallel bands is small compared with that of random markings.

4. MECHANICAL PROPERTIES OF SHEET

In annealed recrystallized sheet, hardness, proof stress, ultimate tensile strength, and elongation all increase by small progressive amounts with diminishing grain-size (Table III). For example, in alloys with a 0·05–0·06-mm. grain-size, and therefore free from random marking effects, the proof stress is about 1 ton lower than in sheet of 0·025-mm. grain-size, which is susceptible to random markings; in

TABLE III.—*Mechanical Properties of Annealed Sheet.*

Rolling Reduction, %	Grain-Size, mm.	Vickers D.P. Hardness Number (10 kg. load)	0·1% Proof Stress, tons/in.^{2}	Ultimate Tensile Strength, tons/in.^{2}	Elongation on 2 in., %
Annealed 30 *min. in a salt bath at* 500° C.					
5	1·0	55	4·8	13·6	24
10	0·15	57	4·9	14·0	24
20	0·060	57	5·6	14·2	23
25	0·057	58	5·6	13·8	23
30	0·042	58	5·6	13·9	26
40	0·040	59	6·0	14·0	26
80	0·025	60	6·6	14·2	26
Annealed 2 *hr. in an air furnace at* 400° C.					
5	0·025 *	64	7·4	14·7	20
10	0·25	52	5·1	13·9	21
20	0·08	52	6·0	14·1	22
25	0·054	53	5·7	14·0	23
30	0·042	53	6·0	14·0	23
40	0·038	55	6·4	14·3	23
80	0·025	55	6·9	14·4	24
Annealed 6 *hr. in an air furnace at* 325° C.					
5	0·025 *	66	7·5	14·4	20
10	0·025 *	67	7·6	14·4	19
20	0·057 †	61	7·1	14·2	23
25	0·057	58	6·2	14·2	22
30	0·05	59	6·2	14·0	21
40	0·042	59	6·5	14·2	23
80	0·026	61	7·1	14·4	25

* Not recrystallized. † Partially recrystallized.

other properties the proportionate differences are smaller. Again, sheet annealed at a low temperature so as to give a non-recrystallized or recovered structure is slightly higher in proof and tensile strength than fully annealed material. Temper-rolled sheet shows a wider range of mechanical properties (Table IV), strength increasing progressively with rolling reduction, but where this does not exceed 2%, ductility is not seriously impaired.

TABLE IV.—*Mechanical Properties of Temper-Rolled Sheet.*

Rolling Reduction, %	Vickers D.P. Hardness Number (10 kg. load)	0·1% Proof Stress, tons/in.²	Ultimate Tensile Strength, tons/in.²	Elongation on 2 in., %
Annealed	54·9	6·9	14·4	24
1	70	8·3	14·6	20
2	73	9·4	14·8	19½
4	82	11·4	15·3	15
10	84	12·2	16·2	9

III.—SIGNIFICANCE OF THE RESULTS

The experiments show that strain markings in annealed recrystallized sheet can be divided into two distinct types. Random markings are associated entirely with deformations of less than 2% and reach maximum intensity with about 1% stretching. They are found only in material with a grain-size below a sharply defined limit of about 0·05 mm. Parallel bands occur in all material irrespective of grain-size, but where this exceeds 1·0 mm. they are completely masked by the orange-peel effect. In annealed material parallel bands reach a measurable intensity at an extension of 2%, thereafter increasing progressively in magnitude up to the point of fracture. On the other hand, in sheet softened by a non-recrystallizing anneal or temper-rolled after a full anneal, parallel bands are of measurable depth at 1% strain, but in neither of these conditions is the surface affected by random markings.

The depth of random markings is about 0·003 in. at 1% extension, while parallel bands do not exceed a depth of 0·0003 in. with extensions of up to 10%, so that random markings are the only ones seriously detrimental in pressing. They can be avoided by subjecting sheet to one of the following procedures:

(1) By adjusting rolling and annealing schedules to give sheet with a final grain-size of 0·05–0·08 mm.

(2) By the application of a small final rolling reduction followed by a low-temperature non-recrystallizing anneal.

(3) By the application of a final temper-rolling reduction to fine-grained annealed sheet.

The third procedure, unlike temper rolling in mild steel, provides material which, over long periods of storage, is immune from any tendency to develop the more conspicuous, or random, type of strain effect on stretching.

It seems likely that the parallel markings, observed on every sample examined in the present investigation irrespective of its physical condition, are of more general occurrence than random markings, and indeed from careful reading of published work it appears probable that parallel banding is the type of marking reported in brass, Duralumin, aluminium, and a variety of other metals in sheet form. Random markings seem to be closely analogous to stretcher-strain effects in soft mild steel although according to Jevons the markings of steel are characterized by depressions on the two opposite faces of the sheet, giving rise to a local necking effect, and not to the kinking observed in the present investigation on aluminium–magnesium alloy sheet.

It is of interest to note that few metals or alloys can readily be obtained with the fine grain-size which the present experiments indicate to be essential to the development of random markings in aluminium–magnesium alloy, and this may well be one reason for the relatively rare occurrence of this phenomenon.

ACKNOWLEDGEMENTS

The authors' thanks are due to Dr. Maurice Cook, for his helpful advice and encouragement, and to various colleagues who have assisted in the experimental work.

REFERENCES

1. J. D. Jevons, "The Metallurgy of Deep Drawing and Pressing". 2nd edn. London: **1942** (Chapman and Hall, Ltd.).
2. E. W. Fell, *Carnegie Schol. Mem., Iron Steel Inst.*, 1936, **26**, 123.
3. A. W. McReynolds, *Trans. Amer. Inst. Min. Met. Eng.*, 1949, **185**, 32.
4. G. A. Knight and G. Murray, *Sheet Metal Ind.*, 1946, **23**, 1741.
5. R. Chadwick, *J. Inst. Metals*, 1933, **51**, 120.
6. A. Le Chatelier, *Rev. Mét.*, 1909, **6**, 914.
7. F. Körber and A. Pomp, *Mitt. K.W.-Inst. Eisenforsch.*, 1927, **9**, 346.
8. T. Sutoki, *Sci. Rep. Tôhoku Imp. Univ.*, 1941, [i], **29**, 673.
9. R. Chadwick and W. H. L. Hooper, *J. Inst. Metals*, 1948–49, **75**, 609.

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THE EFFECT OF PHOSPHORUS ON THE CORROSION-RESISTANCE OF MAGNESIUM AND SOME OF ITS ALLOYS*

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SYNOPSIS

Corrosion tests have been carried out, by total immersion in salt solution, on samples of magnesium alloys containing various amounts of phosphorus and other impurities. It is concluded that the corrosion-resistance of pure magnesium is reduced by the presence of phosphorus in amounts over about 0·002%, but that this effect can be suppressed by addition of manganese. With a normal iron content, the adverse effect of phosphorus is not suppressed by manganese. Phosphorus also exerts a marked adverse effect on the corrosion-resistance of high-purity Elektron A8 (aluminium 8, zinc 0·4, manganese 0·25%, remainder magnesium).

The amount of phosphorus that can be introduced into magnesium is much greater in the presence of iron. Phosphorus increases iron pick-up in remelting A8 alloy.

Phosphorus is precipitated from magnesium by zirconium, and the corrosion-resistance of alloys containing zirconium is unaffected by phosphorus present in the metal used for their preparation.

I.—INTRODUCTION

An important study of factors affecting the corrosion-resistance of magnesium and its alloys has been made by Hanawalt, Nelson, and Peloubet.[1] This showed the general nature of the adverse effects of iron and nickel on corrosion-resistance, the existence of tolerance limits, and the manner in which the positions of these limits can be modified by the presence of aluminium, zinc, and manganese.

From the work of these authors it might be thought that, in order to prepare magnesium and its alloys in the state of highest corrosion-resistance, it would suffice to reduce the iron and nickel contents to below about 0·002% and 0·0005%, respectively, and to add sufficient manganese. This is by no means the case, however. Hanawalt, Nelson, and Peloubet were mainly concerned with the *total* quantities of impurities present, and the conditions of preparation of their samples, namely, slow cooling of the metal in a body of molten flux, were such as to favour settling out of insoluble particles. Since commercial alloys contain varying amounts of insoluble particles and may exhibit microporosity, a knowledge of the extent to which the presence of such defects can reduce corrosion-resistance is a matter of practical importance, and one which it is hoped to discuss in a later paper.

Except where otherwise stated, the present paper refers to metal free from appreciable microporosity or abnormal numbers of particles. Even with such metal considerable variations in corrosion-resistance can occur; for instance, two samples of high-purity § Elektron A8 alloy, with iron and nickel contents below the limits mentioned above and containing 0·25% manganese, can differ in corrosion rate || by a factor of 100.

In considering the reasons for such large differences in corrosion-resistance, the possibility naturally arises that there may be other elements besides iron and nickel that can exert adverse effects. Comparative analyses for metallic impurities of samples of high-purity A8 of widely differing behaviour yielded no indications that any of the impurities detected might be harmful. Some circumstantial evidence suggested, however, that phosphorus might be a factor in determining corrosion-resistance. In particular, it was observed that the lowest corrosion rates obtained with A8 alloy prepared from electrolytic magnesium containing 0·015–0·025% phosphorus were appreciably higher than the lowest rates for the same alloy made from either ferro-silicon magnesium or electrolytic magnesium of very low phosphorus content; there was, however, little difference in average corrosion rate between the alloys made from these two grades of magnesium, at any rate on a production scale. It was accordingly decided to investigate the effect of phosphorus on corrosion-resistance, and an account

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§ It is convenient to use this term to designate Elektron A8 of low iron and nickel contents.

|| Statements concerning corrosion-resistance and corrosion-rate figures in the present paper refer to weight loss (mg./cm.^{2}/day) during 28 days' complete immersion in 3% NaCl solution saturated with $Mg(OH)_2$.

of the work done and results obtained is given below.

II.—INTRODUCTION AND REMOVAL OF PHOSPHORUS

In endeavouring to prepare corrosion specimens which would show unequivocally the effect of phosphorus, it was necessary to find a method of introducing the phosphorus without appreciable iron pick-up. Direct use of red phosphorus, even in conjunction with graphite crucibles and melting tools, proved unsatisfactory, and finally the phosphorus was introduced by plunging into the metal a mixture of A.R. calcium phosphate with sufficient Melrasal Z flux to control burning. This process had to be carried out in graphite crucibles using graphite plunging tools, as otherwise iron pick-up was excessive even at low temperatures, and it appeared that the presence of phosphorus in magnesium appreciably increased iron pick-up from mild-steel crucibles.

1. Solubility of Phosphorus in Magnesium

Using graphite crucibles and tools, the maximum amount of phosphorus that can be introduced into pure magnesium at normal operating temperatures is about 0·09%. If, however, iron crucibles are used, phosphorus contents up to about 0·4% can be achieved, introduction of the extra phosphorus being accompanied by iron pick-up. For each phosphorus content above about 0·09%, a *minimum* iron content is found (see Fig. 1), indicating some association of iron and phosphorus in the molten alloy. This is further confirmed by the observation that, in the presence of much phosphorus, removal of iron from A8 by addition of manganese and settling is incomplete.

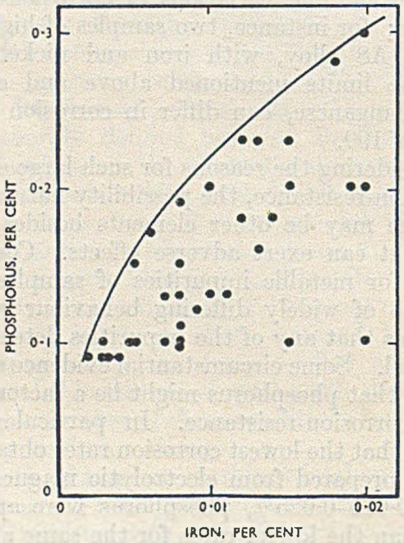

Fig. 1.—Relation Between Phosphorus and Iron Content for Melts Containing More Than 0·08% Phosphorus.

With phosphorus contents above about 0·03%, small cavities are present in the solidified alloys (Figs. 2–5, Plate VII), and, from the number and size of these, the phosphorus content can be roughly estimated. The cavities are mentioned by Vosskühler,[2] but not by Bulian and Fahrenhorst,[3] who found in alloys containing phosphorus a constituent resembling Mg_2Si in habit which they attributed to the compound Mg_3P_2. In addition to these small cavities, metal saturated with phosphorus may develop considerable piping on solidification (Figs. 6 and 7, Plate VII), inflammable vapour, apparently phosphorus, being evolved. From the uniform distribution of cavities in metal containing phosphorus, it appears that the latter is largely in solution in the molten magnesium; and with less than 0·09% phosphorus and in the absence of more than about 0·002% iron, little analytical evidence of segregation has been obtained. With high phosphorus and iron contents, some segregation of phosphorus can occur in the molten alloy, especially if the melt is settled with the top kept hotter than the bottom to minimize convection; irregular phosphorus segregation in the solid metal is frequent. Repeated fluxing of magnesium with a high phosphorus content was found to reduce the phosphorus to about 0·1%. Treatment with a fully chlorinated hydrocarbon, such as hexachlorethane, does not remove phosphorus.

From the foregoing observations it appears probable that phosphorus can be present in molten magnesium, both in the dissolved state and as an insoluble complex with iron.

2. Removal of Phosphorus by Precipitation and Settling

The introduction into magnesium of zirconium and certain other of the less-common elements causes precipitation of any phosphorus present. With zirconium the phosphorus is precipitated in the form of insoluble zirconium–phosphorus particles. If the full zirconium content is not added, or if it is added and the metal is subsequently remelted, the precipitated phosphorus settles out to give a final content of not more than 0·002%. The phosphorus content of zirconium-containing alloys when the full zirconium content is added in one stage may be somewhat higher, e.g. 0·004%, presumably because settling of the precipitated phosphorus is less complete in the more viscous alloy.

When it is desired to remove phosphorus without leaving residual zirconium dissolved in the metal, the dephosphorization may be performed by adding a few per cent. of a zirconium–aluminium master alloy. This is itself insoluble in magnesium and forms insoluble particles on which the phosphorus precipitates and which are easily removed by settling. Dephosphorization without leaving residual precipitant in the final metal can also be effected by additions of rare-earth metals, the excess being removed by fluxing.

PLATE VII.

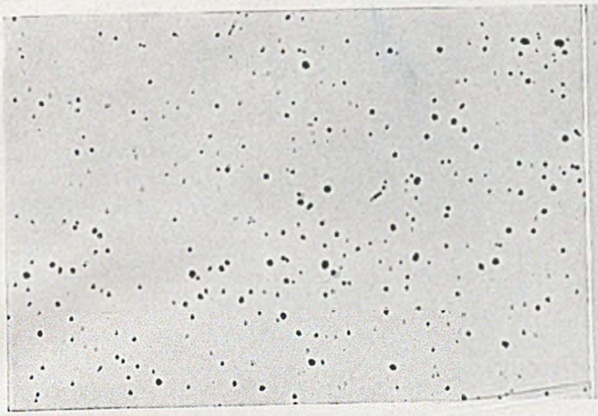

FIG. 2.—Polished Section of Magnesium Containing 0·2% P, 0·02% Fe, Showing Numerous Cavities. Unetched. × 50.

FIG. 3.—Section of Same Metal after Remelting and Repeated Fluxing. P = 0·1%. Unetched. × 50.

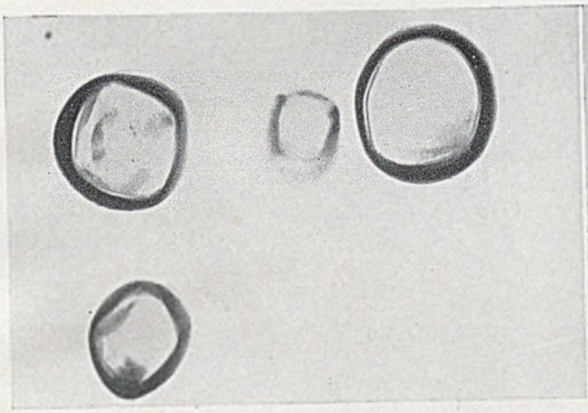

FIG. 4.—Same Metal Polished Electrolytically. × 1000.

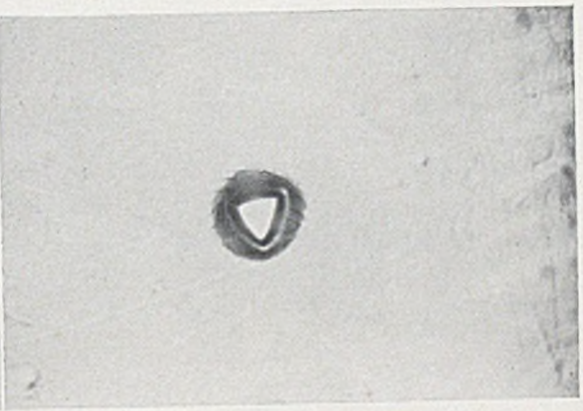

FIG. 5.—Cavity Showing a Triangular Facet. × 1000.

FIG. 6.—Ingot Poured at 750° C. from a Melt of Magnesium Saturated with Phosphorus, Showing Extensive Areas of Oxidation and General " Gassy " Appearance. × ½.

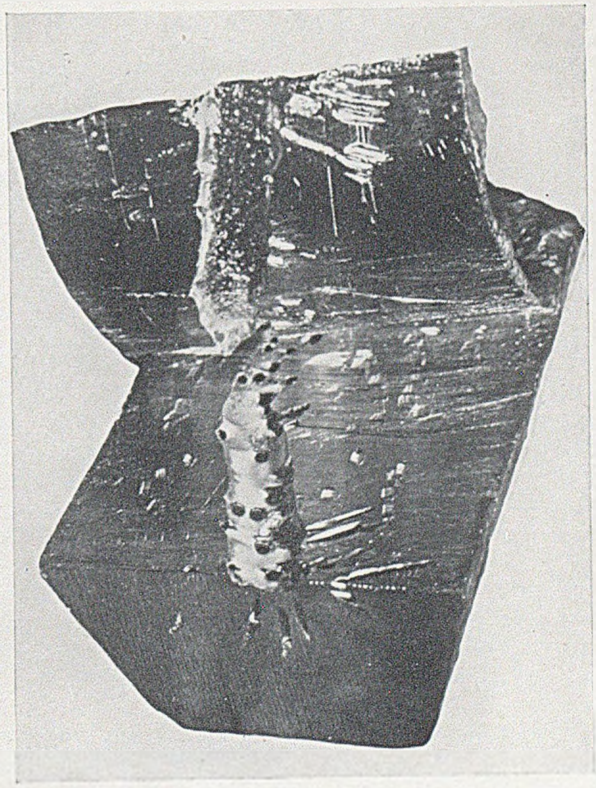

FIG. 7.—Horizontal Section Through the Same Ingot Showing Extensive Piping. × 1.

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III.—CORROSION TEST METHODS

1. Materials

The magnesium used in preparing the corrosion test specimens was made by the ferro-silicon process and its chemical composition was as follows :

Fe, %	P, %	Ni, %	Al, %	Ca, %	Cu, %	Mn, %	Na, %	Pb, %	Si, %	Zn, %
<0·002[a]	Up to 0·005	0·0004	0·005	0·002	0·002	0·003	0·01	0·001	0·002	<0·01

[a] For most of the work magnesium containing less than 0·002% iron was selected. When difficulty was found in preparing samples rich in phosphorus with iron contents below 0·003%, subsequent "blank" and low-phosphorus melts were made from ferro-silicon magnesium with an initial iron content exceeding 0·002%. This appeared to be essential in order to avoid possible confusion of the effects of iron and phosphorus on the corrosion rates.

The aluminium and zinc used were of 99·99% purity. Manganese was added as $MnCl_2$ (iron content 0·2%) or, where small amounts were required, as a magnesium–manganese hardener alloy made from $MnCl_2$. Phosphorus was added either directly as A.R. quality $Ca_3(PO_4)_2$ mixed with Melrasal Z flux or as a hardener alloy with magnesium. Dephosphorization, where required, was carried out by means of a zirconium–aluminium master alloy or, in a few cases, with rare-earth metals.

2. Melts

The following series of virgin melts with varying phosphorus contents were prepared in graphite crucibles on a 10-lb. scale, the phosphorus being added either directly as $Ca_3(PO_4)_2$ or as a hardener alloy made in graphite crucibles :

> Series *A* : Magnesium with low iron and varying manganese content.
>
> Series *B* : Magnesium with at least 0·01% iron[†] and 0·2% manganese.
>
> Series *C* : High-purity Elektron A8.

In addition, a further series was made as follows :

> Series *D* : Remelts of high-purity A8 (prepared on a 250-lb. scale in steel crucibles) with phosphorus added as a hardener alloy made in steel crucibles.

3. Specimens

Melts were cast into horizontal chill bar moulds of "keyhole" section to ensure maximum soundness and uniformity of composition of the three specimens into which each bar was machined. The analyses quoted relate to metal immediately adjacent to that from which the test specimens were taken. The analytical methods used are described elsewhere.[7]

Cylindrical specimens, approximately 0·7 in. dia. × 1½ in. long, were turned from the bar and cleaned with moist pumice powder before immersion in the corrosion bath. This consisted of 3% NaCl solution saturated with $Mg(OH)_2$. Cleaning of specimens after removal from the bath was effected by thorough washing with water and immersion in boiling chromic acid saturated with Ag_2CrO_4. By this means it was possible to remove practically all the corrosion product without dissolving any metal.

The immersion period was 28 days in all cases, and wherever possible all the specimens in one series were corroded at the same time in one bath. Where more than one bath was employed, the specimens in each were representative of the whole composition range of the series. To check that comparable corrosion rates were obtained irrespective of the bath used, separate tests were made on specimens all cut from the same piece of Elektron ZW3 sheet, and also on specimens cut from cast ZW3 slab made by the direct-chill process and known, from previous work, to have uniform corrosion rates. The average results on six pieces of sheet in each bath obtained in three separate and successive tests, and on four pieces of the slab in a test made many months earlier, are given in Table I. The table also gives the maximum and minimum values for the specimens in each bath, showing the degree of variation commonly found between nominally identical specimens.

Table I.—*Variations in Corrosion Rate of ZW3 Alloy Due to Bath Used, and Range of Corrosion Rates among Similar Specimens.*

Bath No.	Sheet			Direct-Chill-Cast Slab		
	No. of Specimens	Corrosion Rate, mg./cm.²/day		No. of Specimens	Corrosion Rate, mg./cm.²/day	
		Average	Max. and Min.		Average	Max. and Min.
1	6	0·335	0·431–0·276	4	0·167	0·249–0·135
2	6	0·408	0·472–0·314	4	0·197	0·262–0·124
3	6	0·264	0·313–0·204	4	0·170	0·190–0·142
4	6	0·263	0·294–0·224	4	0·145	0·156–0·129

IV.—DISCUSSION OF CORROSION TESTS

The results of the corrosion tests are given in Tables II–V and Figs. 8–10. Although the results on individual melts exhibit wide scatter, plots of corrosion rate against phosphorus content for each series disclose, in all cases except that of magnesium containing over 0·2% manganese, tendencies for the corrosion rate to increase with phosphorus content. These trends, shown on a single diagram in Fig. 10, are also evident from inspection of the mean results for each of the groups into which the tables have for convenience been divided (see also Fig. 9).

[†] Electrolytic magnesium contains at least 0·01% iron. Remelting of pure ferro-silicon magnesium once or twice in the usual steel crucibles increases the iron content to a similar figure. The solubility of iron in magnesium appears to be about 0·025% at 650° C.[4-6]

TABLE II.—*Effect of Phosphorus on the Corrosion Rate of Magnesium with Varying Manganese Content (Series* A).

Virgin melts in 10-lb. graphite crucibles.

Range of Manganese Content, %	Range of Phosphorus Content, %	Melt No.	Chemical Analysis			Corrosion Rate, mg./cm.²/day	
			Fe, %	Mn, %	P, %	Mean of 3 Specimens	Range
0–0·05	<0·002	A1	0·004	0·004	0·001 [*]	0·117	0·096–0·155
		A2	0·005	0·027	<0·001	0·226	0·146–0·314
		A3	0·003	0·005	0·001	0·128	0·117–0·135
		A4	0·003	0·004	0·001 [*]	0·199	0·154–0·234
		A5	0·006	0·006	0·001	0·396	0·348–0·423
		A6	0·006	0·004	0·001	0·082	0·062–0·103
		Mean	0·0045	0·008	0·001	0·19	
	0·002–0·01	A7	0·003	0·005	0·002 [*]	4·790	4·48 –5·15
		A8	0·005	0·032	0·002 [*]	0·388	0·276–0·501
		A9	0·002	0·040	0·003 [*]	1·997	1·845–2·221
		A10	0·002	0·004	0·008 [*]	5·100	4·67 –5·52
		A11	0·005	0·028	0·007	0·590	0·572–0·612
		Mean	0·0034	0·022	0·0044	2·57	
	>0·01	A12	0·002	0·050	0·014	1·547	1·355–1·793
		A13	0·003	0·005	0·045	0·316	0·298–0·343
		A14	0·004	0·029	0·031	5·480	5·03 –5·92
		A15	0·004	0·005	0·040	0·700	0·503–0·812
		A16	0·004	0·004	0·090	0·371	0·327–0·454
		Mean	0·0034	0·019	0·044	1·67	
0·051–0·2	<0·002	A17	0·004	0·084	<0·001	0·119	0·103–0·137
		A18	0·002	0·160	<0·001	0·083	0·080–0·086
		A19	0·002	0·150	<0·001 [*]	0·054	0·051–0·058
		Mean	0·0027	0·131	<0·001	0·085	
	0·002–0·01	A20	0·003	0·080	0·003 [*]	1·442	1·315–1·561
		A21	0·002	0·140	0·002	0·082	0·068–0·090
		A22	0·004	0·170	0·003 [*]	0·771	0·716–0·853
		Mean	0·003	0·130	0·0026	0·77	
	>0·01	A23	0·003	0·070	0·012	0·326	0·275–0·372
		A24	0·003	0·100	0·015	2·270	1·969–2·572
		A25	0·004	0·085	0·020	0·178	0·158–0·205
		A26	0·003	0·090	0·099	0·318	0·268–0·343
		A27	0·003	0·070	0·091	0·198	0·194–0·201
		A28	0·003	0·090	0·052	0·300	0·211–0·429
		A29	0·005	0·085	0·066	1·135	0·954–1·429
		A30	0·003	0·140	0·016	3·660	1·99 –6·08
		A31	0·003	0·140	0·061	0·168	0·167–0·171
		Mean	0·0033	0·097	0·043	0·95	
>0·2	<0·002	A32	0·006	0·300	<0·001 [*]	0·385	0·348–0·432
		A33	0·004	0·650	<0·001 [*]	0·170	0·153–0·181
		A34	0·004	0·310	0·001	0·084	0·060–0·098
		Mean	0·0047	0·420	<0·001	0·21	
	0·002–0·01	A35	0·002	0·410	0·002 [*]	0·142	0·106–0·197
		A36	0·003	0·850	0·002 [*]	0·075	0·041–0·101
		A37	0·004	0·660	0·002	0·153	0·140–0·172
		A38	0·004	0·330	0·004 [*]	0·355	0·318–0·393
		A39	0·004	0·610	0·004 [*]	0·210	0·197–0·230
		Mean	0·0034	0·572	0·0028	0·19	
	>0·01	A40	0·004	0·220	0·016	0·614	0·402–0·900
		A41	0·004	0·300	0·015	0·079	0·062–0·089
		A42	0·006	0·650	0·033	0·083	0·057–0·107
		A43	0·003	0·290	0·065	0·213	0·162–0·273
		A44	0·006	0·660	0·068	0·061	0·031–0·080
		Mean	0·0046	0·424	0·039	0·21	

[*] Prepared from ferro-silicon metal after dephosphorization with zirconium-aluminium alloy.

1. MAGNESIUM AND MAGNESIUM-MANGANESE ALLOY (SERIES *A* AND *B*)

From the results for Series *A* melts (Table II and Figs. 9 (*a*)–(*c*)) the following tentative conclusions can be drawn :

(i) Phosphorus has an adverse effect on the corrosion-resistance of pure magnesium.

(ii) The "tolerance limit" for phosphorus is about 0·002% (Fig. 8).

(iii) Addition of manganese suppresses the effect of phosphorus.

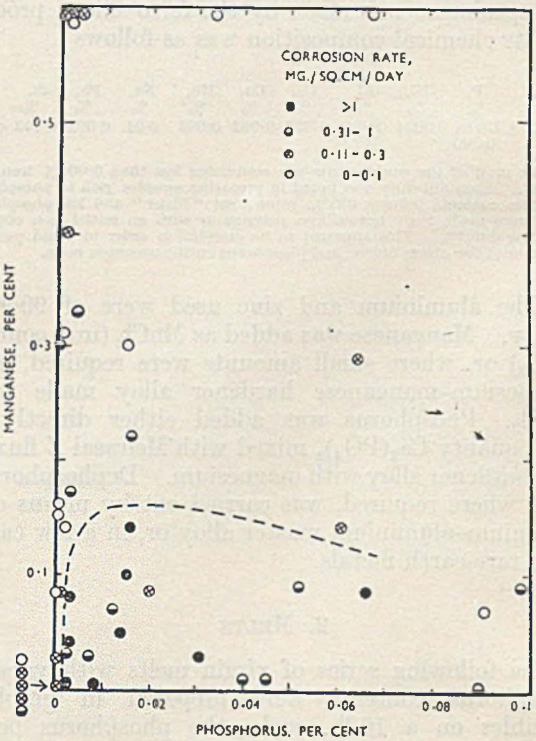

FIG. 8.—Effect of Phosphorus and Manganese on the Corrosion-Resistance of Magnesium.

The point nearest the origin represents the five melts indicated in the margin, which could not be plotted separately in the available space.

All melts with a corrosion rate exceeding 1 mg./cm.²/day lie within the marked area.

TABLE III.—*Effect of Phosphorus on the Corrosion-Resistance of Magnesium Containing at Least* 0·01% *Iron and* 0·2% *Manganese* (*Series* B).

Virgin melts in 10-lb. mild-steel crucibles.

Range of Phosphorus Content, %	Melt No.	Chemical Analysis			Corrosion Rate, mg./cm.²/day	
		Fe, %	Mn, %	P, %	Mean of 3 Specimens	Range
<0·005	B1	0·06	0·29	<0·002	0·601	0·551– 0·673
	B2	0·06	0·29	<0·002	1·185	1·042– 1·258
	B3	0·01	0·41	0·003	0·230	0·178– 0·330
	Mean	0·043	0·33	0·002	0·67	
0·01–0·025	B4	0·015	0·36	0·012	1·43	0·610– 2·32
	B5	0·024	0·33	0·023	1·68	1·46 – 2·03
	B6	0·015	0·25	0·023	1·79	1·500– 1·975
	B7	0·024	0·33	0·023	2·45	1·951– 3·405
	Mean	0·020	0·32	0·02	1·84	
>0·025	B8	0·015	0·21	0·033	5·07	2·162– 9·250
	B9	0·02	0·28	0·06	7·45	1·85 –10·89
	B10	0·02	0·28	0·06	12·20	10·67 –13·78
	B11	0·018	0·23	0·08	2·19	1·73 – 2·46
	B12	0·018	0·24	0·17	2·13	1·68 – 2·60
	B13	0·018	0·20	0·19	11·10	8·42 –13·15
	Mean	0·018	0·24	0·098	6·67	

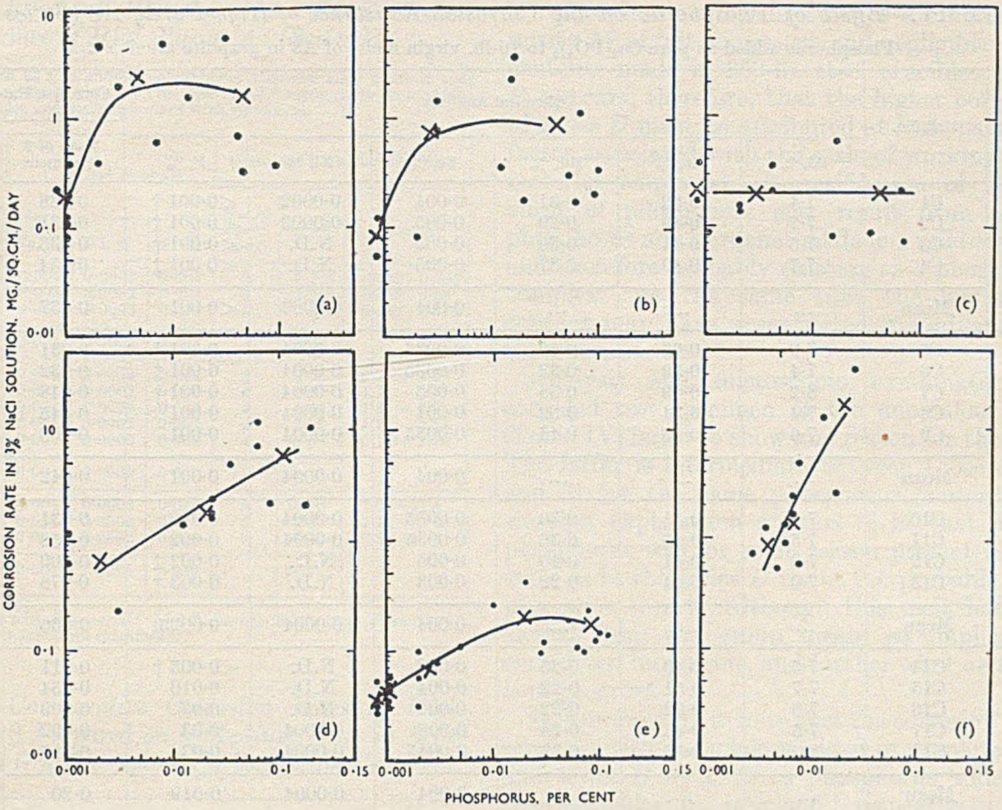

Fig. 9.—Effect of Phosphorus Content on Corrosion-Resistance.

(*a*)–(*c*) Series *A* : Magnesium containing (*a*) $<0·05\%$ Mn; (*b*) 0·05–0·2% Mn; (*c*) $>0·2\%$ Mn.
(*d*) Series *B* : Magnesium containing at least 0·01% Fe and 0·2% Mn.
(*e*) Series *C* : High-purity A8 (phosphorus added as pure $Ca_3(PO_4)_2$).
(*f*) Series *D* : High-purity A8 (phosphorus added as hardener made in steel crucible).

● Individual melts. × Grouped mean results.

The Series *B* results (Table III and Fig. 9 (*d*)) further indicate :

(iv) That when the iron content is " normal ", phosphorus exerts an adverse effect even with $\frac{1}{4}$% manganese present.*

It will be seen that, despite the use of graphite crucibles and tools, some iron pick-up occurred during preparation of the series *A* melts, five melts in particular containing as much as 0·006% iron. None of these shows a high corrosion rate, and it is evident that iron pick-up does not invalidate the results by itself contributing to the observed deterioration in corrosion-resistance which accompanies the introduction of phosphorus. This is in accordance with the high tolerance limit of pure magnesium for iron (0·017%) found by Hanawalt, Nelson, and Peloubet,[1] and the observation by Fox, Bushrod, and Mayer [8] that the corrosion-resistance of Elektron AM503 alloy (magnesium–1·5% manganese) is independent of the iron content.

The mean figure of 0·19 mg./cm.^{2}/day found for magnesium of low iron, manganese, and phosphorus contents accords well with Hanawalt's figure of 0·15 mg./cm.^{2}/day for pure magnesium, obtained by intermittent as opposed to total immersion in salt solution. In general, however, the two forms of test are not dissimilar in effect,[9] and the values are found to be comparable, probably because of the short time of immersion in Hanawalt's test, which is too short to permit much draining, still less drying, of the specimen.

In view of the variable phosphorus content of the ferro-silicon magnesium available for the tests (up to 0·005%), some of the melts were prepared from magnesium dephosphorized with zirconium–aluminium alloy, to assist in the control of phosphorus content in the lower ranges. These melts are indicated in Table II. The evidence available indicates that the dephosphorization process is without appreciable effect on the corrosion rate for a given phosphorus content. This has been verified on a large scale, a rate of 0·16 mg./cm.^{2}/day † being obtained for magnesium so dephosphorized, in agreement with the figure of about 0·2 mg./cm.^{2}/day found for the un-

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* Addition of several per cent. of manganese might, of course, suppress the effect of phosphorus in this case also.

† Mean of eight results ranging from 0·12 to 0·18 mg./cm.^{2}/day.

Table IV.—*Effect of Phosphorus on the Corrosion-Resistance of High-Purity A8 (Series* C).

Phosphorus added as pure $Ca_3(PO_4)_2$ to 10-lb. virgin melts of A8 in graphite crucibles.

Range of Phosphorus Content, %	Melt No.	Chemical Analysis						Corrosion Rate, mg./cm.²/day	
		Al, %	Zn, %	Mn, %	Fe, %	Ni, %	P, %	Mean of 3 Specimens	Range
<0·001	C1	7·4	0·52	0·31	0·004	0·0002	<0·001 †	0·028	0·025–0·031
	C2	7·7	0·49	0·29	0·005	0·0003	<0·001 ‡	0·031	0·024–0·036
	C3	7·5	0·41	0·30	0·003	N.D.	<0·001 †	0·035	0·026–0·047
	C4	7·7	0·49	0·37	0·005	N.D.	<0·001 ‡	0·054	0·039–0·067
	Mean	...	...	...	0·004	0·0003	<0·001	0·037	
0·001	C5	7·9	0·62	0·40	0·0035	0·0004	0·001 †	0·031	0·024–0·039
	C6	7·4	0·59	0·33	0·0035	0·0004	0·001 ‡	0·032	0·026–0·039
	C7	8·2	0·49	0·38	0·005	0·0004	0·001 †	0·048	0·038–0·057
	C8	7·4	0·54	0·32	0·004	0·0004	0·001	0·045	0·035–0·061
	C9	7·6	0·55	0·45	0·0035	0·0004	0·001	0·056	0·040–0·075
	Mean	...	...	...	0·004	0·0004	0·001	0·042	
0·002–0·003	C10	7·4	0·51	0·36	0·0035	0·0004	0·002	0·031	0·031–0·032
	C11	7·5	0·45	0·36	0·0035	0·0004	0·002	0·056	0·045–0·077
	C12	7·2	0·41	0·40	0·005	N.D.	0·002 ‡	0·100	0·068–0·118
	C13	7·0	0·44	0·22	0·003	N.D.	0·003 †	0·078	0·065–0·105
	Mean	...	...	...	0·004	0·0004	0·0025	0·066	
0·005–0·03	C14	7·5	0·49	0·25	0·002	N.D.	0·005 †	0·111	0·067–0·162
	C15	7·7	0·51	0·22	0·004	N.D.	0·010	0·254	0·197–0·329
	C16	7·5	0·52	0·22	0·004	N.D.	0·02	0·409	0·187–0·531
	C17	7·5	0·48	0·28	0·005	0·0004	0·03	0·092	0·082–0·103
	C18	8·0	0·57	0·33	0·0045	0·0004	0·03	0·115	0·089–0·156
	Mean	...	...	...	0·004	0·0004	0·019	0·20	
>0·03	C19	7·7	0·70	0·35	0·005	0·0004	0·05	0·186	0·090–0·322
	C20	9·0	0·57	0·40	0·005	0·0004	0·06	0·103	0·046–0·202
	C21	8·8	0·72	0·48	0·005	N.D.	0·07	0·095	0·060–0·114
	C22	7·0	0·52	0·36	0·005	N.D.	0·08	0·224	0·198–0·263
	C23	7·2	0·32	0·31	0·005 *	0·0002	0·08	2·75	2·48 –3·03
	C24	7·8	0·55	0·39	0·005	N.D.	0·09	0·118	0·097–0·148
	C25	4·3	0·31	0·47	0·005	N.D.	0·10	0·139	0·126–0·154
	C26	7·8	0·50	0·25	0·004	N.D.	0·12	0·131	0·090–0·162
	Mean	...	...	...	0·005	0·0004	0·081	0·14 §	

N.D. = Not determined.
* Confirmed by analyses on the test specimen after corrosion.
† Melts dephosphorized with zirconium–aluminium alloy.
‡ Melts dephosphorized with Mischmetall (rare-earth metals of the cerium group).
§ Excluding Melt C23, the corrosion rate of which appeared to be anomalous.

treated pure magnesium. Table IV suggests that this dephosphorization process is without appreciable effect in the case of high-purity A8 also.

The fact that when the iron content is high the presence of ¼% manganese does not suppress the adverse effect of phosphorus, appears to confirm earlier indications of the possible existence in magnesium of some iron–phosphorus complex.

2. Elektron A8 (Series *C* and *D*)

Results from series *C* (Table IV and Fig. 9 (*e*)) show a broad correlation of corrosion rate with phosphorus content, but none with that of any of the other constituents over the composition range covered. Neither of the processes employed for removing phosphorus appears to have affected the corrosion-resistance appreciably. (One melt with an anomalously high corrosion rate (2·75 mg./cm.²/day) has been excluded in calculating the mean rate for the appropriate group and also in drawing Fig. 9 (*e*); this is discussed later.)

The corrosion rates for the low-phosphorus high-purity A8 average less than one-fifth that of pure magnesium and are therefore very low indeed. Since the iron contents of these specimens extend to 0·005%, it must be assumed that the tolerance limit for iron in magnesium containing 8% aluminium is considerably higher than the figure of 0·002% given by Hanawalt, Nelson, and Peloubet,[1] at least in the presence of a few tenths per cent. of zinc. They reported that 0·5% zinc was without effect on the position of the iron tolerance limit but that 3% zinc raised it to 0·003%.

Corrosion rates from series *D* (Table V and Fig. 9 (*f*)) also show a broad correlation with phosphorus

TABLE V.—*Effect of Phosphorus on the Corrosion-Resistance of High-Purity A*8 (*Series* D).

Phosphorus added as a magnesium–phosphorus hardener alloy made in steel crucibles to 10-lb. melts in graphite crucibles of A8 ° ingot made on a 250-lb. scale in steel crucibles and dephosphorized with zirconium.

Range of Phosphorus Content, %	Melt No.	Chemical Analysis			Corrosion Rate, mg./cm.^{2}/day		
		Fe, %	Ni, %	P, %	No. of Specimens	Mean	Range
0–0·005	D1	0·004	0·0002	0·003	3	0·745	0·675– 0·705
	D2	0·004	0·0002	0·004	2	1·298	0·655– 1·94
	D3	0·003	0·0002	0·005	3	0·544	0·410– 0·718
	Mean	0·004	0·0002	0·004		0·86	
0·006–0·01	D4	0·004	0·0002	0·006	2	0·90	0·726– 1·09
	D5	0·003	0·0002	0·006	4	1·40	1·18 – 1·67
	D6	0·004	0·0002	0·007	5	1·69	1·25 – 1·98
	D7	0·004	0·0002	0·007	2	2·55	1·71 – 3·38
	D8	0·003	0·0003	0·008	4	0·59	0·151– 1·30
	Mean	0·004	0·0002	0·007		1·43	
>0·01	D9	0·004	0·0002	0·013	5	13·10	11·9 –11·4
	D10	0·004	0·0003	0·018	5	2·67	2·14 – 3·60
	D11	0·004	0·0003	0·026	5	37·20	36·2 –38·6
	Mean	0·004	0·0003	0·019		17·7	

° Composition of parent material: Al 7·6, Zn not determined (nominal 0·4), manganese 0·08%, magnesium remainder.

TABLE VI.—*Comparative Corrosion Rates of High-Purity A*8 *Prepared on a Medium Scale in Mild-Steel and Graphite-Lined Carborundum Crucibles.*

Crucible		Melt No.	Chemical Analysis			Corrosion Rate, mg./cm.^{2}/day †
Material	Scale, lb.		Fe, %	Ni, %	P, %	
Mild steel	250	251B2	0·002	0·0005	<0·001	0·78
		420B	<0·002	0·0009	<0·002	0·98
		434B	0·001	0·0005	0·001	1·16
		469B	0·001	0·0006	<0·001	0·30
		470B	0·001	0·0003	<0·001	0·37
		505B	<0·001	0·0008	<0·001	0·22
		579B	<0·001	0·0007	<0·001	0·17
" Tercod " (graphite-lined carborundum)	100	554B	0·001	0·0004	0·002	0·049
		498B	0·0015	0·0003	0·002 °	0·078

° Melt dephosphorized with zirconium–aluminium alloy.
† Mean of 4 specimens.

content, but are much higher than those of series *C* for the same composition. One reason for these higher results might at first sight be taken to be the lower manganese content (0·08%) of the parent material, removal of some manganese having accompanied the dephosphorization process. On the other hand, Hanawalt, Nelson, and Peloubet[1] state that " a few hundredths per cent." of manganese is sufficient to raise the very low iron tolerance limit observed in the presence of aluminium to 0·002% for any aluminium content; and in the present work very low corrosion rates have been obtained with remelted dephosphorized high-purity A8 containing only 0·09% manganese.* Moreover, the corrosion rates for the low-phosphorus alloys of Table V are normal for high purity A8 of full manganese and negligible phosphorus contents made in 250-lb. steel crucibles (Table VI). It appears, therefore, that the higher corrosion rates of series *D* must be attributed at least partly to some factor associated with the scale of working or crucible material employed. Intensification of the adverse effect of phosphorus may result from addition of phosphorus as a hardener made in a mild-steel crucible and therefore probably existing as a phosphorus–iron complex. At the same time the low manganese content may have accentuated the adverse effect of the phosphorus.

Analysis and microscopic examination of the corroded test specimen of the anomalous melt C23 (Table IV) failed to show any reason for the high rate. The latter is intermediate between those of series *C* and *D* for the same phosphorus content, and the simplest explanation appears to be that part of the phosphorus was for some reason present in the same state as in the series *D* melts, i.e. probably as a complex with iron.† Although this melt has been excluded from the group mean, its implications are considered important, and further work on the matter is in progress.

It is interesting to note that the corrosion rates from series *C* (with phosphorus added in graphite crucibles) and series *D* melts (with phosphorus added in mild-steel crucibles) lie, in general, respectively below and above the figure of 0·2 mg./cm.^{2}/day representing pure magnesium. A similar difference is found between the corrosion rates of high-purity A8 made in graphite and steel crucibles when a medium scale is employed (Table VI).

3. ALLOYS CONTAINING ZIRCONIUM

As already indicated, zirconium is a precipitant for phosphorus, and in commercial alloys containing zirconium the phosphorus content is below 0·005%. No indications have been obtained of any effect of phosphorus on the corrosion-resistance of alloys containing zirconium. This is to be expected, since the phosphorus is present in combination with zirconium in the form of insoluble particles and is thus analogous to iron, which in zirconium alloys is present as zirconium-rich particles and does not appreciably affect corrosion-resistance.

V.—EFFECT OF PHOSPHORUS ON IRON PICK-UP IN THE REMELTING OF A8 IN MILD-STEEL CRUCIBLES

In view of the existence of a tolerance limit for iron in high-purity A8 alloy, the question of iron pick-up on remelting the alloy in the usual mild-steel crucibles is one of considerable practical importance.

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* Corrosion rate 0·055 mg./cm.^{2}/day (mean of four results ranging from 0·018 to 0·097 mg./cm.^{2}/day).

† The abnormally high degree of scatter in corrosion rate shown by melts (as distinct from individual test specimens) in the present investigation, and, in particular, the tendency for scatter to increase with increasing phosphorus content, may also be attributable to variations in the state of the phosphorus.

The subject has been studied by Fox, Bushrod, and Mayer,[8] who called attention to the existence of a " seasoning " effect, as a result of which iron pick-up decreases on successive remeltings in the same crucible until, after two or three melts have been made, a roughly constant low iron pick-up occurs. This seasoning of the surface of the crucible is accompanied by decarburization and pick-up of aluminium and magnesium in the surface layers.* Further work has shown that the seasoned surface deteriorates if the crucible is allowed to stand empty for a day or two, probably owing to the absorption of moisture by residual flux adhering to the crucible walls. Washing the crucibles also destroys the " seasoning ". Even with seasoned crucibles, occasional melts high in iron are obtained at temperatures around 800° C. and above. This is no doubt due to local flaking off of the seasoned surface, leaving an area of unseasoned steel exposed to the magnesium.

In view of the probable existence of phosphorus as a complex with iron and the difficulty found in preparing high-phosphorus alloys with low iron content, experiments were made on the effect of phosphorus on iron pick-up in well-seasoned mild-steel crucibles. The results (Table VII) show that the effect is indeed pronounced.

Table VII.—*Effect of Phosphorus on Iron Pick-Up in Remelting A8 in Seasoned Mild-Steel Crucibles.*

(10-lb. Scale)

Phosphorus, %	Melting Cycle	Useful Life of Seasoned Surface *	Average Iron Pick-Up per Remelt, %
<0·001	Held for 30 min. at 800° C.	>12	0·0003
0·02	" " "	2	0·001
0·09	" " "	Nil	0·008

* That is, number of consecutive melts with iron pick-up <0·002%.

Table VIII.—*Effect of Remelting on Iron Content of A8.*

Melt No.	Crucible		Chemical Analysis			Iron Pick-Up, %
	Material	Scale, lb.	P, % Initial	Fe, %		
				Initial	Final	
B358 B362 B363 B1000	Mild steel	10	0·001 0·001 0·001 N.D.	0·003 0·004 0·004 0·0015	0·003 0·0025 0·0025 <0·001	0 −0·0015 −0·0015 −0·001
B361 B366 B231 B236 B237	Cr–Mn–Al steel	20	0·002 0·001 <0·001 <0·001 <0·001	0·003 0·004 0·003 0·003 0·003	0·002 0·001 0·003 0·003 0·005	−0·001 −0·003 0 0 +0·002
B434	Mild steel	250	0·0015	0·0015	0·0015	0

N.D. = not determined.

Further work on remelting the low-phosphorus high-purity alloy has shown that if, instead of holding the melt at fairly high temperatures for a period, the melt is merely raised to 750° C., stirred for 1 min., and poured after standing for, say, 5 min., the iron content may, in fact, fall. Table VIII illustrates this.

VI.—FINAL CONSIDERATIONS

The work described has shown that in addition to iron and nickel a third element, phosphorus, can exert a pronounced adverse effect on the corrosion-resistance of magnesium when present in trace amounts.

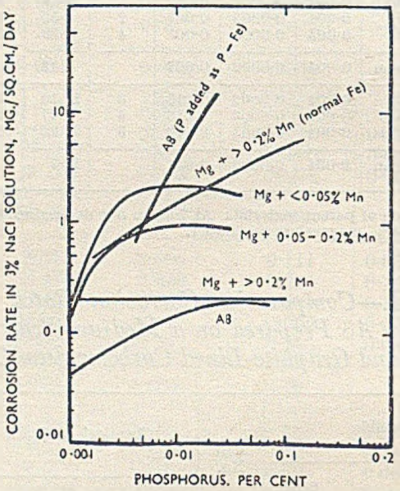

Fig. 10.—Effect of Phosphorus Content on Corrosion-Resistance.

From examination of the individual results, as well as the trends displayed in Fig. 10, it appears that corrosion discontinuities are less marked than with iron and nickel, the phenomena of sharp tolerance limits being not in general evident. The adverse effect of phosphorus shows, however, some analogies with that of nickel :

(1) The most pronounced corrosion discontinuity occurs with magnesium alone.

(2) The adverse effect is suppressed by manganese.

(3) The adverse effect is not suppressed by manganese when aluminium is present.

The precise significance to be attached to the mutual effects of iron and phosphorus in promoting their entry into magnesium (Fig. 1 and Table VII) is not yet clear. It is possible that the solubility of iron in magnesium is increased in the presence of phosphorus and vice versa. On the other hand, some settling out of phosphorus and iron has been observed in the molten alloys, irrespective of whether the

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* This led to the subsequent suggestion of C. J. Bushrod to add aluminium to high-chrome–manganese steel for use in melting magnesium, and subsequent work by J. K. Davies and J. H. T. Petch has shown that such crucibles behave as if they were permanently in a " seasoned " condition.

phosphorus or the iron was introduced first into the magnesium; and this suggests the existence of some insoluble compound present in the molten alloy. The curve of Fig. 1 does not, however, represent a constant iron : phosphorus ratio, and the phosphorus increments, being between five and ten times the iron increments on an atomic basis, do not appear to correspond with likely compounds. Even higher phosphorus : iron ratios have been inferred from comparison of analyses taken from the top and bottom of settled melts. Unfortunately, it has not proved a simple matter to decide from metallographic evidence whether or not phosphorus and iron form a compound insoluble in the molten alloy. The cavities may be gas voids or, as suggested by Vosskühler,[2] they may be formed by the removal of soft particles during the polishing process. With mechanical polishing it is difficult to avoid partially filling cavities with polishing compound, and, with electrolytic polishing, cavities could arise through reaction of phosphide particles with the electrolyte. Very few of the particles described by Bulian and Fahrenhorst[3] and attributed to Mg_3P_2 have been seen in any of the alloys examined.

On the question of the difference in corrosion rates shown by the two series of A8 melts for a given phosphorus content, it has been suggested that this may be largely attributable to the presence of phosphorus in series *D* melts in the form of a complex with iron. If this complex is unstable, or if a supersaturated solution is involved, decomposition might occur on solidification with liberation of iron on subsequent remelting, the liberated iron then settling out under the combined action of aluminium and manganese in the ordinary way. A process of this type would explain the frequent drop in iron content on remelting A8 when the phosphorus content is low. As a consequence it would be expected that the corrosion rate would improve. This has, in fact, been verified and in certain circumstances the remelting of high-purity A8 in steel crucibles has been found to reduce the corrosion rate by a factor of up to one hundred. Such work, however, lies beyond the intended scope of the present paper, the main object of which is to call attention to the adverse effects which phosphorus can exert on the corrosion-resistance of magnesium and its alloys.

ACKNOWLEDGEMENTS

The authors' thanks are due to Mr. W. Unsworth for carrying out most of the melting work, to Mr. W. R. Braithwaite for making some of the corrosion tests, to Messrs. J. B. Wilson and D. R. Knowles for helpful discussions on statistical aspects of the results and for making the majority of the calculations involved, and to the Directors of Magnesium Elektron, Ltd., for permission to publish the paper.

APPENDIX

NOTE ON STATISTICAL ASPECTS OF THE CORROSION TEST RESULTS

Table II (*Series* A : *Effect of Phosphorus on Magnesium of Varying Manganese Content*)

The results do not show a significant * product-moment correlation coefficient (r) between phosphorus content and corrosion rate, nor a significant partial correlation coefficient when the effect of manganese is eliminated. On the other hand, if the true curve of corrosion rate against phosphorus should be sigmoid with a sharp tolerance limit, a significant correlation coefficient would not necessarily be expected. In such a case, however, there should be a significant difference between the test results from samples with phosphorus contents above and below the tolerance limit when tested by the "Student" t test for comparison of means.

If the first occurrence of corrosion rates above 1 mg./cm.^{2}/day is taken as evidence that a tolerance limit has been exceeded (as appears reasonable from a study of the diagram in Hanawalt's[1] paper), then the tolerance limit for phosphorus in pure magnesium will lie between 0·001 and 0·002% (Fig. 8). Applying this tolerance limit to all the melts containing less than 0·2% manganese (Figs. 9 (*a*) and (*b*)), the corrosion test results on melts with higher and lower phosphorus contents are found to differ significantly ($p = 0{\cdot}02$). The same applies to the melts containing less than 0·05% manganese (Fig. 9 (*a*)), the lower degree of significance ($p = 0{\cdot}05$) being due to the smaller number of melts falling below the tolerance limit.

Treated separately, the melts with intermediate manganese content (Fig. 9 (*b*)) indicate a slightly higher phosphorus tolerance limit between 0·002 and 0·003%, and division of the group here yields a significant difference ($p = <0{\cdot}001$). Hanawalt, Nelson, and Peloubet[1] found that the nickel tolerance limit was raised by manganese, and, in view of the general resemblance in behaviour between nickel and phosphorus already noted, the same might apply to the phosphorus tolerance limit.

With manganese content greater than 0·2% (Fig. 9 (*c*)), corrosion rates above 1 mg./cm.^{2}/day were not obtained, and at no point is it possible to divide the melts into two groups so as to yield a significant t test. (With a division at 0·0015% phosphorus $p = 0{\cdot}9$.)

Table III. (*Series* B : *Effect of Phosphorus on Magnesium Containing at Least* 0·01% *Iron and* 0·2% *Manganese*)

The correlation coefficient between phosphorus and corrosion rate is not significant ($p = 0{\cdot}1$). If the results are grouped into those with phosphorus content above and below 0·025%, so as to obtain as nearly as possible groups of equal size, application of

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* The level of significance adopted is a chance probability p of 0·05, i.e. a 1 in 20 possibility that the observed distribution is due to chance.

the t test shows a significant difference in the means ($p = <0{\cdot}01$).

Table IV. (Series C: Effect of Phosphorus on A8)

If, for the reasons discussed on p. 29, melt C23 is excluded, the partial correlation coefficient between phosphorus content and corrosion rate with the effect of manganese eliminated is just significant ($p = 0{\cdot}05$).

Table V. (Series D: Effect of "Phosphorus-Iron" on A8)

The correlation coefficient between phosphorus content and corrosion rate is significant ($p = <0{\cdot}01$). In this case the analytical values obtained for iron and nickel are sufficiently constant to make the derivation of the second-order partial correlation coefficient appear unnecessary.

REFERENCES

1. J. D. Hanawalt, C. E. Nelson, and J. A. Peloubet, *Trans. Amer. Inst. Min. Met. Eng.*, 1942, **147**, 273.
2. H. Vosskühler, see "Technology of Magnesium and Its Alloys" (A. Beck), p. 45. London: **1940** (F. A. Hughes and Co., Ltd.).
3. W. Bulian and E. Fahrenhorst, "Metallography of Magnesium and Its Alloys", p. 31. London: **1944** (F. A. Hughes and Co., Ltd.).
4. E. Fahrenhorst and W. Bulian, *Z. Metallkunde*, 1941, **33**, 31.
5. G. Siebel, *Z. Metallkunde*, 1948, **39**, 22.
6. D. W. Mitchell, *Trans. Amer. Inst. Min. Met. Eng.*, 1948, **175**, 570.
7. "Methods of Analysis of Magnesium and Its Alloys". London: **1946** (F. A. Hughes and Co., Ltd.).
8. F. A. Fox, C. J. Bushrod, and S. E. Mayer, *J. Inst. Metals*, 1947, **73**, 55.
9. L. F. Le Brocq, *J. Inst. Metals*, 1944, **70**, 594 (discussion).

CREEP AND STRESS RUPTURE AS RATE PROCESSES *

1322

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SYNOPSIS

Creep data for an iron–cobalt–chromium–nickel alloy reported by Grant and Bucklin (*Trans. Amer. Soc. Metals*, 1950, 42, 720) are analysed according to the rate-process theory of plastic flow. The data indicate that the theory can be applied over only a limited range of creep rates.

An empirical equation, which relates the applied stress and the temperature to the minimum creep rate, is suggested for the analysis of creep data. This equation is valid only in the absence of structural instabilities.

I.—INTRODUCTION

The phenomenon of creep in metals has long been of great scientific and practical interest. The literature shows that much effort has been expended in analysing each curve and in relating families of curves to establish fundamental relationships between the variables involved in creep. The primary scientific aim has been to obtain a better understanding of the mechanism of creep, whereas the aim of the engineer has been the prediction of strain, strain rate, and life as a function of temperature, stress, and history of the material.

One of the recent important basic studies has considered the flow of metals to be a deformation process which takes place when a " unit of flow " is excited to an activated state. Such units of flow consist of groups of atoms which pass over one another by a shear process. The size of each unit of flow is a characteristic of the material and depends also on the temperature of testing.

The " theory of activated complex " or " rate-process theory " was first proposed by Eyring [1] and was applied to chemical reactions only. It was later applied to studies of viscosity, plasticity, and diffusion,[2] and subsequently by Kauzmann [3] and Dushman, Dunbar, and Huthsteiner [4] to the flow of solid metals. According to this theory, the rate of shear is proportional to the rate at which the unit of flow crosses a free-energy barrier, which is a function of the applied shear stress. By introducing several simplifying assumptions, the following ultimate equation is obtained :

$$r = Me^{-\Delta F/RT} \sinh (Ns) \quad . \quad . \quad . \quad . \quad (1)$$

where

$r =$ minimum creep rate.
$s =$ applied tensile stress.
$T =$ absolute temperature.
$\Delta F =$ free energy of activation.
$R =$ gas constant per mole.
M, N are constants, at constant temperature.

Since

$$\sinh (Ns) \cong \frac{e^{Ns}}{2}$$

when Ns is sufficiently large, equation (1) can be written :

$$\log r = \log r_0 + Ps \quad . \quad . \quad . \quad (2)$$

where r_0 and P are constants at constant temperature.

A further contribution to the study of flow of metals was made by Nowick and Machlin,[5] who derived a new equation for the steady-state rate of creep through the use of the rate-process theory and of the dislocation theory.

In an effort to find a practical application of the rate-process theory, Machlin and Nowick [6] developed an equation from stress-rupture tests which relates the rupture life to the stress and the temperature. This equation is based upon the assumption that the rupture life is inversely proportional to the rate at which the unit of flow crosses the energy barrier. Their suggested analysis of the data was intended to permit both extrapolation and interpolation to longer test times and higher test temperatures.

One of the more important observations common to all these applications of the rate-process theory is that, according to equation (2), a straight line should be obtained when the applied tensile stress is plotted against the logarithm of the minimum creep rate or against the logarithm of the rupture life for each constant temperature of testing. Further speculation, including the calculation of the " size of unit flow " and the calculation of the " free energy " and " entropy of activation ", was based upon the applicability of the creep law as expressed by equation (2).

It is interesting to note that the rate-process theory is consistent with the empirical law suggested by Ludwick.[7] Such semi-log relationships are not in agreement with the observed straight-line relationships between stress and minimum creep rate when the data are plotted on a log/log basis. Since considerable experimental evidence has accumulated over

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a long period of time that such straight-line log/log plots are real, it is difficult to reconcile theory and test data.

The purpose of the present paper is to demonstrate that the creep law expressed by equation (2) cannot be applied when the creep-rate data are extended over a considerably wider range of creep rates than has previously been examined. The creep-rate data recently published by Grant and Bucklin[8] for the S-590 * alloy will be used for this purpose, since these data are much more extensive than any other appearing in the literature.

Before presenting a detailed analysis of these data, a few general considerations should be discussed. Most creep data are reported in the literature in a graphical form. Calculations based on these plots can be greatly affected by the type of co-ordinates employed. It is suggested that the "range of accuracy" be plotted around each experimental point before deciding the position of the "best" experimental curve, especially when a large number of test points are not available. This is of the greatest importance when logarithmic co-ordinates are employed. It can be demonstrated that if a variable x is plotted on a logarithmic scale, a range of accuracy $\pm ax$ plots on that scale as $\pm \log(1 + a)$. Although the range of accuracy is proportional to the variable, the graphical representation of this range is independent of the value of the variable. On the other hand, the graphical indication of constant uncertainties depends on the value of the variable.

The graphical analysis of creep data according to the rate-process theory may lead to rather uncertain results when the data present a considerable scatter. If the graphical method suggested by Dushman *et al.*[4] is used, the location of the straight line drawn in the semi-log plot of stress against creep rate has a great effect in the determination of the extrapolated intercept at zero stress. By re-examination of the data for high-purity aluminium presented by Dushman *et al.*[4] in graphical form, it was possible to calculate an activation energy Q_1 whose values lay between 5000 and 20,000 cal., according to the way in which the best lines were plotted through the limited data on the semi-log basis.

Furthermore, for best results, the analysis of creep data must be restricted to data from samples which show, as far as can be determined metallographically and otherwise, a stable structure and a similar type of deformation. With this in mind, the analysis of the creep data for S-590 alloy will be restricted to that range in which the type of deformation is determined to be high-temperature behaviour,[8] and in which oxidation and other possible sources of instability have no marked effect. It is worth bearing in mind that oxidation was minimized through the addition of 20% chromium, and that the effect of other structural instabilities was greatly reduced by solution-treating and pre-ageing the test-bars for stability.

II.—ANALYSIS OF THE CREEP DATA FOR S-590 ALLOY

The original data (Grant and Bucklin,[8] Table II), from which only the high-temperature-behaviour points were taken, were converted into C.G.S. units, and the logarithm of the minimum creep rate was plotted against the stress. It was observed that straight lines could not be drawn through the experimental points. The trend of the points follows a family of smooth curves, whose slope decreases as the strain rates increase (see Fig. 5). Within a given range of strain rates, the slope of the curves decreases as the temperature decreases.

It was first thought that, although straight lines could not be drawn on a semi-log plot, the experimental data might still be consistent with the rate-process theory, if they fitted the hyperbolic sine equation (1). In order to resolve this uncertainty, a hyperbolic sine curve was drawn near the experimental curve at 922° K. (1200° F.). Fig. 1 shows that the deviation of the theoretical hyperbolic sine curve from a straight line starts at a much lower stress than the deviation of the experimental curve. The same procedure was applied to other curves, with similar results.

Although the experimental data indicated that a semi-log relationship between stress and minimum creep rate cannot be applied to the behaviour of the

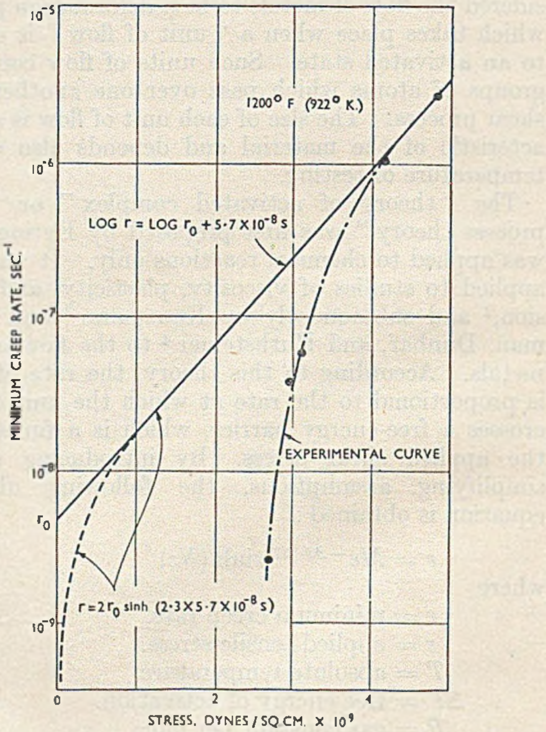

Fig. 1.—Plot of Logarithm of Minimum Creep Rate (r) against Stress (s) at 1200° F. (922° K.), Comparing the Experimental Curve and the Theoretical Hyperbolic Sine Curve.

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* 20% Co, 20% Cr, 20% Ni, 4% W, 4% Mo, 4% Nb, 0·5% C, balance Fe.

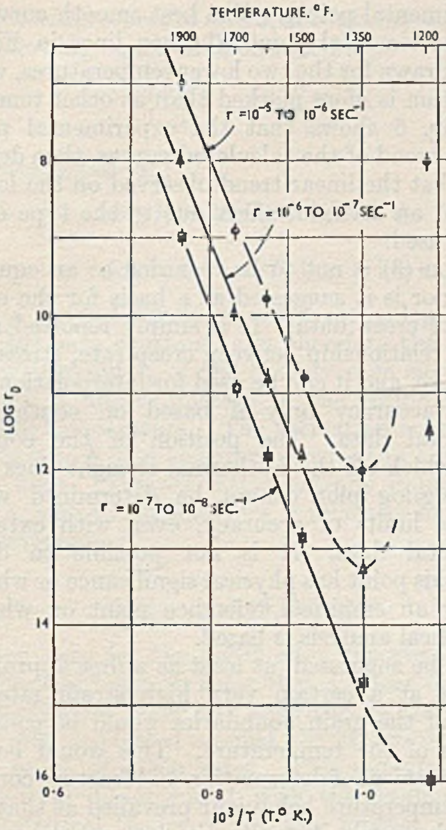

Fig. 2.—Plot of Log r_0 against $1/T$ for Three Ranges of Creep Rates (r).

S-590 alloy, the rate-process theory was applied to the data by deliberately limiting the range of creep rates, the graphical method suggested by Dushman *et al.*[4] being used. The intercepts $\log r_0$ at zero stress were determined for each curve after drawing the best average straight line in the limited range of creep rates. Three ranges, shown below, were selected at the lower creep rate end, where the approach to a straight-line semi-log relationship comes closest:

(*a*) 10^{-5}–10^{-6} sec.$^{-1}$
(*b*) 10^{-6}–10^{-7} sec.$^{-1}$
(*c*) 10^{-7}–10^{-8} sec.$^{-1}$

The values of $\log r_0$ are plotted against the reciprocal of the absolute temperature in Fig. 2.

The uncertainty of the points plotted in Fig. 2 is great, and therefore any conclusion or calculation of the activation energy or the activation entropy is open to doubt. Nevertheless, it appears that a straight line may be obtained when the values of $\log r_0$ are plotted against the reciprocal of the absolute temperature, except where the material approaches the transition from high-temperature to low-temperature behaviour. The trend of the points plotted in Fig. 2 indicates that an activation energy of the order of magnitude of 120,000 cal./g.-atom can be approximately estimated from these data; this value can vary between rather wide limits, according to the way in which r_0 is determined from the semi-log plot of stress against minimum creep rate.

In order to find a more general relationship between stress, minimum creep rate, and temperature, the data for S-590 alloy were replotted, using double logarithmic co-ordinates. The data included in the high-temperature range, in the absence of major sources of instability such as oxidation, show a definite linear trend when the logarithm of the minimum creep rate is plotted against the logarithm of the stress (Fig. 3). It was also observed that these straight lines drawn as shown converge at a single extrapolated point situated at a stress of $10^{10\cdot6}$ dynes/cm.^{2} and at a creep rate of 10^7 sec.$^{-1}$. When the slope of the isothermal lines is plotted against the reciprocal of the absolute temperature, a linear relationship is found (Fig. 4). Therefore, all the data included within the limits mentioned above may be expressed by a simple equation of the type:

$$\log r = a + (b/T - c)(\log s - d) \quad . \quad (3)$$

where:

r = minimum creep rate,
s = applied tensile stress,
T = absolute temperature,
a, b, c, d = constants depending on the material and structure.

For the S-590 alloy the constants are: $a = 7$, $b = 20{,}000$, $c = 8\cdot8$, $d = 1\cdot6$, if the minimum creep rate is expressed in sec.$^{-1}$ and the stress in 10^9 dynes/cm.^{2}.

Data calculated from equation (3) were recorded on a semi-log plot (solid lines in Fig. 5) together with

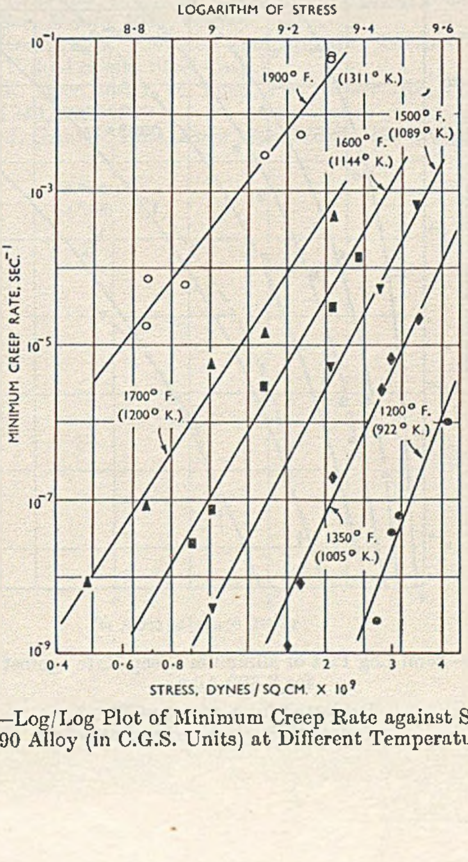

Fig. 3.—Log/Log Plot of Minimum Creep Rate against Stress for S-590 Alloy (in C.G.S. Units) at Different Temperatures.

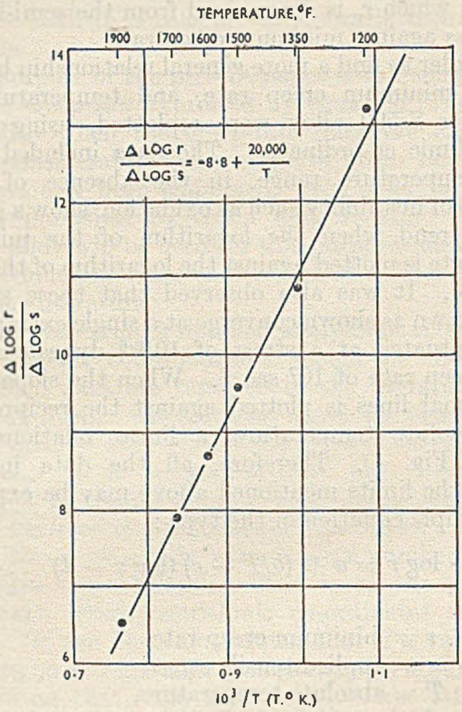

Fig. 4.—Plot of the Slopes of the Straight Lines from Fig. 3 against $1/T$.

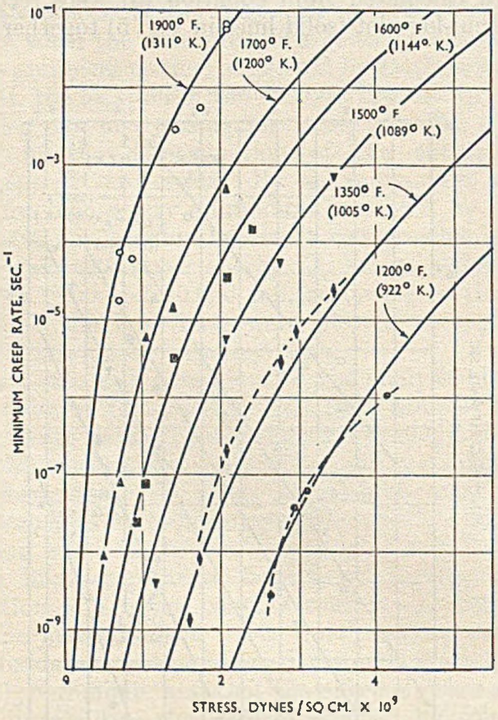

Fig. 5.—Semi-Log Plot of Minimum Creep Rate against Stress for S-590 Alloy.

—— Replotted from curves of Fig. 3.
- - - Curves through experimental points.

the experimental points. The best smooth curves to fit the experimental points (broken lines in Fig. 5) were also drawn for the two lower temperatures, where the deviation is more marked than at other temperatures. Fig. 5 shows that the experimental points follow the trend of the calculated curves, thus demonstrating that the linear trend observed on the log/log plot is not an artificial effect due to the type of co-ordinates used.

Equation (3) is not to be regarded as an equation of state, nor is it suggested as a basis for the extrapolation of creep data. It is simply reported as an empirical relationship between creep rate, stress, and temperature, and it can be used for interpolation with sufficient accuracy only if based on considerable experimental data. The position of the common point at which all the isothermal straight lines meet on the log/log plot cannot be determined within reasonable limits of accuracy, even with extensive experimental data. It is not possible to decide whether this point has physical significance or whether it is only an empirical reference point on which a mathematical analysis is based.

It may be suggested, at least as a first approximation, that at a certain very high strain rate, the strength of the grain boundaries would be constant, regardless of the temperature. This would be true if the mechanism of deformation and fracture common to high-temperature behaviour prevailed at that very high strain rate. Actually, it was proved experimentally that at a much lower strain rate a different mechanism of deformation and fracture becomes the controlling factor. This change is indicated on a log/log plot of minimum creep rate against stress by a change in slope of the experimental curves.

Although the physical significance of the reference point discussed above is doubtful, it is worth mentioning that a common point for all the curves plotted on double logarithmic co-ordinates was found to exist for other alloys, provided major instabilities were not operative. The data by White, Clark, and Wilson [9] for a Cr–Mo–Si steel, show this very clearly. The same evidence is found for the behaviour of 18 : 8 molybdenum stainless steel,[10] and for 3S-H12 aluminium alloy.[11] Even stress-rupture data have a similar trend, as can be shown using the data reported by Grant [12] for the cobalt–chromium " J " alloy.

It is obvious that at constant stress, equation (3) becomes of the same form as Boltzmann's equation :

$$\text{rate} = Ae^{-Q/RT}.$$

Moreover, it can be shown that the successful application of the rate-process theory to a limited range of creep rates, as demonstrated in Fig. 2, can be extended to any range of creep rates in a quite general way. In fact, equation (2), which is consistent with the rate-process theory, can be written :

$$\log r_i = \log r_0 + \frac{d \log r}{ds} \, . \, s \quad . \quad . \quad . \quad (4)$$

for each constant temperature. Differentiating equation (3) with respect to stress, at constant temperature, the following equation is obtained :

$$\frac{d \log r}{ds} = \frac{(b/T - c)}{s}$$

Substituting in (4) :

$$\log r_l = \log r_0 + b/T - c$$

or

$$\log r_0 = -b/T + c + \log r_l \; . \quad . \quad (5)$$

Over a limited range of creep rates, $\log r_l$ may be regarded as a constant, and therefore the logarithm of the intercept r_0 of the best average curve at zero stress is a linear function of the reciprocal of the absolute temperature. The activation energy is proportional to b, and therefore constant in any range of creep rates. The activation entropy is proportional to $(c + \log r_l)$, and therefore it increases as the creep rate increases.

III.—CONCLUSIONS

Creep data for the S-590 alloy reported by Grant and Bucklin [8] clearly show that a linear relationship is not obtained when the logarithm of the minimum creep rate is plotted against the stress over a fairly large range of creep rates. This experimental evidence is in disagreement with the fundamentals of the statistical theory of the rate process. The rate-process theory may be successfully applied only to a limited range of creep rates. If the experimental curves are compared with corresponding curves predicted by the rate-process theory, it may be concluded either that the theory cannot be applied or that the size of the unit of flow decreases (instead of remaining constant) as the strain rate increases at constant temperature.

Extrapolation of creep data based on the results of the rate-process theory cannot be safely carried out. Interpolation leads to good accuracy only if restricted to the same range of creep rates.

A general approximate equation is suggested for interpolation of creep data in a wider range of creep rates. This empirical equation is based upon the experimental evidence of a linear relationship between the logarithm of the minimum creep rate and the logarithm of the stress. It is in good agreement with the results of the rate-process theory only over a limited range of creep rates, and it is found to be valid only in the absence of major instabilities and therefore cannot be safely used for extrapolation of data unless the absence of instabilities can be assured.

ACKNOWLEDGEMENT

The authors wish to express their thanks to the U.S. Bureau of Ships, under whose sponsorship this research was carried out.

REFERENCES

1. H. Eyring, *J. Chem. Physics*, 1935, **3**, 107.
2. H. Eyring, *J. Chem. Physics*, 1936, **4**, 283.
3. W. Kauzmann, *Trans. Amer. Inst. Min. Met. Eng.*, 1941, **143**, 57.
4. S. Dushman, L. W. Dunbar, and H. Huthsteiner, *J. Appl. Physics*, 1944, **15**, 108.
5. A. S. Nowick and E. S. Machlin, *J. Appl. Physics*, 1947, **18**, 79.
6. E. S. Machlin and A. S. Nowick, *Trans. Amer. Inst. Min. Met. Eng.*, 1947, **172**, 386.
7. P. Ludwick, " Elemente der Technologischen Mechanik ". Berlin : **1908**.
8. N. J. Grant and A. G. Bucklin, *Trans. Amer. Soc. Metals*, 1950, **42**, 720.
9. A. E. White, C. L. Clark, and R. L. Wilson, *Trans. Amer. Soc. Metals*, 1938, **26**, 60.
10. " Steels for Elevated-Temperature Service ". **1949**. (United States Steel Corp.).
11. J. E. Dorn and T. E. Tietz, *Proc. Amer. Soc. Test. Mat.*, 1949, **49**, 815.
12. N. J. Grant, *Trans. Amer. Soc. Metals*, 1948, **40**, 585.

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A SIMPLE METHOD OF X-RAY MICROSCOPY AND ITS APPLICATION TO THE STUDY OF DEFORMED METALS *

1323

By R. W. K. HONEYCOMBE,† M.Sc., Ph.D., Member

SYNOPSIS

A method of obtaining images from metal crystals, using a line source of characteristic X-rays, is described. The images can be enlarged to at least 50 diameters to reveal significant microscopic phenomena, in particular distortions arising from slight plastic deformation which are not readily observed by optical micrography. However, optical micrography is of great assistance in the initial interpretation of the X-ray images, and the two techniques should be regarded as complementary. The scope of the method in the study of plastic deformation is illustrated by a series of X-ray and optical micrographs obtained from metal crystals deformed under varying conditions.

I.—INTRODUCTION

It has been realized for many years that X-rays can be used to obtain images of crystal surfaces, but few attempts have been made to apply such methods to the study of deformed crystals. Berg [1, 2] has described two ways by which X-ray images could be obtained. Firstly he suggested the use of a point source of polychromatic X-rays, a method which has recently been used by Guinier and Tennevin [3] to obtain sharp images of crystals. The second method, which Berg adopted for his experiments, employed characteristic radiation from a line source; he showed that if the crystal was set at a suitable Bragg angle, a point-for-point image of the crystal surface was obtained. In this way Berg examined crystals of natural rock salt and detected striations in the image which he attributed to plastic deformation by slip along (110) planes in a [100] direction.

The matter was not investigated further until 1945, when Barrett [4] showed that a similar technique could be applied to the study of many metallurgical problems. Barrett found that a surprising degree of resolution could be obtained by using an ordinary line source of the type frequently available in X-ray tubes. With high-resolution photographic plates, he was able to obtain significant detail in the X-ray images at magnifications of 50 or more. This work was necessarily exploratory in character, but it sufficed to show that X-ray microscopy of this type would be a valuable new way of studying the deformation of metals, as well as other phenomena such as recrystallization and grain growth.

In the present work, the author has used a method based on the same principle and has applied it primarily to a study of the deformation of metal crystals. In conjunction with the classical methods of optical micrography it has proved very useful in studying the inhomogeneities that develop during plastic deformation in both single-crystal and polycrystalline specimens. This work is the subject of a separate paper.[5] In the present paper it is proposed to describe and discuss the technique of X-ray microscopy and to show how some of the results can be interpreted.

II.—DESCRIPTION OF TECHNIQUE

The principle of the method is shown schematically in Fig. 1. The crystal is rotated until a Bragg reflection occurs, and then, if the crystal is perfect, the X-rays reflected are defined by a series of cones with apexes on the line source. The rays forming these cones will give on reflection a point-for-point image of part of the crystal surface on a suitably placed photographic plate. Clearly, the image will be unique because each point on the crystal surface will reflect only one incident ray, namely, that which makes the correct Bragg angle with the set of reflect-

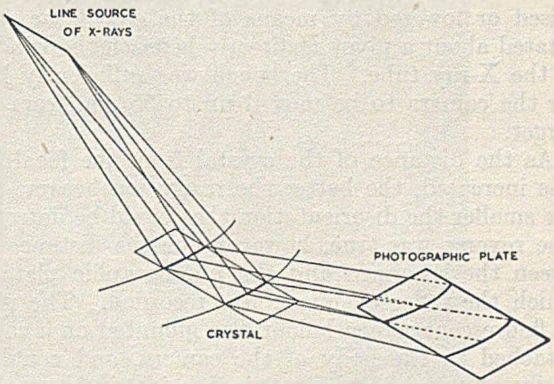

Fig. 1.—Schematic Diagram Showing Principle of X-Ray Technique.

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ing planes. As the crystal is rotated, other planes will, of course, reflect at different Bragg angles and result in similar images of the crystal. If the orientation of the crystal is not uniform, certain regions will not be in a position to reflect when the rest of the crystal is reflecting; this results in white areas in the X-ray image. The simplest case is that of a small included grain, which is revealed as a sharply bounded white area even if the orientation difference is quite small (see, for example, Fig. 5 (*a*), Plate VIII). The method has, however, been applied to the study of much smaller disorientations which occur as a result of plastic deformation. Here the interpretation is not quite so simple and will be dealt with in detail in a later section.

In practice, the apparatus is extremely simple. A demountable X-ray tube was employed for the present work, because the filament could be frequently replaced and the focal line on the target adjusted; normally the source was about 10 mm. long and 1 mm. wide. The target metal was chosen to give a radiation of relatively long wave-length, so that only a thin surface layer contributed to the reflection. Copper ($\lambda_{K\alpha} = 1{\cdot}59$ Å.) was frequently used, but equally satisfactory results were obtained with cobalt ($\lambda_{K\alpha} = 1{\cdot}74$ Å.) and manganese ($\lambda_{K\alpha} = 2{\cdot}10$ Å.) radiation. Fig. 4 (Plate VIII) is a photograph of the experimental arrangement. The window *W* of the X-ray tube was covered by a pair of adjustable slits, whose sole purpose was to limit the size of the beam. The crystal *C* to be examined was mounted on the spindle of a Unicam X-ray camera, and could be rotated about its vertical axis, or raised or lowered to bring different parts of the crystal into the beam. The camera was mounted on an adjustable bench which could be raised or lowered by means of the cantilever *L* or rotated about a pivot *P* directly beneath the window of the X-ray tube. The bench was sufficiently long for the camera to be moved up to 55 cm. from the target.

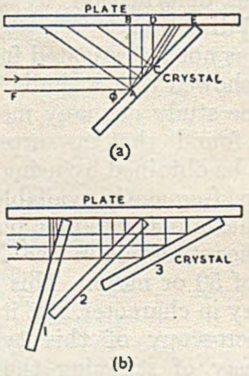
As the distance of the crystal from the focal line was increased, the better the resolution became and the smaller the disorientation that could be detected. The reverse was true, however, for the distance between the specimen and the photographic plate on which the reflected image was recorded. The plate or fluorescent screen holder was mounted on a rod *R* attached to the body of the camera, and could be easily adjusted to the required height. The holder could be moved horizontally so that the photographic plate was brought very close to the surface of the crystal, the usual working distance being between 2 and 5 mm. The holder was adjustable in three directions at right angles and therefore could make any desired angle with the crystal surface.

As the X-ray beam was very broad, it was necessary to reduce scattered radiation by means of a lead tube *T*, and to absorb the direct beam by a movable lead screen *S* which was guided along the camera table in brass slots.

To obtain an X-ray micrograph, the specimen, usually a large single crystal or coarse-grained polycrystal, was mounted in the X-ray camera, as shown in Fig. 4 (Plate VIII), and a fluorescent screen was placed in the holder about 2 cm. from the specimen and parallel to the incident beam. After the eyes had become accustomed to complete darkness, it was possible to see the individual reflections on the screen as the crystal was slowly rotated. A suitable reflection was selected and the plate-holder brought as close to the crystal as feasible, the plate being still parallel to the incident beam. In these circumstances the photographic plate was adequately protected from the direct beam by the thin metal flange of the holder. Even without going to extreme lengths to improve the sharpness of the line source, the resolution of the method proved so satisfactory that the images obtained revealed significant detail when enlarged 50–100 diameters. This necessitated the use of fine-grained photographic emulsions. Kodak Maximum Resolution plates capable of resolving 1000 lines/mm. proved to be extremely satisfactory; the exposure time was usually between 2 and 6 hr. For work requiring magnifications of only 20 times or less, Kodak B.10 plates were adequate, the exposure time varying between 1 and 30 min., depending on the intensity of the reflection. Kodak D.8 developer was used.

III.—SELECTION OF A SUITABLE REFLECTION

During rotation of the crystal, a large number of reflections were usually observed, but all were not suitable for detailed examination. The most obvious cause for rejection was the interference on the plate of two or more reflections. This frequently occurred, particularly with deformed crystals, as each reflection then took place over an extended angular range. The likelihood of interference was also greater because the reflections were frequently larger than 1 cm.² in area. Some reflections were excessively distorted as a result of striking the screen at a very oblique angle. Fig. 2 (*a*) shows the effect when the Bragg angle is altered by choosing a different set of reflecting planes (or by



(a)

(b)

Fig. 2.—Effect of (*a*) Variation of Bragg Angle and (*b*) Variation of Angle between Specimen and Incident Beam.

PLATE VIII.

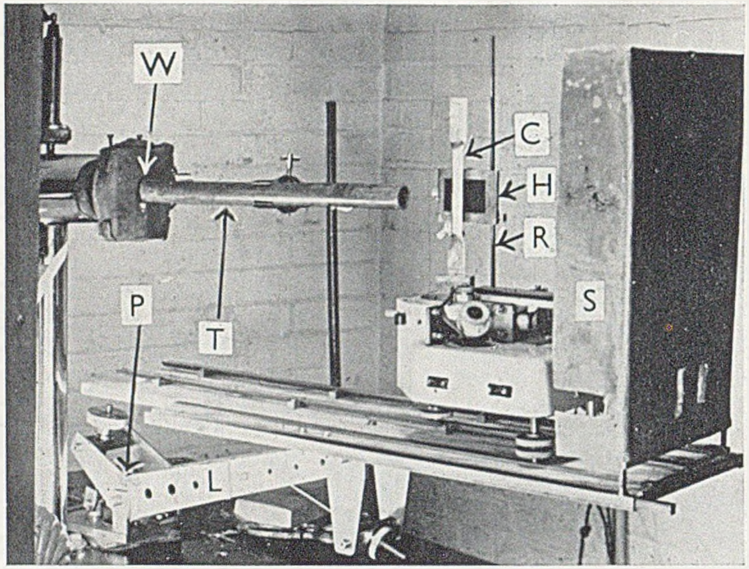
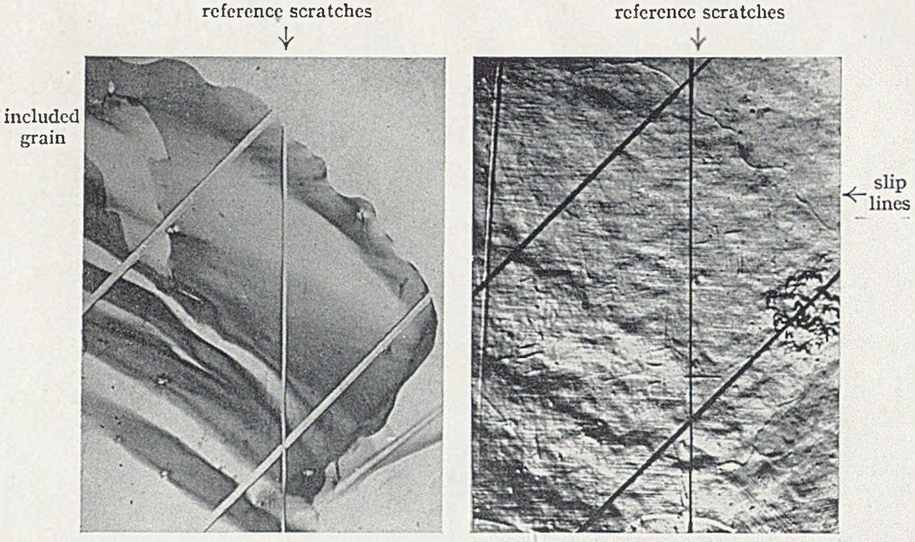
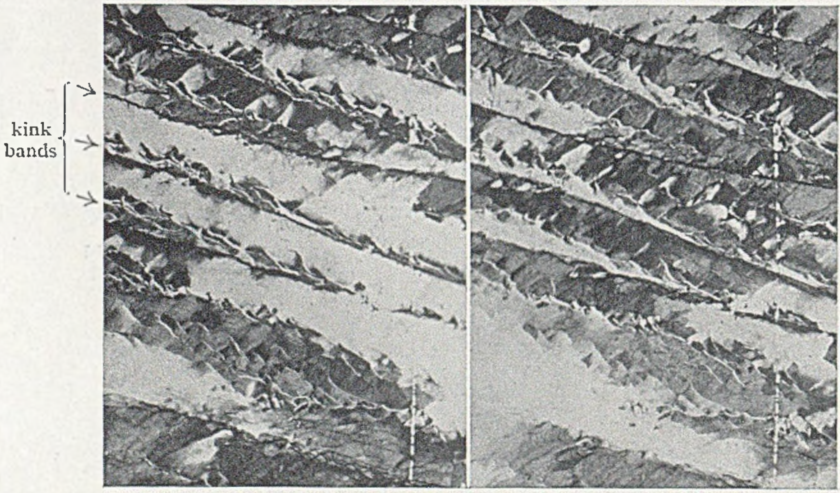


FIG. 4.—Experimental Arrangement for X-Ray Microscopy.



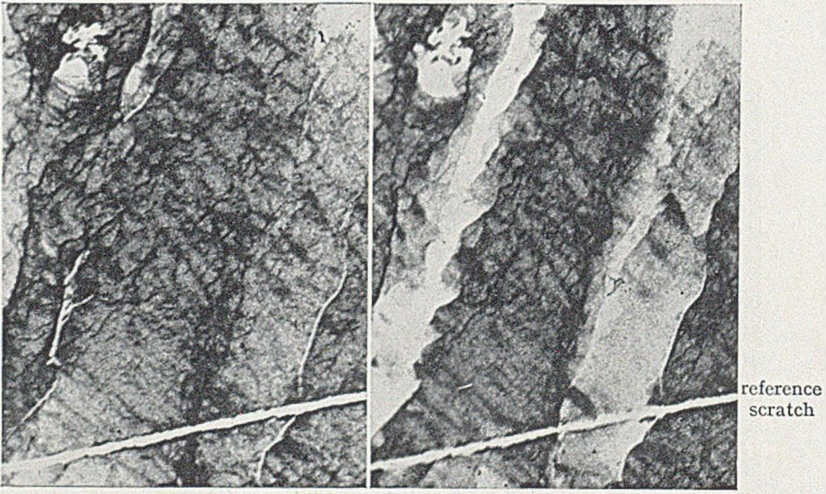
(*a*) X-Ray. (*b*) Optical.

FIG. 5.—A Grain in Al Polycrystal (99·99% pure) After 0·6% Elongation in Tension at Room Temperature. × 6.



(*a*) X-Ray. (*b*) X-Ray After Rotation of 15′.

FIG. 6.—A Grain in Al Polycrystal (99·99% pure) After 3% Elongation During 4 Days' Slow Deformation at 300° C. × 10.

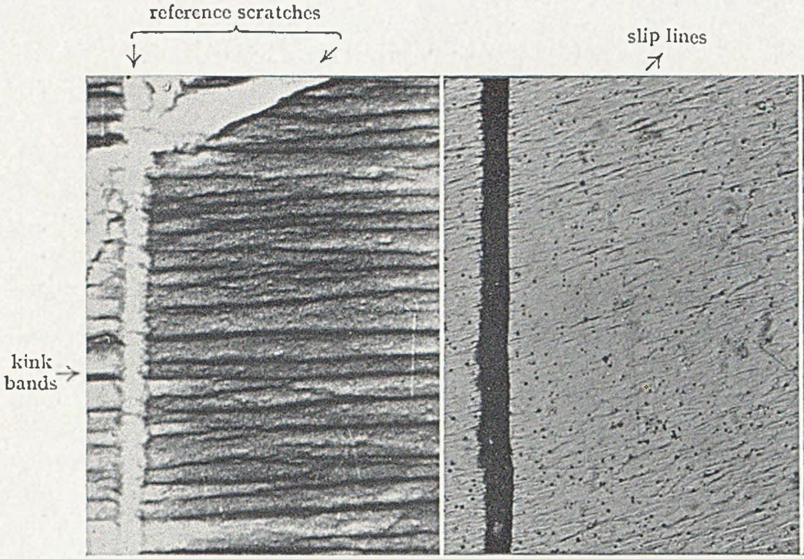


(*a*) Specimen–X-Ray Source = 20 cm. (*b*) Specimen–X-Ray Source = 52 cm.

FIG. 7.—X-Ray Micrographs of Al Crystal (99·99% pure) After 1·7% Elongation by Slow Straining at 300° C. × 20.

All except Fig. 4 reduced by ⅓ linear in reproduction.

[*To face p.* 40.



(*a*) X-Ray. × 25.

(*b*) Optical. × 50.

FIG. 8.—Al Crystal (99·5% pure) After 2·9% Rapid Extension at 450° C.

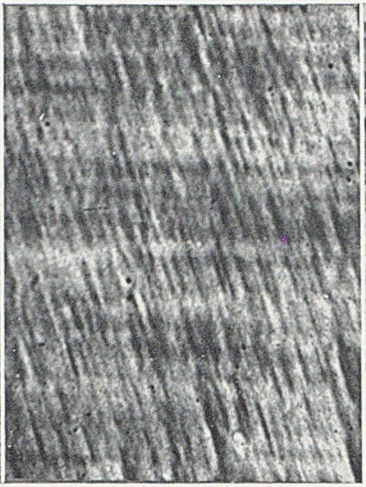


FIG. 9.—X-Ray. Al Crystal (99·5% pure) After 4% Elongation at Room Temperature. × 50.

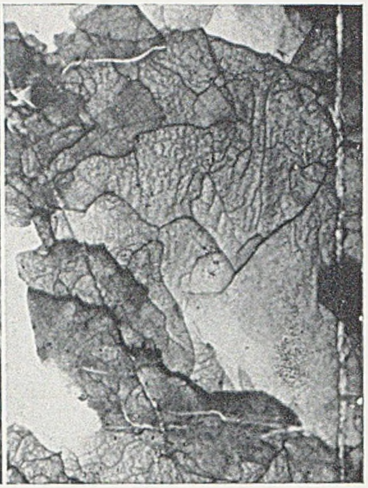
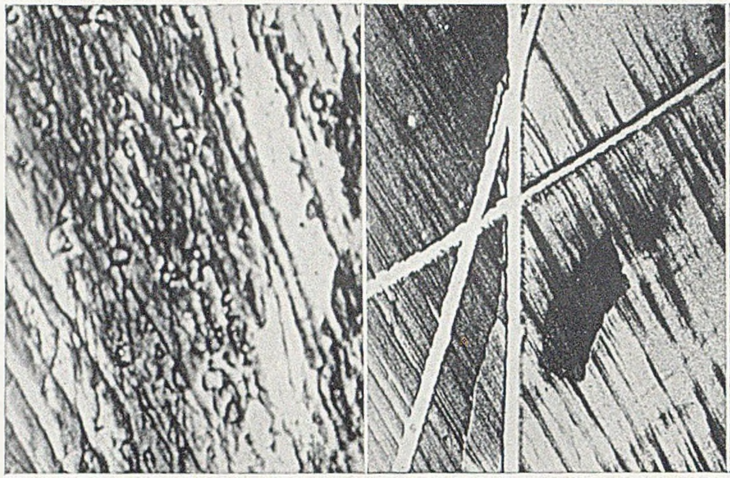


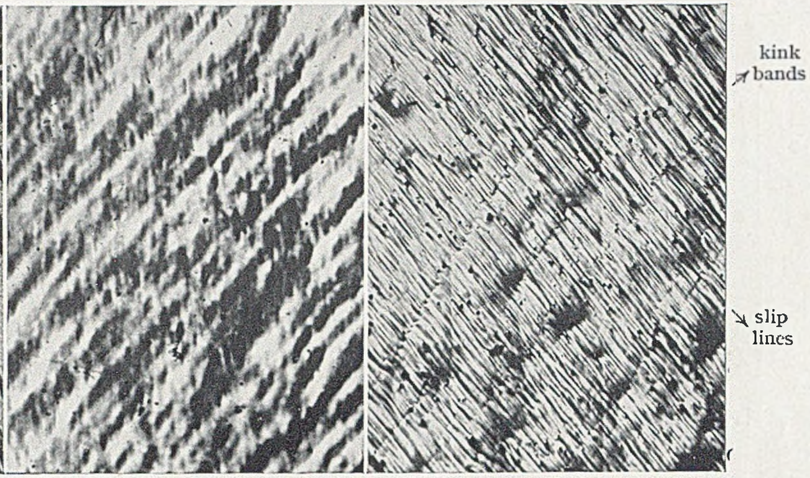
FIG. 10.—X-Ray. Al Crystal (99·99% pure) After 7% Elongation During 13 Days' Slow Extension at 300° C. × 20.



(*a*) X-Ray. After 2·9% elongation.

(*b*) X-Ray. As (*a*) after prolonged electrolytic polishing.

FIG. 11.—Cu Crystal, Originally Ground and then Electrolytically Polished. × 25.



(*a*) X-Ray. After 4·6% elongation. × 25.

(*b*) Optical. After 10% elongation. × 50.

FIG. 12.—Al Crystal (99·5% pure) After Extension at Room Temperature.

All figures reduced by ⅓ linear in reproduction.

PLATE IX.

using a different wave-length), while the angle between the specimen and the photographic plate remains constant (in this case 45°). It is clear that the image is the same size as the area irradiated when the triangle *ABE* is isosceles, in which case the Bragg angle is approximately 56°. As the Bragg angle increases beyond 56°, the image becomes larger, while if the angle is below 56° the image is correspondingly smaller. In practice, the best results are obtained when the reflection is not distorted greatly one way or the other.

Other factors remaining constant, the reflection comes from a larger area of the specimen if it is almost parallel to the incident beam than if it is almost normal to it. This is shown in Fig. 2 (*b*), where a Bragg angle of 45° is used for simplicity. When the photographic plate and the crystal are at a very acute angle, images are frequently obtained from the complete width of the surface presented to the incident beam. There is another reason for putting the plate at a very acute angle to the specimen. In practice, it is found that the resolution of the image is greater, the closer the plate is to the specimen, but some parts of the specimen are inevitably more distant from the photographic plate than others (Fig. 2). This difference is obviously minimized by making the angle between the plate and crystal as small as possible.

IV.—THE APPEARANCE AND INTERPRETATION OF X-RAY IMAGES FROM DEFORMED CRYSTALS

The image from an undeformed crystal is uniform, and no detail can be seen after considerable enlargement. Marked changes in orientation such as those occurring at grain boundaries and twin planes are clearly revealed by sharp breaks in the continuity of the reflection. However, when a metal crystal is deformed, local disorientations usually occur, the presence of which is shown indirectly by asterisms in X-ray Laue photographs. While the detailed nature of these inhomogeneities will be considered elsewhere, it is relevant now to consider how the X-ray images can be modified by them.

The most straightforward change in the image arises when the local rotation of the crystal is so great that this region is not able to reflect any of the incident X-rays. Thus the distorted region will behave exactly like an included grain of different orientation and a white area will appear in the X-ray image. Fig. 5 (*a*) (Plate VIII) illustrates this modification of the image in the case of a reflection from one grain of a coarsely crystalline aluminium specimen, which was deformed by 0·6% elongation in tension at room temperature. The white bands represent lamellar disorientations.

There are also white areas in Fig. 6 (Plate VIII) which correspond with regions of the crystal that, as a result of plastic deformation, have become disoriented so that they no longer contribute to the reflection. In these circumstances, it is reasonable to expect that a slight rotation of the specimen may bring some of the non-reflecting regions into a reflecting position, while at the same time other regions may cease to reflect. In practice, a lightly deformed aluminium crystal will give a reflection over an angular range of 1°–2°, the angular spread increasing with the deformation; but at each particular setting of the crystal within this range a slightly different image will be obtained because of local variations in orientation. For example, Figs. 6 (*a*) and (*b*) (Plate VIII) represent the same reflection from a deformed aluminium crystal which has been rotated 15′ between the two photographs. The same general features are evident in both photographs, namely the lamellar disorientations, but some areas of the crystal which have not reflected in Fig. 6 (*a*) are reflecting in Fig. 6 (*b*) and vice versa.

If a disorientation in the crystal is sufficiently small, then the region concerned may be in a position to select rays which make the requisite Bragg angle from another point on the line source. Fig. 3 shows diagrammatically that this will result in a local displacement of the image which will cause interference on the photographic plate. It will be more pronounced the greater the distance of the plate from the specimen, and this is an additional reason for placing the photographic plate as near as possible to the crystal. If, on the other hand, the distance between the X-ray source and the crystal is increased, then the image will become more sensitive to orientation differences in the crystal. This follows directly from the fact that the angle subtended by the source to any point on the crystal will be smaller the greater the distance between them. The effect of changing this distance on the sensitivity of the technique is illustrated in Figs. 7 (*a*) and (*b*) (Plate VIII), which represent an X-ray reflection from an aluminium crystal after 1·7% elongation by slow straining at 300° C. Fig. 7 (*a*) was taken with the crystal 20 cm. from the X-ray source, and shows no large white regions representing disorientations. However, in Fig. 7 (*b*), which was taken with the crystal 52 cm. from the source, large disoriented regions are revealed.

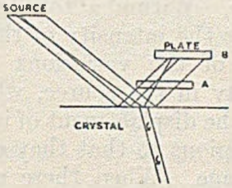


Fig. 3.—Effect of Local Rotation on the Image.

There is obviously a practical limit to disorientations that can be detected with the technique as it stands; this is imposed by the dimensions of the X-ray source and by the workable distance of the specimen from the source. If the line source is assumed to be 10 mm. long and it is foreshortened to 1 mm. by using the X-ray tube in the usual way, then for a crystal 50 cm. away orientation differences of 6′ should be detected. In practice, the disorientations which can be observed are rather smaller than this, but no quantitative estimate has been made. In this respect the method is not as sensitive as one devised by Guinier and Tennevin[3] for investigating the perfection of crystals. Their method employs a point source of polychromatic X-rays and by using the focusing condition it has been possible to detect disorientations as small as 10″. However, this method does not reveal the *distribution* of inhomogeneities in the crystal, and thus does not fall into the category of X-ray microscopy. It should be mentioned that the technique described here is more than adequate for the study of lightly deformed crystals, because after about 1% elongation in tension the resulting disorientations are usually sufficiently large for detection by X-ray microscopy. In fact, after about 10% elongation in tension, the disorientations frequently become so great that the method is then too sensitive. This difficulty can be partly overcome by reducing the distance of the specimen from the X-ray source to a minimum.

It soon became evident that the above considerations were not the only ones defining the nature of the X-ray image. Within a totally reflecting area, it was found that the intensity of the image was not uniform. Some of the variations resulted from a lack of uniformity in the source, while others could be attributed to the displacement of incident radiation by disoriented regions so that they overlapped other parts of the image. When these causes had been eliminated, however, there remained narrow lines of high intensity in many of the X-ray images. By comparison with optical micrographs it was found that distorted regions of the crystal were responsible for the areas of high intensity. For example, one type of inhomogeneity resulting from deformation consists of narrow bands in which the crystal is sharply bent—*kink bands*. The lamellæ *between* these narrow curved regions become disoriented as the deformation proceeds, and a lamellar type of disorientation results (see, for example, Fig. 6, Plate VIII). However, the bands themselves frequently show in the X-ray images as dark lines more intense than the adjacent regions of the crystal. This effect is best observed when the disorientations have been insufficient to produce the white non-reflecting lamellæ in the images. Fig. 8 (*a*) (Plate IX) is an X-ray image from an aluminium crystal deformed by 2·9% elongation at 450° C. The dark bands correspond with narrow kink bands on (110) planes. In other circumstances, kink bands can also be revealed by the absence of a reflection when the rotations within them have been relatively great. These two effects of kink bands on the X-ray image are illustrated in Fig. 7 (*a*) (Plate VIII), in which both black and white lines can be seen. It should be mentioned that if a crystal containing kink bands is annealed at a high temperature, so that the curvatures are eliminated by polygonization, the intensity of the reflections from the bands diminishes markedly and more bands appear as white lines in the image.

The reason for these intense reflections may be the focusing by the bent lattice, but it seems more likely that it is an extinction phenomenon resulting from the distortions, which are in effect concentrations of dislocations. Some support for the latter view is found in the fact that other phenomena in deformed crystals which are associated with concentration of dislocations produce similar effects in the X-ray images. For example there is the slip process, which, although it occurs on a scale rather too fine for complete investigation by the X-ray technique at its present stage of development, nevertheless can be detected and studied under favourable conditions. Here also, the slip bands are revealed in the X-ray images as narrow bands of high intensity (Fig. 9, Plate IX).

Polygonization,[6] although it removes lattice curvatures by diffusion of dislocations and thus causes changes in the intensity of reflections from kink bands, does form narrow walls of dislocations separating regions differing slightly in orientation. These walls are often revealed in X-ray images as dark lines. Fig. 10 (Plate IX) is such an X-ray image taken from an aluminium crystal after 7% elongation in 13 days at 300° C., in which a complicated network of polygon boundaries is visible. In this case the disorientations between the polygons are so small that the majority are reflecting the incident X-rays, and they would not have been detected but for the stronger reflections from their boundaries.

Once it is appreciated that the X-ray image may be modified in the ways mentioned above, it is possible with some experience to interpret the images fairly accurately. Interpretation is greatly aided by using the X-ray method side by side with optical micrography, although frequently the features which are observed in X-ray micrographs after small deformations, are not detected optically till after much heavier deformation. In the next section, a more detailed comparison of the results obtained by the two methods will be given.

V.—A COMPARISON OF X-RAY AND OPTICAL MICROGRAPHY OF DEFORMED CRYSTALS

It should be emphasized first that the features of deformed crystals revealed by X-ray micrography differ from those normally observed by optical micrography. The optical examination of metals depends to a large extent on the observation of differences in surface level, e.g. grain boundaries and slip bands, and of differences in appearance produced by etching. X-ray micrography on the other hand is not particularly sensitive to small changes in surface level; this can be seen by comparison of Figs. 5 (*a*) and (*b*) (Plate VIII), which represent X-ray and optical macrographs taken from one grain of a coarsely crystalline aluminium specimen after 0·6% elongation in tension at room temperature. Optical examination in oblique illumination has revealed slip bands and a number of irregular surface blemishes (Fig. 5 (*b*)), all of which are absent in the X-ray macrograph (Fig. 5 (*a*)). The great advantage of X-ray micrography in the study of deformed crystals is in the detection of small local changes in orientation not revealed by optical microscopy. The X-ray reflection gives a map of the orientation changes which produce optically observable effects only at much heavier deformation. For example, Fig. 5 (*a*) contains several white bands representing disorientations in the crystal; on the other hand optically it is only just possible to detect some of the bands in oblique illumination (Fig. 5 (*b*)) because of slight changes in surface level. Furthermore, there is no indication in Fig. 5 (*b*) of the smaller included grain which is clearly delineated as a white area in Fig. 5 (*a*).

Scratches inscribed on specimens show up clearly in X-ray images (Figs. 5 (*a*), and 7, Plate VIII, and Fig. 8 (*a*), Plate IX) because the severe deformation in the scratch causes local disorientations. For this reason it is essential to eliminate mechanical polishing as a means of preparing the crystal surfaces. This need is further illustrated in Figs. 11 (*a*) and 11 (*b*) (Plate IX). A copper crystal was filed to produce a flat surface, ground, then electrolytically polished to remove all scratches. It was then deformed by 2·9% elongation, but the results of this deliberate deformation were entirely masked in the X-ray image by the effects of filing and grinding (Fig. 11 (*a*)). The crystal was then further electropolished to remove 2 mm. of metal, approximately 25% of its original thickness. The X-ray micrograph taken at this stage revealed the more uniform effects of the tensile deformation and the coarse growth mosaic typical of cast single crystals (Fig. 11 (*b*)). X-ray micrography is thus a simple and sensitive way of studying the extent of deformation of metal crystals by various surface treatments.

The X-ray technique has proved very useful in a study of the plastic deformation of single crystals, the detailed results of which are to be discussed in a later paper.[5] While the method is applicable to an individual grain in polycrystalline aggregates, the use of single crystals is preferable so that simpler stresses can be applied to the specimen during deformation. X-ray micrographic examination of single crystals deformed in tension has revealed the presence of local disorientations or inhomogeneities long before they could be detected optically. For example, Fig. 12 (*a*) (Plate IX) shows an X-ray micrograph of an aluminium single crystal after 4·6% elongation in tension at room temperature, in which lamellar disorientations are evident. The cause of these disorientations proved to be narrow kink bands or regions of curvature which could not be detected optically until after much greater deformation. Fig. 12 (*b*) (Plate IX) is an optical micrograph taken on the same crystal after 10% elongation, in which the narrow kink bands can just be seen; however, there is no indication of the marked disorientations which have resulted in the crystal. In Fig. 8 (*a*) (Plate IX), which is an X-ray micrograph from an aluminium crystal deformed by 2·9% elongation in tension at 450° C., the disorientations have not been sufficient to show up as white regions in the image, but the distorted regions appear as intense black bands. The corresponding optical micrograph reveals no sign of the kink bands, the only feature being the rather ill-defined high-temperature slip bands (Fig. 8 (*b*)).

It is not obvious at first sight that the bands of Figs. 8 (*a*) and 12 (*a*) (Plate IX), are identical, but a careful comparison of X-ray and optical micrographs at various stages of the deformation confirms this view. The most convincing evidence is obtained from the orientation of the bands within the crystals. The determination of the orientation of the bands is a simple matter in optical micrography if the orientation of the crystal itself is known; but with X-ray micrographs some distortion of angular relationships existing in the crystal usually results in the X-ray image. This follows because the reflected image may strike the photographic plate at any angle in space. A correction factor can be obtained by inscribing on the specimen two fine scratches making different angles with the axis. These will appear in the X-ray image (e.g. Fig. 11 (*b*), Plate IX), and the angles can again be measured.

Now it can be shown that:

$$\tan \alpha = A \tan \beta + B$$

where α = angle on X-ray image, β = actual angle on specimen surface, and A and B are constants for a given reflection. With two pairs of angles, two simple equations are obtained from which the factors A and B can be determined.* Thus the angles made by bands on surfaces of the crystal can be found when the bands are visible only in X-ray images. The orientation of the bands is then obtained by plotting the angles obtained from two faces of the crystal on a stereographic projection containing the crystal orientation. By this method it was found that the bands in Figs. 8 (*a*) and 12 (*a*) (Plate IX) occurred on (110) planes in the early stages of the deformation. In all the crystals examined these bands possessed the same crystallographic habit independent of temperature and speed of deformation.

X-ray micrography has revealed a second lamellar type of disorientation, but in most of the crystals examined the bands were initially almost parallel to the active slip plane, namely (111) in aluminium and copper. Fig. 11 (*b*) (Plate IX) is an X-ray image of a copper crystal deformed to 2·9% elongation by tension in which these lamellar disorientations are visible. Because of their crystallography they could easily be confused with slip bands, but optical examination at much greater deformations confirmed that such bands were present in the crystal and showed that they were regions in which a second slip system had operated, causing local disorientations.

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* B is usually quite small and can often be neglected.

Thus optical examination revealed how the bands were formed, while X-ray microscopy proved to be a much more sensitive means of detecting them.

The slow deformation of crystals at elevated temperatures usually results in the formation of similar inhomogeneities. For example, Figs. 6 (*a*) and (*b*) (Plate VIII) are X-ray micrographs taken from an aluminium crystal after four days' slow deformation at 300° C. (3% elongation). The large-scale lamellar form of the image is due to disorientations resulting from the formation of kink bands, which under these conditions cause almost macroscopic disorientations in aluminium of high purity. However, slow deformation at high temperatures is a rather different problem from that of deformation at low temperatures, because polygonization can occur simultaneously with the deformation. The work of Wood and his collaborators [7, 8] has shown the important part played by this phenomenon in the creep of aluminium polycrystals at elevated temperatures. By X-ray powder methods and optical micrography, they have found that strain-free fragments are formed within the individual crystals at an early stage of the deformation. But these methods do not reveal the origin of the fragments, nor do they show their exact mode of distribution within the grain and how it changes as the deformation proceeds further. X-ray microscopy is proving of value in this field, as it is capable of showing the mode of occurrence of the polygons at an early stage of the deformation, and the behaviour of individual fragments of a crystal can be followed. It has been mentioned already that the polygons are first revealed in X-ray micrographs by the intense reflections from their boundaries (Fig. 10, Plate IX); in this way the fragments can be detected in the early stages of the deformation when the orientation differences are too small to be revealed by absence of reflections. As the deformation proceeds the disorientations between the polygons become greater and they no longer all reflect simultaneously. This stage is shown in Figs. 6 (*a*) and (*b*) (Plate VIII) in which the long lamellæ are no longer reflecting uniformly—there are white areas within dark lamellæ which represent large individual polygons.

In conclusion, it is not suggested that the X-ray method described here is incapable of improvement. Certain alterations such as the provision of a much smaller source and a reduction of the divergence of the X-ray beam will certainly increase the sensitivity of the technique. However, the aim of this paper has been to show that the method as it stands provides a simple and effective way of studying the deformation of metal crystals, which yields information not normally obtained by optical micrography or more standard X-ray methods. The interpretation of the X-ray images can be a source of difficulty, particularly if the deformation is severe, but if the experiments are restricted to lightly deformed specimens and include optical micrography, then significant results can be obtained.

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REFERENCES

1. W. Berg, *Naturwiss.*, 1931, **19**, 391.
2. W. Berg, *Z. Krist.*, 1934, **89**, 286.
3. A. Guinier and J. Tennevin, *Compt. rend.*, 1948, **226**, 1530.
4. C. S. Barrett, *Trans. Amer. Inst. Min. Met. Eng.*, 1945, **161**, 15.
5. R. W. K. Honeycombe, *J. Inst. Metals*, 1951–52, **80** (in the press).
6. R. W. Cahn, *J. Inst. Metals*, 1949–50, **76**, 121.
7. G. R. Wilms and W. A. Wood, *J. Inst. Metals*, 1948–49, **75**, 693.
8. W. A. Wood and W. A. Rachinger, *J. Inst. Metals*, 1949–50, **76**, 237.

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SYMBOLS AND ABBREVIATIONS FOR USE IN "METALLURGICAL ABSTRACTS"

Å.	Ångström units = 1×10^{-8} cm.
abs.	absolute
A.C.	alternating current
addn.	addition(s)
amp.	ampere(s)
amp.hr.	ampere-hour(s)
appn.	application(s)
approx.	approximate(ly)
aq.	aqueous
at.-%	atomic per cent.
at. wt.	atomic weight(s)
atm.	atmosphere(s) [pressure]
A.W.G.	American wire-gauge
B. & S.	Brown & Sharpe (gauge)
b.c.c.	body-centred cubic
°Bé.	degree Baumé [scale]
B.H.N.	Brinell hardness number(s)
b.h.p.	brake horse-power
b.p.	boiling point(s)
B.Th.U.	British thermal unit(s)
B.W.G.	Birmingham wire-gauge
°C.	degree Celsius (formerly centigrade) [scale]
C	coulomb(s)
cal.	calorie(s)
c.c.	cubic centimetre (s)
c.d.	current density(ies)
C.G.S.	centimetre-gramme-second (units)
chem.	chemical
cm.	centimetre(s)
coeff.	coefficient(s)
compn.	composition(s)
conc.	concentrated (adj.)
const.	constant (adj. or noun)
contg.	containing
cryst.	crystalline
crystn.	crystallization
cwt.	hundredweight(s)
D.C.	direct current
d	density(ies)
detn.	determination(s)
dia.	diameter(s)
dil.	dilute (adj.)
dm.	decimetre(s)
D.P.N.	diamond pyramid (hardness) number(s)
elect.	electric, electrical(ly)
electrochem.	electrochemical(ly)
e.m.f.	electromotive force(s)
estn.	estimation(s)
eV.	electron volt(s)
°F.	degree Fahrenheit [scale]
F	Faraday's constant
f.c.c.	face-centred cubic
f.p.	freezing point(s)
ft.	foot; feet
ft.-lb.	foot-pound(s)
g.	gramme(s)
g.-atom	gramme-atom(s)
g.-mol.	gramme-molecule(s)
gal.	gallon(s)
grav.	gravimetric(ally)
H.F.	high-frequency
h.c.p.	hexagonal close-packed
H-ion	hydrogen-ion
h.p.	horse-power
h.p.-hr.	horse-power-hour(s)
hr.	hour(s)
in.	inch(es)
in.-lb.	inch-pound(s)
indust.	industrial
°K.	degree Kelvin absolute [temperature scale]
kc.	kilocycle(s)
kc./s.	kilocycle(s) per second
kg.	kilogramme(s)
kg.cal.	kilogramme-calorie(s)
kg.m.	kilogramme-metre(s)
km.	kilometre(s)
kV.	kilovolt(s)
kVA.	kilovolt-ampere(s)
kW.	kilowatt(s)
kWh.	kilowatt-hour(s)
kX	Crystal Ångström(s) = 1000 Siegbahn X-units
l.	litre(s)
lb.	pound(s)
L.F.	low-frequency
liq.	liquid (adj.)

M	molar [solution]
m.	metre(s)
m.amp.	milliampere(s)
math.	mathematical(ly)
max.	maximum (adj. or noun)
Mc.	megacycle(s)
Mc./s.	megacycle(s) per second
mech.	mechanical
met.	metallurgical(ly)
mg.	milligramme(s)
ml.	millilitre(s)
min.	minute(s); minimum (adj. or noun).
mm.	millimetre(s)
m.m.f.	magnetomotive force(s)
mol.	molecule(s)
m.p.	melting point(s)
mμ	millimicron(s) = 1×10^{-7} cm. = 10 Å.
mV.	millivolt(s)
N	normal [solution]
N.P.L.	National Physical Laboratory (Teddington, Middlesex)
No.	number(s)
N.T.P.	normal temperature and pressure
Oe.	oersted(s)
opt.	optical(ly)
oz.	ounce(s)
P.C.E.	pyrometric cone equivalent
p.d.	potential difference [electric]
*p*H	hydrogen-ion concentration
phys.	physical(ly)
p.p.m.	part(s) per million
prepn.	preparation
prodn.	production
pptn.	precipitation
P.S.	proof stress
qual.	qualitative(ly)
quant.	quantitative(ly)
°R.	degree Réaumur [scale]
recrystn.	recrystallization
ref.	reference(s)
resp.	respective(ly)
r.p.m.	revolution(s) per minute
sci.	scientific
sec.	second(s)
sepn.	separation(s)
soln.	solution(s)
sp.	specific
sp. gr.	specific gravity(ies)
spectrochem.	spectrochemical(ly)
sq.	square
S.W.G.	standard wire-gauge (Imperial)
tech.	technical(ly)
temp.	temperature(s)
t.p.i.	threads per inch
U.T.S.	ultimate tensile stress(es)
V.	volt(s)
VA.	volt-ampere(s)
vol.	volume(s)
W.	watt(s)
Wh.	watt hour(s)
wt.-%	weight per cent.
yd.	yard(s)
Y.P.	yield point(s)
%	per cent.
γ	microgramme = 1×10^{-6} g.
λ	wave-length
μ	micron(s) = 1×10^{-4} cm.
μμ	1 millionth micron = 1×10^{-10} cm. = 0·01 Å.
Ω	ohm(s)
°	degree(s) arc
′	minute of the arc
″	second of the arc
/	per
<	less than
>	greater than
≮	not less than
≯	not greater than
⩽	equal to or less than
⩾	equal to or greater than
≠	not equal to
≡	identically equal to
≅	approximately (or essentially) equal to
~	about
∝	varies as
∥	parallel
⊥	perpendicular

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 19 SEPTEMBER 1951 PART 1

1—PROPERTIES OF METALS

***The Mechanical and Physical Properties of [Commercially] Pure Aluminium and Some Aluminium[-Magnesium] Alloys at the Temperature of Liquid Oxygen.** H. Mäder (*Metall*, 1951, 5, (1), 1–5).—The mech. properties of pure Al (99·5%), its binary alloys with 3, 5, and 7% Mg, and a ternary alloy with Mg and Mn, have been determined on autogenously welded and on unwelded sheets (4 mm. thick) of annealed material at room temp. (+ 20° C.) and at the temp. of liq. O_2 (−183° C.). The results, which are given in tables and graphs, show that: (1) the static tensile properties of both welded and unwelded materials are higher at −183° C. than at room temp., the difference between the two sets of values being greatest with Al and becoming less marked with increase in Mg content; the ternary alloy has values approx. equal to those of the binary alloy contg. 3% Mg; (2) the bendability of welded material is higher at low temp. (− 150° C.) than at room temp.; it decreases with increase in thickness of the sheet; the ternary alloy has values approx. equal to those of the binary alloy contg. 5% Mg; (3) impact strength is lower at − 183° C. than at room temp.; it decreases with increasing Mg content and, in all cases, is considerably lower—by ~75%—after welding; and (4) dynamic tensile strengths at −183° C. are little different to those at +20° C., although decrease in temp. does slightly improve the strength after welding—which is only ~20–25% of that before welding; the ternary alloy is inferior to all other materials. The coeff. of linear expansion, α: (i) increases with increase in Mg content, and (ii) decreases with temp., and, for all materials $\alpha \simeq 17 \times 10^{-6}$/°C. at −120° C. The mean sp. heat, c_p, is practically independent of compn., and falls from 0·224 cal./g. between +20° and +100° C. to 0·180 cal./g. between +20° and −183° C. It is concluded that the binary alloys with 3 or 5% Mg are eminently suitable—with a slight preference for the latter compn.—for the construction of welded containers for the transport and storage of liquefied gases.—E. N.

***Application of Microradioanalytic Method to the Study of the Diffusion of Copper in Aluminium in the Presence of Calcium.** Mladen Paić (*Rev. Mét.*, 1951, 48, (2), 116–120).—The resistance of the Al–Cu–Mg alloys to corrosion can be increased by cladding them with pure Al, and this increased resistance is further improved when the coating contains small amounts of Ca, as a result, it is said, of the inhibiting action of the Ca on the rate of diffusion of the Cu. In order to investigate this observation, which was made by Bungardt (*Z. Metallkunde*, 1940, 32, 363; *Met. Abs.*, 1941, 8, 306) from microscopical examination, P. submitted specimens with cladding contg. different amounts of Ca to microradioanalysis (*Compt. rend.*, 1941, 213, 572, *et al.*; *Met. Abs.*, 1942, 9, 84), and so followed the Cu content of the coating as a function of the distance from the surface and the time of heating. A comparison of the gradients of the Cu content in coatings heated to 500° C. for various times showed that the rates of diffusion of the Cu increased with the Ca content of the coating. Hence, the fact that the presence of Ca in the coating increased its resistance to corrosion could not be due to the non-existent inhibiting effect of Ca on the rate of diffusion of Cu. The Cu reached the outer surface of an Al cladding 0·15 mm. thick after heating for ~80 min. at 500° C. 7 ref.—J. H. W.

***The Pentavalent State of Americium.** L. B. Werner and I. Perlman (*J. Amer. Chem. Soc.*, 1951, 73, (1), 495–496).—A note. The existence of Am^{V} is corroborated. Its prepn. and properties are described.—J. R.

***Transitions in Chromium.** M. E. Fine, E. S. Greiner, and W. C. Ellis (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 56–58).—A report of a study of the thermal dependence—between −200° and +200° C.—of a number of the properties of Cr. The samples were prepared: (1) by powder-metallurgy methods, and finally annealed at 1400° C. in purified He, the final product having a purity of 99·8%, and (2) by electroforming from aq. soln., after which they were packed in Cr, and vacuum annealed at 1000° C. Discontinuous changes were found in the Young's modulus, internal friction, coeff. of expansion, elect. resistance, thermal e.m.f., and thermoelectric power, which are evidence of a transition near room temp. It takes place over a large temp. range, progresses only as the temp. is changing, within the experiment range is independent of the rate of heating and cooling, and is reversible with no detectable hysteresis; on heating, the transition begins at ~ −20° C., increases in intensity to a critical value near 37° C., but is not complete until ~140° C. The difference between the observed lattice const. and those calculated from the expansion data suggests that the transition involves a departure from the ideal b.c.c. arrangement; the difference corresponds to 0·04% more atoms (on the average) associated with a unit cell above the transition than below and an equal decrease in the proportion of unit cells. The extra atoms by short-range movements perhaps assume interstitial (face-centred) positions giving the denser packing; or, alternatively, denser packing could arise from the disappearance of vacant lattice sites on heating. Young's-modulus data disclosed evidence for another transition near −150° C. This modulus effect, taking place between −160° and −120° C., with a critical temp. at −152° C., is similar to the one near room temp., but is of much smaller magnitude. 14 ref.—E. N.

***Cobalt Self-Diffusion: A Study by the Method of Decrease in Surface Activity.** R. C. Ruder and C. E. Birchenall (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (2), 142–146).—The diffusion coeff. of Co into Co (99·9 + %) and Ni (99·9 + %) have been measured, in an atmosphere of H, by the decrease in surface activity of thin (~10^{-7} cm.) films of Co contg. the radioactive isotope Co^{60} plated on to the diffusion surfaces. The Co self-diffusion coeff. obey the equation: ${D^{Co}}_{Co} = e^{-61,900/RT}$ cm.^{2}/sec.; for dil. Co diffusing into Ni: ${D^{Co}}_{Ni} = 1{\cdot}46\ e^{-68,300/RT}$ cm.^{2}/sec. 7 ref.—E. N.

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

***Elementary [Magnetic] Domains in a Cobalt Single Crystal.** L. F. Bates (*J. Phys. Radium*, 1951, **12**, (3), 322–324).—[In French]. Magnetic powder diagrams (Bitter diagrams) obtained on a Co single crystal, either demagnetized or in a magnetic field $\parallel$ or $\perp$ to the crystal axis, are illustrated and discussed.—J. S. G. T.

***Magnetic Saturation Intensity [of Cobalt] and Some Other Related Measurements [in Iron–Nickel–Aluminium Alloys].** W. Sucksmith (*J. Phys. Radium*, 1951, **12**, (3), 430–435; discussion, 435–436).—[In English]. The variation of the saturation intensity of a Co single crystal was studied at 300°–600° C. This variation shows the usual trend up to 400° C., where a phase change takes place. The calculated value of the Curie temp. of cubic Co is $\sim$1404° K., above which temp. the hexagonal crystal lattice is replaced by a b.c.c. lattice. At the transition temp., the magnetization rises discontinuously by $\sim$1·5%. Variation of saturation magnetization with degree of order in binary alloys is discussed with ref. to the Fe–Ni–Al series of alloys. The importance of measurements of magnetic saturation in the study of phase equilibria is stressed. 15 ref.—J. S. G. T.

***Creep of Annealed High-Purity Copper : Effects of Stress, Temperature, Mechanical and Thermal History.** W. D. Jenkins and T. G. Digges (*Metal Ind.*, 1951, **78**, (20), 403–405).—An account of an investigation of creep tests on annealed O-free, high-conductivity Cu under tension at 110°, 250°, and 300° F. (45°, 120°, and 150° C.) now being carried out at the U.S. National Bureau of Standards.—J. H. W.

***Grain-Boundary Effect in Surface-Tension Measurement [of Solid Copper].** Harry Udin (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 63).—A note. In the earlier investigation of the surface tension of solid Cu (*ibid.*, 1949, **185**, 186; see *Met. Abs.*, 1948–49, **16**, 531), it was assumed that the grain-boundary energy in a pure metal was very small in comparison with the energy of the external surface, and, accordingly, the internal surface energy of the coarse-grained specimens was neglected. However, experimental evidence has since shown that far from being small, the grain-boundary energy is generally $\sim\frac{1}{3}$ the energy of an external surface, and, therefore, there is a significant error in the reported const. When the grain-boundary energy is taken into account, the surface tension of solid Cu at its m.p. is found to be 1650 ± 100 dynes/cm., instead of 1370, as previously reported; the new temp. coeff. is $-$ 0·55 dynes/cm./°C. The details of an experiment to determine the grain-boundary tension are being worked out. 4 ref.—E. N.

***Diffusion of Copper Along the Grain Boundaries of Nickel.** R. S. Barnes (*Nature*, 1950, **166**, (4233), 1032–1033).—It is shown that the rate of diffusion of Cu into Ni in the solid state is greater along the grain boundaries than through the grains. Cu and Ni strips were annealed in H, to remove oxide films, and hot-rolled together. The sandwich was annealed at 1000° C. for 36 hr. in A. The cross-sectional surface was then polished and etched electrolytically. Photomicrographs show greater diffusion along the grain boundaries than across them. Twin boundaries have no effect on diffusion rates. The diffusing Cu atoms do not lock the Ni grain boundaries and so prevent grain growth during annealing. Rough estimates of the ratio of diffusion rates along the grain boundaries and across them give values between 1·5 and 4. The vol. diffusion of Cu into Ni is not dependent on the orientation of the grain.—B. R. T. F.

***The Ferromagnetism of Metallic Dysprosium.** F. Trombe (*J. Phys. Radium*, 1951, **12**, (3), 222).—Metallic Dy is characterized by a ferromagnetic Curie point and distinct saturation effects at $\sim$105° K. At 179° K. a very pronounced magnetic anomaly, possibly attributable to a second low-temp. ferromagnetic state, stable at $>$175° K., occurs.—J. S. G. T.

Gallium : A By-Product Metal. A. P. Thompson and H. R. Harner (*J. Metals*, 1951, **191**, (2), 91–94).—A review of its history, ores, extraction, phys. and chem. properties, and uses. Prodn. in the U.S.A. was estimated to have been $\sim$200 lb. in 1948. 11 ref.—E. N.

***Effect of Heat-Treatment on the Electrical Properties of Germanium.** H. C. Theuerer and J. H. Scaff (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 59–63).—Ingots of Ge were prepared by reduction of GeO$_2$ at 650° C. by H, fusion of the resulting sponge metal at 1000° C.—when the purity was 99·8%—final melting under an atmosphere of He, and controlled solidification from the bottom upwards. Under these conditions normal segregation takes place and most of the impurities concentrate in the last region to freeze. The major part of the ingot is, therefore, purer than it would be if cooled randomly; the orderly advance of the freezing surface gives a succession of surfaces of const. compn., and the distribution of the impurities follows from the shape of the solidifying interface at different stages of solidification. Conductivity and resistivity measurements show that in the as-cast state the ingot is an *n*-type conductor with peak reverse voltages ranging from 50 to 200 V., depending on the position in the specimen. From the similarity of these voltage contours to the contours of freezing surface, it is evident that the peak reverse voltages are systematically related to purity, being highest at the bottom of the ingot—the purest—and lowest at the top. On heating the specimen, contained in a quartz tube under an atmosphere of He, in the range 500°–900° C. for 24 hr. and then cooling in air, conductivity measurements show that conversion to *p*-type conduction takes place, the purest metal transforming at the lowest temp., and the general curvature of the *n*/*p* boundaries again suggests that these are regions of equal purity formed during the freezing process. At temp. as high as 900° C., however, the impurity-rich top is still unconverted to *p*-type. The thermal reversion is completely reversible, for after heating at 500° C. for 24 hr. the whole ingot is again *n*-type, with the same characteristics as the original, thus showing that no substantial diffusion of impurities takes place during the heat-treatment process. The results are interpreted in terms of changes in the balance between acceptors and donors resulting from an increase in acceptor levels as the temp. of heat-treatment is increased. Two possible mechanisms for the increase in acceptor concentration are given, one postulating a change in solubility of an acceptor impurity, the other postulating the formation of acceptor centres by thermally produced lattice defects, but further experiments are necessary to determine which mechanism is correct. 5 ref. —E. N.

***On the Crystallization of Very Thin Films of Gold.** Antoine Colombani and Gaston Ranc (*Compt. rend.*, 1951, **232**, (14), 1344–1346).—The behaviour of very thin films of Au deposited, by thermal evaporation in a vacuum produced by an oil-diffusion pump in the presence of liq. air, on amorphous (Plexiglas) or crystalline (rock salt) supports at normal temp. was studied. (1) *On Plexiglas.*—With a film thickness $<$50 Å., the resistivity is an increasing function of the time. (2) *On rock salt.*—Below 35 Å. the resistivity increases with the time; above 35 Å. it diminishes, and the resistivity of the deposit on rock salt is always below that of the deposit on Plexiglas. C. and R. draw a number of conclusions from these facts and from an investigation of the films with opt. and electron microscopes.—J. H. W.

***Lattice-Distortion Spectrum of Evaporated Gold.** P. G. Wilkinson (*J. Appl. Physics*, 1951, **22**, (4), 419–423).—Thin Au films were deposited on liq.-air-cooled surfaces by evaporation in various low pressures of N, and the energy spectrum of the lattice distortions responsible for the high initial resistivity was determined by following the change of resistivity as the specimen was gradually allowed to warm. In the range investigated, max. were found at 80, 100, and 120 $\times$ 10^{-14} erg. The number of each of these types decreased as the N pressure during deposition was raised. The final resistivity depended on the film thickness in a manner which agreed with theory.—R. W. C.

***Contact Potentials of Evaporated Iron Films in Air and Nitrogen at Low Pressure.** Norman Hackerman and Leland L. Antes (*Science*, 1950, **112**, (2912), 471).—The variation of contact potential with time for metal surfaces prepared by evaporation is discussed, with special ref. to Fe. In connection with the studies, electron-diffraction patterns of evaporated metal films produced within an electron-microscope column were continuously observed from the time of their deposition, it being possible to follow the formation of an oxide pattern as a function of time and pressure. The first perceptible change in Fe at a pressure of 27 μ Hg occurred in 1½ min., as compared with 27 min. for Ni.—J. R.

***The Dissolution of Metals Over the Temperature Range 25° to −60° C. I.—The Dissolution of Magnesium in Hydrochloric Acid–Methyl Alcohol–Water Solutions.** A. B. Garrett and R. R. Cooper (*J. Phys. Colloid Chem.*, 1950, **54**, (4), 437–444).—G. and C. have measured the dissolution rate of Mg in HCl–CH_3OH–H_2O soln. in the range 25° to −60° C. The reaction was found to be first-order with respect to concentration of HCl. The effect of stirring speed indicated that the reaction is largely diffusion-controlled.—D. M. D.

***High-Temperature Structure and Thermal Expansion of Some Metals as Determined by X-Ray Diffraction Data. I.—Platinum, Tantalum, Niobium, and Molybdenum.** (Edwards, Speiser, and Johnston). See col. 29.

***On Compact-Dispersed Silver.** George-Maria Schwab (*J. Phys. Colloid Chem.*, 1950, **54**, (4), 576–580).—S. has measured the pore and particle size of an Ag compact, produced by the electrolytic reduction of fused AgCl, in three ways: (i) by catalytic action, (ii) from the sp. penetration, and (iii) by X-rays. It is found that the mean pore width $\simeq$ mean particle width $\simeq 5 \times 10^{-4}$ cm.—D. M. D.

Technetium: Element 43. J. C. Hackney (*J. Chem. Education*, 1951, **28**, (4), 186–190).—Early predictions of an element of atomic no. 43 in the same periodic sub-group as Mn are mentioned, and theoretical conclusions regarding its structure, arrived at before the first prepn. of the element by Perrier and Segrè in 1937, are briefly reviewed. The preparation, isolation, and chem. properties of Tc are discussed, and the nuclear properties of the isotopes are tabulated. 55 ref.—J. R.

***Electrical Properties of Grey Tin.** G. Busch, J. Wieland, and H. Zoller (*Helv. Phys. Acta*, 1951, **24**, (1), 49–62).—[In German]. Grey Sn of high purity (99·9963%) was prepared by prolonged cooling of spectroscopically pure metallic Sn. The variation of conductivity with temp. was then determined by measuring the Q factor of a coil contg. a core of the grey Sn powder or its alloys, at frequencies up to 30 Mc./s. ($1/Q = R/\omega L$, where R = ohmic resistance of the core, L = its self-inductance, and ω = frequency). The following results were found: (1) conductivity at 0° C. = $5 \times 10^3\ \Omega^{-1}$ cm^{-1}., (2) a decrease in conductivity with decrease in temp., according to the exponential law, (3) an intrinsic conductivity, with an activation energy of ~0·1 eV., down to −100° C., (4) large increases in conductivity at low temp., with increasing addn. of Al (up to 0·489 at.-%) as alloying element. Measurements were also made, by conventional D.C. methods, of the Hall effect and of the remarkably high changes of resistivity in magnetic fields. It is concluded that grey Sn is a semiconductor of high elect. resistance, with properties very similar to those of Si and Ge. 17 ref.—E. N.

***Influence of Gas Adsorption on the Rate of Reduction of Nickel Oxide by Tungsten.** Jean-Pierre Kiehl (*Compt. rend.*, 1951, **232**, (18), 1666–1668).—A mixture of Ni oxide and W powders was placed in a SiO_2 reaction tube, degassed at 450° C. in a vacuum of 10^{-5} mm. Hg, and then annealed for 15 min. at different temp. in various gases and in a vacuum of 6×10^{-6} mm. Hg. The gases used were: He, Ne, A, N, and CO_2. Below 780° C. the rate of reduction of the Ni oxide by W varies according to the nature of the gas adsorbed. At const. temp. and pressure the rate is an increasing function first of the temp. of liquefaction of the gas and then of the quantity of the gas adsorbed. The $\alpha \rightarrow \beta$ transformation of W, determined by a dilatometric and X-ray investigation, is indicated by the variation of the chem. reactivity of the metal: the temp. of the transformation is lower the greater the quantity of gas adsorbed. Above 780° C. the rate of reduction of the oxide is virtually the same for all the gases studied in consequence of the very marked diminution of the adsorption. Above 850° C. the gases no longer influence the rate of reduction, probably because of the rapid diminution of the reacting surface of the grains by recrystn. The effect of a vacuum is very small; it can, therefore, be assumed that, in the temp. range studied, the rate of reduction tends towards zero as the pressure approaches zero. These results confirm the results of previous work (Forestier and K., *ibid.*, 1949, **229**, 197).—J. H. W.

***The Production of Uranium and Thorium in the Powder Form.** P. Chiotti and B. A. Rogers (*U.S. Atomic Energy Commission Publ.*, 1950, (**AECD-2974**), 21 pp.).—Both metals can be produced as powder by forming powdered hydride from the solid metals and then decomposing the hydride under conditions such that no sintering occurs. UH_3 is formed most rapidly at ~225° C., and curves are shown relating the rate of reaction and the dissociation pressure of UH_3 with temp. H may be purified by passing over Cu at 600° C. and dried with $Mg(ClO_4)_2$ or by passing over U turnings at 600°–700° C. An apparatus and procedure are described for converting 100–250-g. charges to powder, the size of the particles being controlled by the operating conditions. Th forms two hydrides, ThH_2 and $ThH_{3\cdot75-4}$, so that the procedure for Th is more complicated and consists of four operations: (1) formation of ThH_2 at 600°–650° C.; (2) formation of higher hydride at <320° C.; (3) decomposition of higher hydride at <500° C. under atm. pressure; and (4) decomposition of ThH_2 at 700° C. at low pressure. Crystallographic and thermodynamic data relating to the three hydrides are included.—B. W. M.

Zirconium. G. L. Miller (*Murex Rev.*, 1951, **1**, (8), 184–193).—M. discusses the prodn. of Zr from $ZrCl_4$, including the purification and densification of the tetrachloride, its reduction with Mg, the sepn. of $ZrCl_4$, and the prepn. of Zr metal from the reduced sponge. The fabrication of the Zr metal is also described. The properties of Zr, including corrosion-resistance, machinability, plastic working, &c., are outlined. M. concludes with a discussion of the appn. of Zr, and the prodn. of the metal in the U.K. 23 ref.—R. J.

***A Metallurgical Investigation of Silver Chloride.** R. D. Moeller *et al.* (*U.S. Atomic Energy Commission Publ.*, 1950, (**AECU-811**), 25 pp.).—The results are given of various phys. measurements on AgCl, including a general survey of the literature. The material was cast into various types of ingot, and microhardness tests were made after a series of cold reductions. Metallographic examinations were made on polished sections, using polarized and non-polarized reflected light and also transmitted light. The tensile properties of sheet material were measured by conventional methods, and the alloying characteristics of AgCl with other alkali halides were considered. In many respects, the phys. properties of AgCl are very similar to those of the common metals, the main differences, such as those in elect. conductivity and nature of the bonding, being due to the ionic nature of the crystal.—B. W. M.

Mechanical Properties of Metals. H. W. Gillett (*Machine Design*, 1951, **23**, (3), 143–148, 198, 200).—G. stresses that very few figures obtained from conventional mech. tests have any appn. in design, as service behaviour does not utilize the exact property thus measured. The normal mech. tests are critically discussed, and examples are quoted showing how the results are frequently inconsistent. For special design purposes the properties stated to be of greatest value are yield strength and elastic modulus. The true stress/strain diagram is useful, as it reflects the work-hardening behaviour of the material, a feature of value in plastic forming processes. G. discusses how valuable information on usable deformation, combined stress, and directionality can be applied in design problems. Other subjects discussed include the limitations of tensile data, sampling, hardness, bend and transverse tests, toughness and brittleness, the effects of repeated stress, and the influence of high temp. It is emphasized that conventional tests give no information on a large number of properties, and service tests are essential.—D. M. L.

Hardness and Wear. Wilhelm Späth (*Metalloberfläche*, 1950, [A], **4**, (12), 177–180).—A review of the relationships between the mech. properties known as "hardness" and "wear," factors which can be defined and measured in various ways. 11 ref.—E. N.

***The Entropies of Melting of Metals.** R. A. Oriani (*J. Chem. Physics*, 1951, **19**, (1), 93–97).—The melting process can be visualized as consisting first of an expansion and then a disordering of the crystal structure. The entropy contribution from the first effect is calculated by analogy with the process occurring in a vol. expansion within an individual phase, the entropy contribution from the melting disorder being then obtained by difference from the measured entropy of fusion. Tentative conclusions of heuristic value are drawn from the relative magnitudes of the various quantities. A simple model for the melting process is adopted, making possible a correlation between the amount of disordering during melting and the structural contribution to the entropy of melting. —J. R.

***Relation Between the Atomic Vibrations of Metals in the Solid State and Their Heat of Fusion.** Léon Jollivet (*Compt. rend.*, 1951, **232**, (10), 966–968).—It is known that the atomic vibrations of metals are closely related to the sp. heat (Einstein and Debye) and to the melting temp. (Lindeman); J. showed that there is also a simple relation between the atomic vibrations of solid metals and their heat of fusion. He calculated, ν, the frequency limit of the atomic vibrations by three methods and plotted this factor against n, the principal quantum number of the outer layer of the atom for a number of metals. The f.c.c. and h.c.p. metals lay on one curve and the b.c.c. on another. These curves were straight lines up to $n = 5$, above which the values were dispersed. The straight lines corresponded to the equation: $H = \alpha(n - 1)Nh\nu$, where H in cal./g.-atom is the heat of fusion, α is a const. equal to 1·63 for the f.c.c. and h.c.p. metals and to 0·90 for the b.c.c. The heats of fusion are thus proportional to the quantum of vibrations of the atoms for the max. frequency of the lattice. The coeff. of proportionality depends on two terms, the one characterizing the crystal structure and the other the energy level of the outer electrons of the atoms. —J. H. W.

Development of Improved Metals to Determine Success of Atomic Energy Programme. —— (*Steel*, 1950, **126**, (15), 91–92, 94, 97).—Prepared from data presented in the 7th Semi-Annual Report of the Atomic Energy Commission, Jan. 1950. Amongst the subjects reviewed are strength of metals and creep, diffusion in solid metals, thermodynamics of metals, corrosion, and the effect of radiation.—E. J.

†Effects of Radiation on Materials. J. C. Slater (*J. Appl. Physics*, 1951, **22**, (3), 237–256).—A declassified excerpt from an American survey of the field, dealing with the subject in a general way, and without ref. About a third is taken up by a concise survey of the atomic and electronic factors which determine a number of important phys. and mech. properties of solids, including metals. There follows a section sketching the present status of collision theory as applied to the slowing down of fast particles in solids; the particular point is made that radiation damage must be almost entirely due to displaced nuclei rather than to electronic excitation, and for this reason heavy and relatively slow particles should wreak most damage. The considerations advanced are then applied separately to metals, semi-conductors, ionic solids and ceramics, and to molecular solids. The close analogy between work-hardening and radiation damage is emphasized, and the relevance in this connection of the recrystn. temp. is stressed. Order–disorder changes, pptn.-hardening, surface removal, and alloy formation through retention of fission fragments are discussed. In comparison with the other categories of materials, radiation damage to metals would be expected to be small, particularly for metals with a low recrystn. temp. —R. W. C.

†Internal-Friction Measurement: Method of Investigating Structural Changes in Metals. [—I.–III.] A. S. Darling (*Metal Ind.*, 1951, **78**, (12), 223–225; (14), 271–273; (15), 291–292).—A method of obtaining information on the damping of metals, unobtainable by other means, depends on the effect that internal changes have on the relationship between stress and strain. Zener ("Elasticity and Anelasticity of Metals," University of Chicago Press, 1948) has carried out an analysis leading to general equations on the assumption that an initial stress is applied to a specimen and maintained const. The stress will relax with time owing to the progress of various changes in the microstructure, and the time of relaxation will be uniquely determined by the factor causing this stress reduction. D. derives the general equations and discusses the effect of temp. on the stress/strain relationship and of thermo-elastic coupling on the damping capacity of metals. He concludes that although internal-friction measurements offer a method of investigating changes in structure of metals under the action of stress, theory is much in advance of reliable experimental results and considerably more systematic work is required. 40 ref.—J. H. W.

The Electrostatic Component of the Force of Friction. R. Schnurmann (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 62–66; discussion, 66–68).—Experimental evidence, previously available, suggests that jerky ("stick–slip") motion of a frictional contact operated at a const. rate is of electrostatic origin. The electrification of frictional contacts is illustrated by discharge currents made visible on the screen of a cathode-ray oscilloscope. With a metallic friction element loaded against a rotating steel disc, rapid oscillations appear on the screen; their amplitude is independent of load and speed. In discussion, *J. S. Courtney-Pratt* maintained that there is no evidence that elect. forces form any significant part of frictional forces. 12 ref. —J. S. G. T.

***Surface Welding and the Extreme-Pressure Lubricant.** G. I. Finch and R. T. Spurr (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 79–81; discussion, 81–82).—When two metal surfaces are brought into contact, surface oxide films prevent intermetallic welding; the films may be ruptured by sliding. Cu single crystals weld more readily than similar crystals mech. polished to form work-hardened surface layers, or than electropolished polycryst. materials where deformation in each crystal grain is restricted at its boundary. Al, which is harder than Sn, welds much more easily than Sn to Cu, since Cu_2O is worn away more quickly by hard Al_2O_3 than by SnO. The ease with which metal surfaces weld on sliding depends upon their rigidity, and the nature and thickness of their surface films. Extreme-pressure additives react with the oxide film to form a layer of chloride, oxychloride, sulphide, &c., which is more fusible than the metal. 5 ref.—J. S. G. T.

***An Investigation of Surface Damage, Using Radioactive Metals.** E. Rabinowicz (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 82–85).—In indust. practice, it is generally more important to reduce wear than to reduce friction. Radioactive hemispherical sliders of Cu, Cd, Zn, Pt, Ag, and stainless steel were slid over a bottom plate, and the transfer of activity to the latter recorded by placing it in contact with a photographic plate which blackened locally owing to the radioactive material torn from the hemisphere during sliding. This method will detect down to 10^{-12} g. in favourable cases. Metal transfer occurs in the form of discrete fragments, and well-lubricated surfaces reduce the transfer, compared with clean surfaces in the ratio 400, or more, : 1. On heating well-lubricated surfaces, the coeff. of friction and metal transfer increase markedly at the m.p. of the surface film. Metal transfer, in contrast to friction, is sensitive to temp. change below the surface film's m.p. 7 ref.—J. S. G. T.

***The Adsorption of Lubricant Films: A Study by Radioactive Tracers.** A. C. Moore (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 54–56; discussion, 56–57).—The chem. reaction between radioactive metal foils, (Pt, Au, Zn, Cd, and Cu), and adsorbed monolayers of stearic acid, ethyl stearate, and octadecyl alcohol, was studied. It is directly verified that when reaction occurs the system shows good lubricating properties at temp. above the bulk m.p. of the lubricant and up to the softening point of the soap. If no reaction takes place, lubrication occurs only below the bulk m.p. of the lubricant. Reaction does not cease with the formation of a monolayer of soap on the metal; if sufficient materials are present, thick films of soap are built up. The weight of metal detected in the tests is of the order

10^{-7} g./cm.^{2} of true surface area. The results obtained by the use of radioactive tracers agree with those obtained by electron-diffraction methods. 16 ref.—J. S. G. T.

***A Study of Boundary Lubricant Films by Electron Diffraction.** J. W. Menter (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 52–53; discussion, 53–54).—The structures and orientations of long-chain hydrocarbon compounds, including fatty acids, and metallic soaps on polished surfaces of Pt, Ag, Cu, Zn, Cd, mild steel, and Ni, were studied by electron diffraction. It is shown that the first monolayer is oriented with the hydrocarbon chains approx. normal to the surface; material above the first monolayer shows a different orientation and structure. Structure and orientation of the monolayer are unaffected by rubbing except in one case, viz. octacosanoic acid. The diffraction patterns remain unchanged over a considerable rise of temp., but at a critical temp., characteristic of the surface films and the chemical nature of the surface, the diffraction patterns change until ultimately reversible two-dimensional melting of the film occurs. Good boundary lubrication requires a closely packed solid film of oriented molecules on the metal surface. 16 ref.—J. S. G. T.

***The Effect of Lubrication on Friction, Wear, and Abrasion.** P. Grodzinski (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 86–89; discussion, 89–90).—Experimental results relating to the effect of lubricant (water) upon the coeff. of friction and abrasion between a cast-iron disc and a lubricated diamond surface are given. Best combinations of abrasive, lap material, and lubricant found by Knight and Case (*Trans. Amer. Soc. Mech. Eng.*, 1915, **37**, 297) are tabulated for abrasives SiC and Al_2O_3 and Cu, cast iron, and steel. 11 ref.—J. S. G. T.

***Lubrication with Materials in the Solid State.** F. T. Barwell and A. A. Milne (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 39–42; discussion, 42–43).—Coeff. of friction between two rubbing surfaces of mild steel loaded at 40 lb./in.^{2} and moving at relative velocities of 0·24 and 9·45 ft./min. using 13 solid lubricants, ranging from the worst (talc, 0·5) to the best (sodium stearate, 0·08), were measured. Only graphite, metallic soaps, and paraffin wax were found to be effective solid lubricants. Results are given of running tests in which mild steel treated by Sn or Cr plating and phosphatizing was continuously reciprocated against mild steel either unlubricated or lubricated with paraffin wax (m.p. 55° C.) A phosphatized surface, lubricated with paraffin wax, gave the best results. It is suggested that the wax has a sufficiently high yield value to prevent contact between the surfaces and that it is sufficiently non-reactive towards the metal to make easy the cleavage at the wax/mild steel interface. In the discussion, *E. A. Smith* referred to apparent contradiction between results obtained by B. and M. and by Clayton relating to graphite lubrication, and indicated how these could be reconciled.—J. S. G. T.

***The Lubrication of Metals by Long-Chain Organic Compounds.** K. V. Shooter (*Physics of Lubrication* (*Brit. J. Appl. Physics Suppl.* No. 1), **1951**, 49–51; discussion, 51).—Results of lubrication tests made with the Bowden–Leben apparatus (*Proc. Roy. Soc.*, 1939, [A], **169**, 371; *Met. Abs.*, 1939, **6**, 206) using fatty acids as lubricants between rubbing surfaces of Cu, Fe, Mg, Zn, and Cd, show that reaction and soap formation must occur and that more than a monomolecular layer of soap must be present on each sliding surface for good lubrication to result. 11 ref.—J. S. G. T.

Theory of D_0 for Atomic Diffusion in Metals. Clarence Zener (*J. Appl. Physics*, 1951, **22**, (4), 372–375).—A recent treatment by Wert and Z. (*Phys. Rev.*, 1949, [ii], **76**, 1169; *Met. Abs.*, 1949–50, **17**, 637) of interstitial diffusion is here extended to self- and substitutional diffusion at low concentrations. The statistical-mech. basis of the theory is presented in detail and a general equation derived: $\Delta S = \gamma\beta H/T_m$, where β is a dimensionless quantity involving an elastic modulus and T_m the melting temp., ΔS is the entropy of activation, H the heat of activation, and γ a const. of order of unity. The const. are evaluated for the vacancy and the ring diffusion models for diffusion (the above equation being shown to hold in both cases). An important conclusion is that ΔS is always positive. The best experimental values of appropriate diffusion const. agree well with the equation, and from the values of γ required to obtain a fit it is concluded that the vacancy mechanism operates in f.c.c. metals, the ring mechanism in b.c.c. ones. —R. W. C.

***Influence of a Transverse Magnetic Field on the Conductivity of Thin Metallic Films.** E. H. Sondheimer (*Phys. Rev.*, 1950, [ii], **80**, (3), 401–406).—The influence of a transverse magnetic field on the conductivity and Hall coeff. of thin metallic films, whose thickness is comparable with the mean free path of the conduction electrons, is considered theoretically. Analogous to the increase of zero-field elect. resistivity, the Hall coeff. of such films is predicted to be greater than the bulk value. If the film thickness is small compared with the mean free path, the resistivity is predicted to increase to a max. with increasing magnetic field, and then to oscillate. —P. C. L. P.

***On the Origin of Intermittent Activation in the Ferromagnetics.** Robert Forrer (*Compt. rend.*, 1951, **232**, (19), 1746–1748).—F. recently introduced the conception of intermittent activation (*ibid.*, 1950, **230**, 1254, 1584), and a study of alloys confirmed that this intermittent activation produces ferromagnetism only when it tends towards the $4p$ state, having filled the $4s^2$ state (*ibid.*, 1950, **231**, 1130, 1444; *Met. Abs.*, 1950–51, **18**, 602). The intermittent activation allows of intermittent dissociation of pairs of electrons in the $3d$ state; its cause is attributable to the magnetic moment of the atom itself, and the total intermittent activation, a, is proportional to the number, n'', of pairs of electrons and to the number, n', of single electrons, so that a = const. $\times\, n'' \times n'$. In the series of elements: V–Cr–Mn–Fe–Co–Ni, n'' increases from 0 to 5, while n' decreases from 5 to 0; a thus has a parabolic form as a function of the number of electrons in the $3d$ state. The number of interactions, N, that can be inferred from the Curie points of some combinations of the MnSb type (θ = 70° C. for CrTe, 314° C. for MnSb, and 328° C. for $FeSi_{1\cdot 1}$) confirms the parabolic form. For the normal ferromagnetics, on the other hand, the number of interactions is given by N = const. $\times$ $(n'' - 2)n'$; so that metallic Mn is in effect not ferromagnetic, while Fe has a strong molecular field. Metals and alloys of the f.c.c. type (Co-γ, Ni, Fe–Ni-γ, Co–Ni-γ) have a permanent activation of at least a half electron, and the parabola is thus displaced a half-unit to the right. The intermittent activation is thus primarily an internal effect of the atom, depending on the atomic number. It is probable that it exercises by its intercepts an appreciable effect on the spatial distance of neighbouring atoms and consequently on the choice of the crystal lattice.—J. H. W.

***The Electrical Resistance at Low Temperatures of Some Ferromagnetic Metals [and Alloys] in a Magnetic Field.** C. J. Gorter (*J. Phys. Radium*, 1951, **12**, (3), 279–280; discussion, 280).—A brief, provisional account is given of the researches of J. Smit relating to the elect. resistance of Ni, Fe, and Ni–Fe, Ni–Co, and Ni–Cu alloys in magnetic fields up to 22 kOe. at temp. between atmospheric and that of liq. H. The effect upon resistance of field orientation relative to the elect. current increases considerably at low temp., and attains an approx. value of a multiple of 10% in some alloys at the lowest temp. In the case of the pure metals, however, the effect upon resistance of field orientation relative to the current is very small, and the increase of resistance due to the magnetic field has its normal value. In the case of the alloys, the slope, dR/dH, on the descending part of the R/H graph in strong fields vanishes at very low temp. except in the case of the 42 : 58 Ni–Fe alloy, doubtless due to abs. magnetic saturation being very nearly attained. The 76 : 24 Ni–Fe alloy shows anomalous behaviour attributed to the development of superstructure in the alloy.—J. S. G. T.

†Thermal Effects Associated with Magnetization Processes. L. F. Bates (*J. Phys. Radium*, 1951, **12**, (3), 459–469; discussion, 469–470).—[In English]. Results of experimental studies of thermal effects associated with magnetization processes, in low and moderate fields, during the last 25 years, and B.'s apparatus used for this purpose, are discussed. Many of the results are discussed in terms of a theory due to Stoner and Rhodes (*Phil. Mag.*, 1949, [vii], 40, 481; *Met. Abs.*, 1949–50, **17**, 9). The main results obtained are: (1) thermal changes for initial magnetization of all the pure ferromagnetic metals are the same, initial cooling being followed by magnetocaloric heating; (2) there is a "folding back" effect in the curves of thermal changes for a magnetization cycle; (3) internal strain in Ni produces very large effects on the magneto-thermal curves; and (4) metals and alloys show interesting differences. An attempt is made to deduce thermal behaviour from available magnetic, magnetocaloric, and magnetostriction data. The result is satisfactory in the case of Ni under severe internal strain, but less so with well-annealed materials, where effects of crystal anisotropy are important and difficult to estimate. 14 ref.—J. S. G. T.

***Magnetic Drag.** Louis Néel (*J. Phys. Radium*, 1951, **12**, (3), 339–350; discussion, 350–351).—Magnetic drag in magnetic materials is separated into a reversible and an irreversible component and the parts played by these in the magnetization of substances are discussed in terms of two fictitious magnetic drag fields varying with time and superposed upon the applied magnetic field. The reversible field is attributable to progressive stabilization of spontaneous magnetization which is related to material diffusion within ferromagnetic substances. Magnetostriction constraints are insufficient to explain this diffusion stabilization, which, it is suggested, is attributable to magneto-cryst. coupling, for which a mechanism is proposed. Irreversible drag, on the other hand, is attributable to thermal fluctuations, arising in various ways, of which the most important, apparently, is due to fluctuations of the internal dispersive field associated with variations in the direction of spontaneous magnetization. Numerical data relating to the resp. magnitudes of the two types of magnetic drag are given. 34 ref.—J. S. G. T.

***Irreversible [Magnetic] Drag in Weak [Magnetic] Fields.** Jean-Claude Barbier (*J. Phys. Radium*, 1951, **12**, (3), 352–354).—Experiments carried out with a number of ferromagnetic metals and alloys, including Fe, Ni, Co, Alnico V, Co–Zn, and Ni–Zn ferrites, designed to study the variation of magnetization with time, and the effect of an alternating magnetic field upon remanent magnetization in weak fields, confirm the hypothesis of an irreversible magnetic drag due to thermal fluctuations, contained in Néel's theory of magnetic drag (preceding abstract).—J. S. G. T.

***The Coercive [Magnetic] Field.** Louis Weil (*J. Phys. Radium*, 1951, **12**, (3), 437–447; discussion, 446).—Becker's and Kersten's resp. contributions to the theory of the coercive magnetic field of ferromagnetics, and Néel's introduction of internal demagnetizing fields are briefly reviewed, and some confirmations of the theories are discussed. The coercive field of finely granulated materials is attributed to: (1) internal stresses, and (2) magnetic anisotropy and anisotropy of form of the grains; experimental results confirm the roles played by these two factors. Upper and lower grain dimensions and the part played by packing in the prodn. of the coercive field are discussed with the aid of the theory and experimental data. 61 ref.—J. S. G. T.

†**Magnetic Domain Patterns.** R. M. Bozorth (*J. Phys. Radium*, 1951, **12**, (3), 308–320; discussion, 320–321).—[In English]. The technique and interpretation of magnetic domain patterns, as revealed by magnetic powder diagrams, are discussed at considerable length by ref. to illustrations of domain patterns characterizing Fe in various states of magnetization, and after mech. or electrolytic polishing, and under increasing tension, Co, Co–Ni alloy, Fe–Si alloy (showing that domains sometimes cross grain boundaries), and Alnico V. 20 ref.—J. S. G. T.

Delimitation of the Rayleigh Domain in Alternating Fields in the Light of Recent Theoretical and Experimental Work. I. Epelboim (*J. Phys. Radium*, 1951, **12**, (3), 361–370; discussion, 370–371).—Departures from Rayleigh's laws relating to magnetization in weak fields, shown by some materials of high magnetic permeability, are referred to. E. has shown that strips and wires of high-permeability material, polished electrolytically, are heterogeneous in the direction of their thickness, probably owing to irregular diffusion of atoms during thermal and mech. treatment, and he attributes departures from Rayleigh's laws shown by these materials to this heterogeneity, which sets a field limit to the applicability of these laws. 39 ref.—J. S. G. T.

Dynamics of the Bloch Boundary and High-Frequency [Magnetic] Permeability. R. Becker (*J. Phys. Radium*, 1951, **12**, (3), 332–337; discussion, 337–338).—A Bloch boundary in a ferromagnetic crystal is the crystal region in which the spontaneous magnetization of the crystal changes from one permissible direction to another. In the region of initial magnetic permeability, magnetization takes place as a result of an increase of domains favourably oriented at the expense of unfavourably oriented domains. This necessitates a displacement of Bloch boundaries separating the domains, and it is shown that these boundaries possess an inertial mass which can be evaluated exactly. At very H.F., the Bloch boundaries are unable to follow variations of the magnetic field, owing to their inertial mass. This hypothesis of inertial mass of the boundaries is used to discuss the observed disappearance of magnetic susceptibility. Among matters discussed are: friction due to internal induced current, Snoek's theory of ferromagnetic resonance (*Physica*, 1948, **14**, 207), and inhomogeneity of the magnetic field due to a skin effect. 12 ref.—J. S. G. T.

*†**Collective Electron Ferromagnetism in Metals and Alloys.** E. C. Stoner (*J. Phys. Radium*, 1951, **12**, (3), 372–387; discussion, 387–388).—[In English]. A theory of ferromagnetism, based on the Fermi–Dirac statistical electron theory, in which it is assumed that the effective electrons (or holes) are arranged in a parabolic strip, and that the energy of exchange is proportional to the square of the magnetization, is developed. Under certain restrictive conditions, forms of the magnetization curves corresponding to those in Weiss's theory are derived, and it is shown that spontaneous magnetization occurs below a certain temp. which is the Curie point. Saturation is incomplete even at 0° K. In the case of Ni, with approx. 0·6 hole/atom in the *d* band, marked agreement is found between theoretical deductions and experimental values for the variation of magnetic susceptibility with temp., the variation of spontaneous magnetization, and the sp. electronic heat both at low and at high temp. By a natural development of the theory, a satisfactory explanation can be given of the properties of many Ni alloys. Magnetic saturation is incomplete at 0° K. in the case of many alloys of various compositions and for Fe. In the case of Fe and Co the assumption of a parabolic form of electronic band is not a satisfactory approximation. The influence of various forms of band is being studied. 28 ref.—J. S. G. T.

Ferromagnetic Resonance. C. Kittel (*J. Phys. Radium*, 1951, **12**, (3), 291–300; discussion, 300–302).—[In English]. The phenomenon of ferromagnetic resonance absorption was discovered experimentally by Griffiths in 1946 (*Nature*, 1946, **158**, 670; *Met. Abs.*, 1947, **14**, 4), after theoretical discussion of the matter by Gans and Loyarte (*Ann. Physik*, 1921, **64**, 209) and by Dorfmann (*Z. Physik*, 1923, **17**, 98; *J. Inst. Metals* (Abstracts), 1925, **34**, 493). A math. theory of the phenomenon is developed by the aid of classical resonance theory, followed by a quantum-mech. interpretation of the phenomenon. Short-wave resonance and measurements of the gyromagnetic coeff. are correlated, and recent experimental results are briefly discussed. A theory of the effect of frequency upon the initial permeability of ferrites, which agrees with recent experimental results, is developed. 38 ref.—J. S. G. T.

***A New Conception of the Electronic State of Ferromagnetics: Introduction of the Notion of Intermittent Activation.** Robert Forrer (*J. Phys. Radium*, 1951, **12**, (3), 402–413; discussion, 413).—A math. theory of ferromagnetism, applicable to metals and alloys and based upon the intermittent activation of adjacent electrons in the crystal lattice, is developed. The theory enables non-ferromagnetics (Cr, Mn) to be differentiated from spontaneous ferromagnetics (Fe, Co, Ni) and provoked ferromagnetics (alloys of Cr and Mn). The magnetic moments of numerous ferromagnetic alloys are calculated. Values of the Curie const. of Fe and Ni, calculated by aid of the theory, agree with experimental values. The magnetic moments of Fe–Cr and Fe–V alloys are interpreted in terms of the antiferromagnetism characterizing Fe and either Cr or V together with a superposed intermittent electronic activation. The appearance of ferromagnetism at a definite compn. in the alloys Fe–Ni-γ, Fe–Co-γ, Pt–Co, Pt–Cr, Pt–Mn is attributable to progressive intermittent electronic activation. 17 ref.—J. S. G. T.

Rotations [of Magnetic Moments] During Magnetization. (Initial Magnetization, Coercive Field). Charles Guillaud (*J. Phys. Radium*, 1951, **12**, (3), 492–497).—Values of the initial magnetization, (α), and of the coercive field, (H_c) are deduced theoretically from a consideration of the rotation of elementary magnetic moments against the magneto-cryst. and tensile stresses produced by an external magnetic field. Theoretical values of α and H_c and some of the const. occurring in the theory are compared with experimental values obtained for the alloys MnBi, Mn_2Sb, Co and Co ferrites, with fairly satisfactory agreement in some cases. The results confirm magnetic moment rotation as the mechanism of magnetization. 16 ref.—J. S. G. T.

The Gyromagnetic (Magneto-Mechanical) Ratios of Some Ferromagnetics of the Iron Group. André J. P. Meyer (*J. Phys. Radium*, 1951, **12**, (3), 303–304; discussion, 304).—Values of the Landé factor, g, of Fe and Ni, determined by measurement of the Einstein–de Haas effect by the method of forced resonance employed by Cœterier (*Helv. Phys. Acta*, 1935, **8**, 522) confirm the value $g = 2{\cdot}00$ found by Sucksmith, Chattok, and Bates, and indicate that electron spin is the sole source of effective magnetic moment in metals. The same conclusion is reached in the case of the isoelectronic alloys of Fe with Ni or Co, FeNi, FeCo, and Fe_3Ni, which have very different Curie points and saturation magnetic moments. The difference of moments of these isoelectronic alloys cannot be attributable to a variable orbital moment, as suggested by Forrer (*ibid.*, 1940, [vii], **1**, 74; *Met. Abs.*, 1940, **7**, 205). In the discussion, *E. C. Stoner* directed attention to difficulties arising from these conclusions in interpreting some of Barnett's quite different results, and in explaining the phenomenon of magneto-cryst. anisotropy in the absence of orbital moments.—J. S. G. T.

Recent Developments in the Theory of Antiferromagnetism. J. H. Van Vleck (*J. Phys. Radium*, 1951, **12**, (3), 262–274; discussion, 274).—[In English]. A review with 17 ref.—T. G.

†High Negative Molecular Fields, and the Different Mechanisms of Passage from Antiferromagnetism to Weiss Paramagnetism. G. Foëx (*J. Phys. Radium*, 1951, **12**, (3), 153–159; discussion, 159–160).—Metals, oxides, and salts may possess magnetic properties intermediate between the paramagnetism, independent of temp., and Weiss paramagnetism. They obey Weiss's law, but with values of the Curie const., (C), much too large to have the customary significance, and very high values of $\Delta(= -\theta°$, where $\theta°$ is the Curie temp.). These facts are illustrated by the case of γ-Fe, studied by Sucksmith and Pearce and others (*Proc. Roy. Soc.*, 1938, [A], **167**, 189; *Met. Abs.*, 1938, **5**, 575). At temp. below 0° C., Pt in its magnetic behaviour resembles γ-Fe. Wucher (*Compt. rend.*, 1950, **230**, 730; *Met. Abs.*, 1949–50, **17**, 705) studied the magnetic properties of solid soln. of diamagnetic metals in Pd and found that values of the Curie const. and of Δ are a function of the electronic concentration in the dissolved metal; contrary to expectation, however, Δ increases with increase of this concentration. It is suggested that the electronic structure of Pd metal is an equilibrium mixture of 32·5 ions in the d^8 state and 67·5 neutral atoms in the d^{10} state. Structures of binary alloys of Pd with Au, Ag, Al, and Pb are briefly discussed, and difficulties encountered in applying Weiss's hypothesis of the molecular field are referred to. 32 ref.—J. S. G. T.

***Ferromagnetism in Weak [Alternating] Magnetic Fields, and the Curie Point.** R. Goldschmidt (*J. Phys. Radium*, 1951, **12**, (3), 355–360; discussion, 360).—Desirable properties of ferromagnetic materials employed in the construction of apparatus and appliances for use in weak alternating magnetic fields are briefly referred to. The development of magnetic materials in the last 150 years is briefly reviewed, with illustrative examples of the hysteresis cycles of typical products introduced at various epochs, with especial ref. to Rayleigh cycles, Jordan const., and magnetic viscosity. The anomalous behaviour of ferromagnetics in various fields and at different temp. is briefly referred to, more especially with ref. to initial permeability and losses in the neighbourhood of the Curie point; a material contg. ~4% Si, used in transformers and having two Curie points, and Ni–Fe and Ni–Si alloys, with or without a preferred direction of magnetization, are described. In the discussion, *K. Hoselitz* contradicted G. in maintaining that a substance contg. 4% Si consists of two phases with two Curie points. Alloys of this, and much higher, Si content are single phase and have one Curie point only.—J. S. G. T.

***Ferromagnetism, Paramagnetism, and Cohesive Energy of Transition Metals and Their Alloys.** (Berg). See col. 24.

Non-Classical Reaction Kinetics. Henry Eyring and Peter Gibbs (*Science*, 1951, **113**, (2926), 104–105).—Hardness, elect. resistivity, and magnetism of metals, when treated as rate processes, provide examples of non-classical kinetics. From the appn. of the classical expression for the net sp. rate of surmounting barriers, E. and G. derive an equation for excess velocity in plastic flow, and proceed to interpret the data in terms of crystal geometry and penetration of electrons through the electrostatic barriers separating the initial and final configurations. The interpretation is then extended to a consideration of the effects of alloying elements on plasticity and elect. conductivity, between the mechanisms of which there is an intimate similarity. Curves of the experimental variations of Brinell hardness and resistivity for some binary systems of similar elements are reproduced, the parabolic character of the relationships in general and the similarity of the hardness and resistivity curves for the systems Pt–Pd and Pd–Au being evident.—J. R.

***Fermi–Dirac Functions of Integral Order.** P. Rhodes (*Proc. Roy. Soc.*, 1950, [A], **204**, (1078), 396–405).—Math. After a brief indication of the types of phys. problems in which they arise—transport effects, e.g. elect. and thermal conductivity, and in assemblies of electrons distributed in a band of standard form—an account is given of methods of evaluation of the Fermi–Dirac functions, $F_n(\eta) = \int_0^\infty \{x^n/(e^{x-\eta}+1)\}dx$, for positive integral values of n. 18 ref.—E. N.

2—PROPERTIES OF ALLOYS

***The Mechanical and Physical Properties of [Commercially] Pure Aluminium and Some Aluminium [-Magnesium] Alloys at the Temperature of Liquid Oxygen.** (Mäder). See col. 1.

***The Distribution of Tensile Properties in Extruded Sections [of a Duralumin-Type Alloy].**—I. H. Kostron (*Metall*, 1950, **4**, (21/22), 451–458).—A study is reported of the tensile properties—0·2% P.S., U.T.S., and elongation—as determined on miniature test-pieces, in the directions of the three axes of various shaped, extruded, and age-hardened sections of a Duralumin-type alloy. The tabulated results are discussed in relation to the " press effect," i.e. the higher tensile strength existing in a direction ∥ the extrusion axis than ⊥ it, for which a ratio of 1·21 : 1 has been derived theoretically by Unckel. The ratio of the P.S. (1·19–1·30) was found to be in close agreement with Unckel's ratio, but a lower ratio (1·10–1·19) was found for the U.T.S. Elongation was primarily dependent on the number and shape of the particles of intermetallic compounds within the specimen, for they are found as stringers in sections cut ∥ the extrusion axis and as discrete rounded particles in those cut ⊥ the extrusion axis. Although slight differences in properties were observed between specimens taken at the beginning and towards the end of extrusion, such differences vanished completely if the material was re-aged after another soln.-treatment (500° C., ½ hr., and water quench). Also, after re-ageing, a decrease was observed in the P.S., especially in the direction of the extrusion axis; little difference was observed in the U.T.S.; the elongation increased.—E. N.

***The Influence of Excessive [Solution-Treatment] Temperatures on the [Tensile and] Fatigue Strengths of Aluminium–Copper–Magnesium Alloys.** P. Brenner (*Metall*, 1950, 4, (23/24), 502–504).—Extruded bars of two Duralumin-type alloys were soln.-treated for 1 hr. at 480°–550° C., quenched in H_2O, and aged at room temp.; their microstructures and mech. properties—tensile and fatigue (on notched and on polished test-pieces) were then determined. The mech. properties of the alloy contg. the lower Cu (4·1%), Mg and Mn (0·6% each) were found to decrease slowly with increase in soln. temp. to 540° C., and thereafter, rapidly. The mech. properties of the alloy with higher contents of Cu (4·4%), Mg (1·3%), and Mn (1%) showed some improvement with increase of temp. up to 500° C., but thereafter deterioration was rapid. The deleterious effects of excessive soln.-treatment temp. are shown to be due to the onset of incipient melting of the particles of the intermetallic phases present at the grain boundaries of the alloy, thereby causing notch effects within the specimen; the temp. at which such melting begins decreases with increase in the amount of alloying additions present.—E. N.

***Age-Hardening [of Aluminium, Magnesium, and Copper Alloys].** L. F. Mondolfo (*J. Metals*, 1951, 191, (2), 95).—An A.I.M.M.E. Research in Progress summary. A study—by means of hardness and electrolytic-potential measurements—of the alloy systems of: (1) Al with (*a*) Cu 4%, (*b*) Ag 25%, (*c*) Cu 4, and Mg 1·7%, and (*d*) Zn 6 and Mg 2%; (2) Mg with Al 9%; and (3) Cu with Be 2%, indicates the following tentative conclusions: there are at least two types of age-hardening, depending on the complexity of the transformation required to form the lattice of the precipitate from the lattice of the solid soln. A simple transformation, as, for example, a b.c.c. precipitate forming from a f.c.c. soln., takes place in one step, and only one hardness peak results. With more complex transformations, at least one intermediate lattice is formed and at least two hardness peaks occur. The first hardening results from the transformation from solid soln. to intermediate lattice, and the second hardening from the intermediate lattice to stable precipitate, if there is only one intermediate lattice formed. Hardening does not result from actual pptn., but from the straining of the lattice caused by segregation and atomic movement before the formation of the new phase. Completion of the transformation, by causing a more stable condition, results in softening.—E. N.

***Effect of a Low Rate of Cold Working on the Kinetics of Hardening of Solid Solutions of Aluminium.** Aurel Berghézan (*Compt. rend.*, 1951, 232, (17), 1560–1562).—B. (with P. Lacombe) has already shown the effect of polygonization treatment and of the rate of quenching on the kinetics of decomposition of Al solid soln. (*ibid.*, 1949, 228, 1733, and 1951, 232, 974; *Met. Abs.*, 1949–50, 17, 19; 1950–51, 18, 698); he now considers a third factor, that of light cold-working of the 8% Zn and the 4% Cu alloys previous to ageing. Microspecimens straight from quenching were given reductions of 3, 6, and 10% and then subjected to prolonged ageing at different temp. between 20° and 150° C. before determining their properties with a Chevenard micromachine. Graphs showed that, contrary to the rapid increase of elongation and breaking stress of alloys aged normally at 20° C. after quenching, the hardening of the lightly cold-worked alloys is greatly retarded and even stopped in certain cases. This result was confirmed by measurements of the microhardness, by the Portevin–Le Chatelier effect (*ibid.*, 1923, 176, 507; *J. Inst. Metals* (Abstracts), 1923, 30, 607), and by simple cold bending after quenching or hot before it: in the latter case the effect is still more marked. At temp. above normal, ageing may be slowed up or stopped according to the temp. of annealing and the nature of the solid soln. B. discusses the significance of these results and concludes that light cold work or simple bending on freshly quenched alloys can partly destroy the tendency of supersaturated solid soln. of Al to spontaneous hardening by ageing at room temp. This fact, which is in contradiction to the generally accepted idea that cold work accelerates ageing, suggests that there is a limit of cold work above which this acceleration can be produced.—J. H. W.

***Influence of the Rate of Cooling After Homogenization on the Kinetics of Decomposition of Supersaturated Solid Solutions of Magnesium and of Silicon in Aluminium.** H. Jolivet and M. Armand (*Rev. Mét.*, 1951, 48, (2), 91–96).—To the two well-known factors influencing the kinetics of decomposition of supersaturated solid soln., namely the degree of supersaturation and the temp., should be added a third: the rate of cooling after homogenization. J. and A. have investigated the effect of this third factor in the case of the Al–11·1% Mg alloy annealed at 425° C. for 36 hr. and the Al–0·8% Si alloy annealed at 550° C. for 36 hr., using a Chevenard differential dilatometer. The alloys were maintained for 1 hr. at the temp. of homogenization and cooled in: (1) water at 20° C., (2) boiling water, (3) air, and (4) the furnace (7° C./min.), and then reheated and maintained at the temp. of decompn. The results showed that: (1) pptn. in the zone in which it occurs relatively rapidly is considerably aided by previous ageing, and (2) the effect of this ageing on the rate of pptn. in this zone is much greater than that of a phase previously precipitated during slow cooling.—J. H. W.

***Segregation of Iron and Manganese in Some Aluminium Casting Alloys.** W. H. Glaisher (*Metallurgia*, 1951, 43, (257), 127–131).—The results of investigations into the troublesome gravity segregation which may occur with alloys within or near to the limits of specifications D.T.D. 424 and L.A.C. 112, owing to the presence of undue amounts of Fe and Mn, are presented. In order to ensure uniformity of compn. in castings, alloys of these types should, after melting, be heated to at least 50° C. above the temp. at which segregation would occur on slow cooling, or to a higher temp. if castings are to be made immediately, and should be thoroughly stirred. In using alloys of compn. such that gravity segregation is likely to occur in thick-section castings, as low a pouring temp. as possible consistent with proper running of the castings should be used, provided this does not lead to segregation in the crucible. A procedure for sampling this type of alloy for analysis is given.—F. M. L.

Trends in Aluminium Casting Alloys. Walter Bonsack (*Trans. Amer. Found. Soc.*, 1950, 58, 69–75).—Cf. *Amer. Foundryman*, 1949, 16, (4), 49; *Met. Abs.*, 1949–50, 17, 567.—V. K.

***Characteristics of Some Aluminium–Zinc–Magnesium–Copper Casting Alloys.** W. E. Sicha and H. Y. Hunsicker (*Trans. Amer. Found. Soc.*, 1950, 58, 333–341; discussion, 341–345).—On the bases of a study of compositional variations on casting and mech. properties of Al–Zn–Mg–Cu casting alloys, the following two compn. were found to possess optimum properties: for sand casting, Zn 6·5, Mg 0·7, Cu 0·5, and Ti 0·15%, for gravity die-casting Zn 6·5, Fe 1·0, Cu 0·5, Mg 0·35, and Ti 0·1%, balance Al. These alloys can be aged at room or elevated temp. and exhibit good corrosion-resisting and brazing characteristics.—V. K.

***Investigations to Determine the Suitability of Light Alloys and of Alloys of the Zamak Type for Purposes of Screw Cutting [on a Lathe].** (Farrenq). See col. 67.

Aluminium-Base Friction Alloy. —— (*Fonderie*, 1951, (63), 2408–2409).—Describes the compn. and use of Al-base bearing alloys used in U.S.A., Germany, Switzerland, England, and France, to replace the usual low-m.p. alloys, the Sn-bearing Cu alloys, and the Cu–Pb alloys.—J. H. W.

***On the Production of Powdered Alloys of Chromium and Manganese.** Pierre Jolibois and Bernard Fleureau (*Compt. rend.*, 1951, 232, (13), 1272–1274).—Trombe and Foëx (*ibid.*, 1950, **230**, 2294) obtained cast Cr in a solar furnace; Grube and Flad (*Z. Elektrochem.*, 1939, **45**, 835; *Met. Abs.*, 1940, **7**, 57) obtained the metal very slowly beginning at 950° C. by reducing the oxide in very carefully dried H. J. and F. confirmed the latter work in a reduction apparatus previously used in the reduction of phosphates (J. and J. C. Hutter, *Compt. rend.*, 1949, **228**, 1389). They easily obtained powdered alloys of Fe and Cr with this apparatus by precalcining a mixture of nitrates in a given proportion. Efforts to obtain Mn by the reduction of its oxides with H have always failed (cf. Aoyama and Oka, *Sci. Rep. Tôhoku Imp. Univ.*, 1933, [ii], **22**, 824; *Met. Abs.*, 1934, **1**, 3; and Dannatt and Ellingham, *Discussions Faraday Soc.*, 1948, (4), 126). However, mixtures of the oxides of Fe and Mn can be reduced in a very dry atmosphere. An atomic ratio of unity seems to be the limit above which the reduction becomes impossible, and nothing in Gayler's fusibility diagram (*J. Iron Steel Inst.*, 1933, **128**, 293; *Met. Abs.*, 1934, **1**, 294) will account for this limit. With these alloys of Fe and Mn the reduction of the two oxides takes place simultaneously; with the Fe and Cr oxides, the Fe is reduced first and the presence of the metallic Fe appears to have only a small effect on the reduction of the Cr oxide, which takes place 400° C. higher. In both cases, the powders obtained are very homogeneous, and, later, their X-ray spectra will be studied.—J. H. W.

***Phase Equilibria in an Ordering Alloy [Cobalt–Platinum] System.** J. B. Newkirk, R. Smoluchowski, A. H. Geisler, and D. L. Martin (*J. Appl. Physics*, 1951, **22**, (3), 290–298).—The order–disorder change in Co–Pt alloys near the CoPt compn. was studied by X-ray diffraction, elect.-resistivity measurement, and micro-examination. There is extensive evidence for the existence of a two-phase region on either side of the compn. CoPt, in which ordered and disordered material co-exist. Any one grain generally contains ordered and disordered patches on a fine scale; one of the structures is a Widmanstätten pattern, with (110) planes of the disordered cubic phase || (101) planes of the ordered tetragonal phase. Calculation shows these planes to have the best fit of any pair. It is considered that there is a good case for assuming the general existence of two-phase regions in alloy systems subject to order–disorder changes.—R. W. C.

***The Cohesion of Alloys: I.—Intermetallic Systems Formed by Copper, Silver, and Gold, and Deviations from Vegard's Law.** M. A. Jaswon, W. G. Henry, and G. V. Raynor (*Proc. Phys. Soc.*, 1951, [B], **64**, (3), 177–189).—Objections are put forward against seeking to extend the Wigner–Seitz treatment of metallic cohesion to alloys. A semi-empirical approach is proposed, applicable to dil. solid soln. The change in cohesive energy due to alloying is computed by considering polyhedra of 13 atoms (for the close-packed cubic lattice)—a central atom and its 12 nearest neighbours; other neighbours are neglected. The size of such a polyhedron in a solid soln. will differ from the appropriately weighted mean of the sizes in the pure constituent metals, and the deviation (which is proportional also to the deviation from Vegard's law) is directly related to the cohesive energy. The relation is derived, starting from a treatment by Fürth (*Proc. Roy. Soc.*, 1944, [A], **183**, 87; *Met. Abs.*, 1944, **11**, 394). The assumption has to be made that the variation with vol. of the energies of the components of an alloy is the same as in pure metals. The size of polyhedron which will minimize the energy of the alloy is computed, and from this the expected deviation from Vegard's law is obtained. With this deviation serving as criterion, the predictions of the theory are tested on binary mixtures of Cu, Ag, and Au. The sign of the deviation is correctly predicted for each of the dil. solid soln., but quant. agreement is good only for the Cu–Ag terminal soln. There are satisfactory reasons for the less good agreement in the other cases. The appn. of the new ideas to the problem of the *extent* of solid solubility is briefly indicated. 14 ref.—R. W. C.

***The Thermal Conductivity of a Copper–Nickel Alloy at Low Temperatures.** J. K. Hulm (*Proc. Phys. Soc.*, 1951, [B], **64**, (3), 207–211).—The thermal conductivity of an 80 : 20 cupro-nickel has been measured at temp. from 1·89 to 21·9° K. The conductivity falls rapidly at low temp. Following Makinson's theory, the conductivity is split up into electronic and lattice terms. Of the several forms of heat-wave scattering contributing to the second term (which is $\propto T^2$), it is concluded that only electronic scattering is of importance in the temp. range studied. 12 ref.—R. W. C.

***Effect of Ternary Additions on the Age-Hardening of a Copper–Silver Alloy.** Harold Margolin and Walter R. Hibbard, Jr. (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (2), 174–180).—Ten ternary alloys contg. Cu 96, Ag 3, and 1 at.-% of a third element (Al, As, In, Mg, Mn, Ni, P, Sb, Si, or Zn) were prepared, and after cold working, soln.-treatment, and quenching, a study was made—by means of X-ray, microscopic, and hardness measurements—of their behaviour on ageing at 350°–550° C. for periods up to 120 hr.; a binary alloy contg. Cu 97 and Ag 3 at.-%, similarly treated, was used as a comparison standard. The pptn. phenomena which occurred were found to include: (1) segregation of Ag (and third element where involved) beyond equilibrium solubility limits in the case of Cu–Ag, and probably in the ternary alloys as well, to form a transition Cu-base structure rich in Ag; (2) limited coherent growth of the precipitate, involving coherency possibly between precipitate and transition Cu lattice and coherency between transition Cu and the matrix Cu lattice poorer in Ag; and (3) breaking away of the precipitate, recrystn., and coarsening. Analysis of the hardness data revealed: (*a*) no correlation with grain-size or temp. of ageing; (*b*) max. hardness increment decreases in the order: P and Mg, Mn, Ni, In, Al, Si, As, Sb, and Zn; and (*c*) the elements, arranged in decreasing order of their influence on the rate of hardening are Sb, As, In, Mg, Si, Zn, Al or Mn, and Ni. Precipitate in Widmanstätten form was found, in all alloys, along intragranular boundaries, and in recrystallized nodules before attainment of max. hardness. The effect of the ternary element was related to differences between transition and precipitate lattice parameters and analogies with soln. and strain-hardening. 22 ref.—E. N.

***Thermoelastic Analysis of the Transformations of Bronzes.** Robert Cabarat, Pierre Gence, Léon Guillet, and Albert Portevin (*Compt. rend.*, 1951, **232**, (11), 1041–1042).—Using the apparatus previously described by C. (*ibid.*, 1943, **217**, 229) for the detn. of the modulus of elasticity and the logarithmic decrement of alloys and their variation with the temp., the authors studied the transformations in the solid state that occur in the Cu–Sn alloys contg. 20–32% Sn. The 24·9% Sn alloy (very near the eutectoid), cooled at the rate of 1° C./min., has two phases: α (f.c.c. with 7% Sn) and δ (large cubic lattice with 32% Sn). When heated, it undergoes an isothermal transformation, and above 520° C. a b.c.c. solid soln. is formed. The complete disappearance of the δ-phase at 20° C. involves a considerable diminution in the modulus of elasticity. Analogous phenomena occur in the case of the hypoeutectoid alloy contg. 20·2% Sn; in the annealed stage this alloy contains the same phases as the above alloy, but at 20° C. the α-phase is in excess. The variations of the properties near the temp. of transformation are less marked, since above 520° C. this alloy still contains an appreciable amount (38%) of the α-phase. An alloy contg. 32% Sn corresponds to the δ-phase ($Cu_{31}Sn_8$) and decomposes at 590° C. to a mixture of two phases whose lattices are hexagonal (contg. 26 atoms) and b.c.c. The variations of the properties are analogous to those of the preceding alloys.—J. H. W.

***Effect of Superheating and Casting Temperatures on Physical Properties and Soldification Characteristics of Tin Bronzes.** (Ames and Kahn). See col. 59.

***Studies of Cast Tin and Lead–Tin Bronzes, with Regard to the Effects of Nickel Additions.** A. H. F. Goederitz (*Metall*, 1950, **4**, (23/24), 495–498).—A report of an uncompleted study of the effects of adding Ni to sand- and chill-cast 86 : 14 Cu–Sn, and Cu–Pb (12–14%)–Sn (6–9%) bearing metals. In the case of the binary alloy, increasing addn. of Ni (up to 4%)—thereby replacing some of each of the basic constituents—cause corresponding improvements in the mech. properties, probably as a result of grain refinement. This beneficial effect is much smaller when similar addn. are made to the ternary alloys. The properties of all the materials are compared with those of Cu–Mn (7%)–Si (3%) bearing metals. It is concluded that German specifications for Cu–Sn and Cu–Pb–Sn bearing metals are such that, while still remaining within the permissible limits of compn., small addn. of Ni and Mn, and smaller addn. (~0·5%) of Sb, Al, Mg, and Si, can be made with consequent improvements in properties.—E. N.

***The Elastic Parameters of Beta-Brass.** Robert A. Artman and Donald O. Thompson (*J. Appl. Physics*, 1951, **22**, (3), 358).—A letter. Using several single crystals of brass contg. 55·12 at.-% Cu, Young's modulus was determined by a dynamic method and the rigidity modulus by a static one. The elastic parameters computed from the moduli are given, and differ slightly from other workers' values. The differences are attributed to small differences in compn., and some phys. conclusions are drawn from them.—R. W. C.

***Optimum Aluminium Additions in Commercial Yellow Brass.** R. A. Colton and R. H. Gilbert (*Trans. Amer. Found. Soc.*, 1950, **58**, 423–427; discussion, 427–428).—Cf. *Amer. Foundryman*, 1950, **17**, (4), 135; *Met. Abs.*, 1949–50, **17**, 806. —V. K.

***Melt Quality and Fracture Characteristics of 85 : 5 : 5 : 5 Red Brass.** (Baker, Upthegrove, and Rote). See col. 59.

***Remarks on the Subject of the Behaviour of Cuprous Metals Submitted to a Permanent Mechanical Stress.** J. Hérenguel and M. Scheidecker (*Rev. Mét.*, 1951, **48**, (3), 173–181; discussion, 181).—H. and S. undertook an investigation into the phenomena of intergranular failure of brasses under pure mech. stress without corrosion, determining the exact method of rupture and the temp. and time of appn. This included a study of the influence of internal stress superposed on the external forces and the evolution of cracks during the subsequent met. operations. The alloy principally used was the 72 : 28 Cu–Zn alloy. They found that the binary Cu–Zn alloys are subject to considerable intergranular cracking when under continuous stress in the temp. interval from 150° C. to the m.p. The extent of the crack is related to the stress, the temp., the time, and the grain-size. The internal tension that may exist in an alloy so stressed is not essential for bringing about failure; this is produced in annealed alloys or at temp. above the recrystn. temp. for cold-worked specimens. In the absence of all internal stress, failure does not appear to occur. In subsequent recrystn. the network of cracks becomes intragranular; on heating the crack breaks up into discontinuous elements; there is some kind of consolidation of the crack so that it is no longer susceptible to opening by tension; the crack becomes coated with oxide by heating in air. The phenomenon occurs even for small Zn contents, increases with the content to a max. at ~10% Zn, and remains virtually const. up to the limit of the α-solid soln. The mechanism of this cracking is still not clear, but it appears to be connected with dezincification.—J. H. W.

***Plastic Deformation and Diffusionless Phase Changes in Metals : The Gold–Cadmium Beta Phase.** L. C. Chang and T. A. Read (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 47–52).—A report of X-ray, elect.-resistance, and motion-picture studies of the diffusionless phase change in crystals of the β Au–Cd alloy. In the alloy contg. 47·5 at.-% Cd the high-temp. phase, β_1, has an ordered CsCl structure, 2 atoms/unit cell, and $a = 3{\cdot}3165 \pm 0{\cdot}0005$ kX.; the low-temp. phase, β′, is orthorhombic, $a = 3{\cdot}1476$, $b = 4{\cdot}7549$, $c = 4{\cdot}8546 \pm 0{\cdot}0005$ kX, with 2 Cd atoms at 0, 0, 0; 0, $\frac{1}{2}$, $\frac{5}{8}$, and 2 Au atoms at $\frac{1}{2}$, 0, $\frac{1}{2}$; $\frac{1}{2}$, $\frac{1}{2}$, $\frac{1}{8}$. Transformation takes place at ~60° and ~80° C. on cooling and heating, respectively. The alloy contg. 49·0 at.-% Au undergoes transformation from an ordered b.c.c. to a tetragonal structure at ~36° C. on cooling; and a reverse transformation on heating to 40°–55° C., depending on the previous history of the alloy. The orientation relationships in the lower-Cd alloy were found to be $(011)_{\beta_1}$ ‖ to $(001)_{\beta'}$, and $[\bar{1}\bar{1}1]_{\beta_1}$ ‖ to $[\bar{1}\bar{1}0]_{\beta'}$. Although, normally, multiple-interface transformation occurs, single-interface transformation can be obtained under carefully controlled conditions, and in the latter case, with both alloys, when a single crystal of β_1 transforms by multiple-interface mechanism into a crystal aggregate of β′, a single crystal of the initial orientation is invariably obtained when the reverse transformation occurs. Experimental evaluation shows that the crystallographic mechanism involved in the β_1–β′ transformation is a simple homogeneous shear $(\bar{3}31)_{\beta_1}$ $[323]_{\beta'}$ of ~3° plus a possible homogeneous contraction of ~0·015 kX units along the *b* axis of the orthorhombic cell. The nucleation (and therefore the transformation temp.) of the β′ phase on cooling is primarily dependent on the existence of imperfections within the specimen which have their origin in growth from the melt or in previous plastic deformation. The elect. resistivity of the low-Cd alloy decreases on transformation to the orthorhombic phase, while that of the higher-Cd alloy increases on transformation to the tetragonal phase. The orthorhombic phase has a characteristic rubber-like behaviour, which is not found in the tetragonal phase. 13 ref.—E. N.

***Coefficients of Thermal Expansion of Gold–Cadmium Alloys Containing 47·5 at.-% Cadmium.** Lo-Ching Chang (*J. Appl. Physics*, 1951, **22**, (4), 525–526).—A note. An oscillation technique for single crystals is described which gives precision values for the lattice spacing of any plane with a large Bragg angle. The method was applied to measuring the expansion coeff. of both forms of the substance (change-point ~60°–80° C.). The low-temp. orthorhombic form proved to be extremely anisotropic, with two negative coeff.—R. W. C.

***Magnetic Saturation Intensity [of Cobalt] and Some Other Related Measurements [in Iron–Nickel–Aluminium Alloys].** (Sucksmith). See col. 3.

Notes on Magnesium Alloys.—I.–II. F. A. Fox (*Engineer*, 1951, **191**, (4958), 148–150; (4959), 180–182).—A general article, outlining the manufacture, properties, and uses of Mg-base alloys. The various processes for the extraction of Mg are briefly described, with an account of the properties of the commercial alloys in the cast, wrought, extruded, and forged conditions. The effect of particular alloying elements —Zr, Al, Mn, Ce, &c.—on the phys. and mech. properties of the alloys are discussed and the tensile properties of the common Elektron alloys are tabulated. Heat-treatment of Al- and Zn-contg. alloys is discussed and the fatigue properties as indicated by the Wöhler and Haigh tests and their relation to static properties, are dealt with in some detail. F. concludes with sections on surface protection and indust. appn. of Mg alloys.—D. K. W.

***Influence of Structural Inhomogeneities on the Hot-Working Properties of Magnesium–Aluminium Alloys.** W. Lott (*Metall*, 1950, **4**, (21/22), 458–462).—A study is reported on the effects of the Mn content on the workability (drop forging and rolling) of fully homogenized ingots of Mg-Al 6 and Mg-Al 7—binary alloys of Mg with 5·5–6·5 and 6·5–8% Al, resp. The specified Mn content is 0·05–0·4%, and this is added to the melt in the form of $MnCl_2$, which decomposes in contact with the molten Mg to form : (1) Mn, which improves the corrosion-resistance of the alloy, and (2) gaseous Cl, which scours the oxide impurities in the melt, so that they form a surface slag which can be skimmed off. It was found that although other workers have reported solubilities of 0·15–0·3% Mn at the m.p. of the 7% Al alloy, the solidified metal contains <0·15% in solid soln., any excess being precipitated as Al_6Mn, with the consequent formation of forging cracks. Not only must some Mn be present for the aforesaid reasons, but some must also be present to prevent coarse crystal formation during the prolonged homogenization-treatment—20–30 hr. at 390° C.—which occurs in Mn-free alloys. It is concluded, therefore, that the specified limits for Mn should be 0·1–0·3%, with an optimum figure of 0·1%; under these conditions degrees of reduction of up to 85% can be easily obtained during the prodn. of semi-finished articles by rolling and drop forging. 11 ref.—E. N.

***The Ageing of Sand-Cast Mg–Al–Zn Alloys.** T. E. Leontis and C. E. Nelson (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (2), 120–124).—A comprehensive study is reported of the changes in tensile properties, microstructure, and dimensional stability, as a function of the time and temp. of ageing, after soln.-treatment, of the two Mg-base alloys : (1) AZ92A, contg. Al 9, Zn 2, and Mn 0·2%, and (2) AZ63A, contg. Al 6, Zn 3, and Mn 0·2%. The results show that: (*a*) a wide variation is properties can be obtained by selection of ageing temp. and time, and (*b*) while both alloys exhibit a measurable amount of growth on ageing, suitable ageing treatments render them dimensionally stable to subsequent exposure at temp. prevailing in engine applications. 14 ref.—E. N.

***Developments of High-Strength Magnesium Casting Alloy ZK61.** J. W. Meier and M. W. Martinson (*Trans. Amer. Found. Soc.*, 1950, 58, 742–750; discussion, 750–751).—Mech. properties of both cast and of heat-treated test-bars of Mg–Zn–Zr alloys were studied over the range of 1–11% Zn and 0·1–0·9% Zr. Optimum properties were obtained with a 6% Zn–0·7% Zr alloy. The alloy is soln.-treated between 425° and 500° C. for 2 hr. and aged at 150° C. for 48 hr. 29 ref.—V. K.

***A Study of the Grain Refinement and Protection Against Inflammability of Magnesium–Aluminium Alloys.** Albert Keil and Pierre Bertin (*Recherche aéronaut.*, 1951, (20), 3–9).—A suggested classification of the methods of grain refinement of Mg and its alloys is formulated depending upon the compn. of the nucleus of crystn. Experimental results are included to show that the casting conditions have a marked effect on the grain-size and consequent mech. properties of Mg–Al alloys. Protection from inflammability is effected by the addn. of ~0·01% Be, which does not induce increased grain-size. The value of Zn for grain refinement is not affected by the Be and the Mg–4% Al–1% Zn–0·01% Be alloy is particularly recommended for good mech. properties combined with protection from inflammability up to 780° C.—H. A. H.

***Transformation Temperatures of the Definite Compounds MnAs, [MnSb], and MnBi, and Their Relation to a Probable Mechanism of Antiferromagnetism.** Charles Guillaud (*J. Phys. Radium*, 1951, 12, (3), 223–226; discussion, 226–227).—The crystallographic and thermomagnetic properties of the compounds MnAs, MnSb, and MnBi are discussed, and the thermomagnetic properties of MnAs and MnBi interpreted in terms of a temp. of transformation from ferromagnetism to antiferromagnetism, rather than in terms of a Curie temp. The Mn ions form parallel layers with anti-parallel magnetization between adjacent layers. The resp. ionic saturation magnetic moments are (in Bohr magnetons): n_{MnBi}, 3·52; n_{MnSb}, 3·53; and n_{MnAs}, 3·40. 8 ref.—J. S. G. T.

***Relations Between Structural State and Magnetic Properties of Iron–Nickel Alloys of Approximate Composition Ni_3Fe.** Émile Josso (*J. Phys. Radium*, 1951, 12, (3), 399–401; discussion, 401).—The changes of sp. vol. associated with the order $\rightleftarrows$ disorder transformation of alloys of compn. approximating to that of Ni_3Fe (Permalloy) are utilized to determine the equilibrium diagram of the transformation. Changes of compn. of the alloys were effected either by varying the Ni content of the alloy or by adding other constituents. The establishment of order in Ni_3Fe is accompanied by a rise of Curie point, and an increase of the longitudinal magnetostriction coeff. Changes of the Curie point produced by variation in the degree of order can be accurately determined by the addn. of a small quantity of Mo to the alloy Ni_3Fe. The magnetostriction coeff. of Permalloy in the ordered state is twice as great as that in the disordered state of the alloy.
—J. S. G. T.

***Viscosity of the Sodium–Potassium System.** Curtis T. Ewing, Joseph A. Grand, and R. R. Miller (*J. Amer. Chem. Soc.*, 1951, 73, (3), 1168–1171).—Tabulated data from viscosity detn. made by a closed-viscometer technique on Na, K, and three Na–K alloy compn. from 60° (or the m.p.) to 200° C. show that the values for the alloys vary uniformly between those for the metals. The mixture equation $\phi = x_{\mathrm{Na}}\phi^{\circ}_{\mathrm{Na}} + x_{\mathrm{K}}\phi^{\circ}_{\mathrm{K}}$, where ϕ is the fluidity and x is the weight fraction, was found adequately to satisfy the data. The temp. variation of viscosity for each of the five metals or alloys can be effectively expressed by Andrade's equation $\eta v^{\frac{1}{3}} = Ae^{C/vT}$ (*Phil. Mag.*, 1934, [vii], 17, 698) where η is the viscosity, v the sp. vol., T the abs. temp., and A and C are const. for any given liquid.—J. R.

***The Tungsten–Aluminium Equilibrium Diagram.** O. Kubaschewski (*Z. Metallkunde*, 1950, 41, (5), 159–160).—An annotated equilibrium diagram, embodying the results of recent investigations.—E. N.

***The Uranium–Aluminium Equilibrium Diagram.** E. Scheil (*Z. Metallkunde*, 1950, 41, (5), 159–160).—An annotated equilibrium diagram, embodying the results of recent investigations.—E. N.

***Ten Zirconium Alloys Evaluated.—I.–II.** F. B. Litton (*Iron Age*, 1951, 167, (14), 95–99; (15), 112–114).—L. tabulates and discusses the mech. and oxidation resistance of binary Zr alloys contg. Hf, Ti, Al, Ta, Nb, W, Mo, Cu, O, or N. Hf up to 8·2% does not affect the tensile properties; Al, Ti, W, Nb, O, and N increase the U.T.S. Alloys contg. <1% W or Mo tested in the condition as-rolled at 400° C. have strengths comparable with those of alloys with 4–50% alloy addn. N and O both improve the ductility. The oxidation resistance of Zr is not improved by alloying with small amounts of Al (~0·05%) or N (<0·013%), and is slightly increased by O. Arc-melted metal has better resistance to O than induction- or graphite-crucible-melted Zr. In general, Zr alloys are not as resistant to oxidation as the commercially pure metal. A metallographic investigation showed that Zr alloy systems are predominantly eutectoidal.—J. H. W.

***Infra-Red Photoconductivity of Certain Valence Intermetallic Compounds.** J. G. N. Braithwaite (*Proc. Phys. Soc.*, 1951, [B], 64, (3), 274–275).—A letter. The prepn. and photoelectric threshold and peak wave-lengths of a number of tellurides, and of Zn_3As_2, SnS, and Sb_2Se_3, are reported.
—R. W. C.

***The Cohesion of Alloys. II.—Some General Metallurgical Implications.** W. G. Henry, M. A. Jaswon, and G. V. Raynor (*Proc. Phys. Soc.*, 1951, [B], 64, (3), 190–194).—Cf. *ibid.*, p. 177; *Met. Abs.* this vol., col. 17. Certain general conclusions regarding solubility in binary solid soln. are drawn from the theory described previously. It is an empirical fact that for a given amount of either expansion or contraction of a solute atom, the change in binding energy is always greater for contraction. It follows from this that of the two terminal solid soln. in a binary system, the one which has the larger ions as solvent should have the wider compn. range. The experimental correlation between compressibility and solubility range is also accounted for by the theory. An older view, on which the polarizing power and polarizabilities of the metal ions (as derived from experiments with salts contg. these ions) is an important factor in determining solubilities, is shown to require revision. Such effects are unimportant for large ions with a low charge (such as the metals of group IA), though they are probably significant in the case of ions such as Al^{3+}.—R. W. C.

***The Cohesion of Alloys. III.—Extension to High Solute Concentrations, and Application to the Alloys of Group IA Metals.** W. G. Henry and G. V. Raynor (*Proc. Phys. Soc.*, 1951, [B], 64, (3), 195–206).—Cf. *ibid.*, p. 190; preceding abstract. The theory described in the preceding parts is here extended to more conc. solid soln. The size deviation between the 13-atom polyhedron (v.s.) in the actual solid soln. and that computed as a weighted mean of those in the constituents, is associated with an energy increase Δ. This is assumed independent of concentration for dil. alloys, since the 12 atoms surrounding the central atom of the polyhedron are generally all of one kind. This is no longer true for conc. soln., and Δ becomes a function of concentration, assumed linear. The gradient of the relation is estimated by considering the 50 at.-% alloy in turn as a soln. of A in B, and of B in A. It is shown that if Δ is large, limited or vanishing solid solubility is expected; and if small, extensive or complete solid solubility is predicted. Calculations in this connection for all the possible binary systems of Na, K, Cs, and Rb, are in agreement with the theory. Where extensive solubility is predicted on one side of the diagram only, there may nevertheless be complete miscibility, but with a steep dip in the solidus (as is the case for Na–Cs). The theory is further extended to deal with the form of the liquidus and the freezing range, and the position of the eutectic point.

In all these connections, the qualitative predictions of the theory are confirmed by experiment.—R. W. C.

***Applications of Reaction Kinetics to Metallographic Problems [in Alloys].** George-Maria Schwab and George Petroutsos (*J. Phys. Colloid Chem.*, 1950, **54**, (4), 581–597).—S. and P. have obtained evidence as to the degree of electron saturation of the first Brillouin zone in alloys other than those of the Hume-Rothery type. The activation energy in the dehydrogenation of formic acid vapour was measured for various catalysts of different compn. in 6 alloy systems. It was found that in the system Au–Fe there is a deviation from the simple Hume-Rothery rules, in agreement with elect. and magnetic data. In the system Fe–C the hardness of martensite can be traced to its increased electron concentration. In the system Au–Sb, the compound $AuSb_2$ of pyrite type has a relatively empty zone, and in the system Cu–Sb the compound Cu_2Sb behaves similarly. In the system Cu–Mg, the compound Cu_2Mg corresponds to a saturated α phase, while the compound $CuMg_2$ has an extremely low degree of zone saturation. In the system Au–Pb the compound Au_2Pb, isomorphous to Cu_2Mg, corresponds to it also in its electron distribution.—D. M. D.

†Recent Progress in the Field of Permanent Magnets. K. Hoselitz (*J. Phys. Radium*, 1951, **12**, (3), 448–457; discussion, 457–458).—[In English]. The compn., magnetic, phys. and mech. properties, and heat-treatment of 13 magnet steels, the magnetic properties of 7 pptn.-hardening, 10 isotropic, and 8 anisotropic diffusion-hardening alloys, and of 7 permanent-magnet materials made from powders, are tabulated and discussed in considerable detail. The coercive forces of permanent-magnet alloys used in practice are of the order 500–1000 Oe., and the heterogeneity produced by pptn. or diffusion can, to some extent, account for observed values of the coercive force, although the detailed mechanisms are not known. Some alloys of Pt and some of Mn possess extremely high coercive forces; their cost or difficulty of prepn. limit their practical appn., but they are of considerable sci. interest. 41 ref.—J. S. G. T.

The Electrical Resistance at Low Temperatures of Some Ferromagnetic Metals [and Alloys] in a Magnetic Field. (Gorter). See col. 10.

***Influence of Order on Magnetic Properties [of Alloys].** R. Smoluchowski (*J. Phys. Radium*, 1951, **12**, (3), 389–397; discussion, 397–398).—[In English]. A new theory of magnetic saturation in binary alloys, based upon fluctuations of electronic concentration in all equivalent groups of atoms, is developed. In the case of the b.c.c. lattice these groups contain the first and second neighbours, and the theory agrees with experimental data relating to Fe–Co alloys. In the case of f.c.c. lattices, e.g. Fe–Ni alloys, the group contains nearest neighbours only. The theory enables one to calculate the effect of order upon such magnetic properties as saturation moment and magnetostriction, and theoretical and experimental results agree. The effects of order upon Curie temp., magnetic anisotropy, coercive force, and magnetic permeability and the effect of magnetic properties upon order phenomena are also discussed. 22 ref.—J. S. G. T.

***Atomic [Magnetic] Moments and Curie Points of Face-Centred Cubic Isoelectronic Alloys of the Iron Group.** P. Taglang (*J. Phys. Radium*, 1951, **12**, (3), 414–417; discussion, 417).—A study of the magnetic properties of 4 isoelectronic Fe–Co–Ni alloys, having a f.c.c. lattice, reveals the existence of a linear relation between the Curie temp. and abs. atomic magnetic moment. Two components of the ferromagnetic moment are distinguished, viz. (1) an extrapolated moment, independent of the Curie temp. but variable to the extent of 1 Bohr magneton/electron in the electron class, and (2) a supplementary moment, dependent upon the Curie temp. This latter result conforms with Néel's theory of dynamic ionic equilibrium ("Le Magnétisme," 1940, vol. II, p. 105).
—J. S. G. T.

***Ferromagnetism, Paramagnetism, and Cohesive Energy of Transition Metals and Their Alloys.** T. G. Owe Berg (*J. Phys. Radium*, 1951, **12**, (3), 418–429).—[In English]. A theory which extends the customary representation of the magnetic moment of binary alloys in terms of the resp. atomic numbers and electronic *d* is extended to include intermetallic compounds. The formulæ derived are applied, *inter alia*, to calculate the magnetic susceptibility of Ni, and the calculated results agree with experimental values obtained by Fallot (*Ann. Physique*, 1936, [xi], **6**, 305; *Met. Abs.*, 1938, **5**, 260). The formulæ, applied to binary Fe alloys, differentiate between 3 types of alloys, viz. perfect solid soln., a "metastable" phase, and an intermetallic-compound phase. 17 ref.—J. S. G. T.

***New Techniques and Results in the Measurement of Magnetostriction [of Alloys].** J. E. Goldman (*J. Phys. Radium*, 1951, **12**, (3), 471–475; discussion, 475).—[In English]. Apparatus and technique using elect. strain gauges for the measurement of linear and vol. magnetostriction are described. Results are given for magnetostriction in oriented 3·5% Si steel (Hipersil), and in a very small single crystal of an Fe–30% Co alloy in the form of an oblate spheroid of dimensions 3·5 mm. in dia. and 0·3 mm. thick. Values of the magnetostriction const. in the Fe–Co alloy are: $\lambda_{100} = 1{\cdot}5 \times 10^{-5}$; $\lambda_{111} = 8{\cdot}5 \times 10^{-5}$. G. suggests that the mechanism of nucleation in Alnico V consists in the pptn. of a phase rich in Fe and Co contg. approx. 30% Co. Nuclei with axes making a small angle with the matrix grow preferentially, and magnetic anisotropy of the alloy results from cryst. anisotropy of the precipitate. The relation between magnetostriction and order–disorder transformation is briefly discussed with ref. to Fe–Co alloy and to Ni_3Fe (Permalloy). 15 ref.
—J. S. G. T.

***Ferromagnetism in Weak [Alternating] Magnetic Fields, and the Curie Point.** (Goldschmidt). See col. 14.

The Gyromagnetic (Magneto-Mechanical) Ratios of Some Ferromagnetics of the Iron Group. (Meyer). See col. 13.

Rotations [of Magnetic Moments] During Magnetization. (Initial Magnetization, Coercive Field). (Guillaud). See col. 13.

†High Negative Molecular Fields, and the Different Mechanisms of Passage from Antiferromagnetism to Weiss Paramagnetism. (Foëx). See col. 13.

***A New Conception of the Electronic State of Ferromagnetics: Introduction of the Notion of Intermittent Activation.** (Forrer). See col. 12.

***†Collective Electron Ferromagnetism in Metals and Alloys.** (Stoner). See col. 12.

Gmelins Handbuch der anorganischen Chemie. Legierungen der Platinmetalle: Osmium, Iridium, Platin. (——). See col. 91.

3 — STRUCTURE

(Metallography; Macrography; Crystal Structure.)

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

***Chemical Polishing of Aluminium and Its Alloys.** J. Hérenguel and R. Segond (*Rev. Mét.*, 1951, **48**, (4), 262–266).—Electrolytic polishing is used in most laboratories for the polishing of Al and its alloys. With the baths usually employed, i.e. H_2SO_4 and H_3PO_4 oxidized with either HNO_3 or H_2CrO_4, high c.d. is required which in large specimens results in large currents, thereby introducing practical difficulties. Chem. polishing has, therefore, been adopted in some laboratories for light alloys, using similar baths to the above. H. and S. investigated the possible appn. of this process on an indust. scale to the chem. polishing of Brillalumag 3 (Mg 3%), Alumag 50 (Mg 5, Mn 0·4%), and 99·5% Al. They found that it is possible to use two distinct methods of chem. attack, both giving a good polish: (1) a turbulent region with an abundant evolution of gas, and (2) an attack forming at the surface of the specimen a viscous film rich in Al salts, analogous to that observed in electrolytic polishing. They defined the range of compn. and temp. for the appn. of each method. A generally applicable bath consists of H_3PO_4 70, H_2SO_4 25, HNO_3 5% by vol., the attack being carried on for from 30 to 120 sec. This reveals: (1) furrowing, which indicates local variations of compn., (2) etching, whose significance has yet to be interpreted, and (3) insoluble phases, shown, not as planes but as 3-dimensional vol.—J. H. W.

***Modifications of the Structure of Aluminium During Creep.** G. Wyon and Ch. Crussard (*Rev. Mét.*, 1951, **48**, (2), 121–130).—W. and C. briefly review the current explanations for the fragmentation of the grains of a metal by creep, which are as important as they are debatable, and describe their own experiments with 99·99% Al in the form of plates 0·8–1·0 mm. thick, heat-treated, electrolytically polished, and attacked in certain cases with a mixture of HCl, HNO_3, and HF. Creep, usually at 310° C. but also up to 600° C., was recorded on a Chevenard machine. After testing, the specimens were examined at different magnifications, and then repolished, attacked by the 3-acid mixture, and examined for corrosion pits. Some two dozen photomicrographs are reproduced. Creep at very high temp. (towards 600° C.) introduces broken boundaries into the interior of the metal and these are very mobile at this temp. and constitute a reversible phenomenon. Once the specimen has been cooled to below the temp. of total solubility of the impurities present, these boundaries become very stable. With creep at 250°–300° C. micrography reveals: thick slip bands; waves in the surface forming "cellules" corresponding to successive bending; large angular kinking probably also arising from progressive bending; a net-work of swellings in the smaller grains or near the boundaries of the larger ones, arising from the boundaries of polygonization. The first types of deformation occur at the lower temp., the latter at higher temp.; the "cellules" are the more frequent the more imperfect the initial grain. A number of modifications of the intergranular boundaries were observed and are recorded. There is a correlation between the modification of the micrographic structure, the appearance of X-ray asterism, and the form of the creep curve, and the observations are amply explained by the theory of dislocations: following the distribution of the dislocations formed during creep progressive bending or well-defined kinking is observed, which depends on the obstacles that the dislocations encounter. In the conditions of attack employed, large corrosion pits appeared only on groups of dislocations; isolated dislocations gave small pits. 14 ref.—J. H. W.

Interatomic Distances in Co_2Al_9. Linus Pauling (*Acta Cryst.*, 1951, **4**, (2), 138–140).—[In English]. A reply is made to the criticism by Douglas (*ibid.*, 1950, **3**, 19; *Met. Abs.* 1949–50, **17**, 571) that Pauling's theory of the intermetallic bond predicts a wrong value for the Co–Al bond in Co_2Al_9 crystals. P. states that his theory was incorrectly applied, and shows how a correct calculation gives 2·516 Å. (compared with the experimental value of $2{\cdot}47_0$ Å.). The Al–Al distance is also discussed. Arguments are advanced why Co atoms should transfer one electron to Al (in opposition to an earlier view proposing the reverse exchange); a calculation is presented on this assumption giving exact agreement between theoretical and experimental values of the Co–Al bond distance.—R. W. C.

***The Torsion Texture of Copper.** W. A. Backofen (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1950, **188**, (12), 1454–1459).—Specimens of O.F.H.C. Cu and of electrolytic T.P. Cu were twisted, in a torsion testing machine at const. rates of 120° and 215°/min., up to shear strains of 7·7, and their pole figures were then determined by the Geiger-counter X-ray spectrometer method. The results show that the torsion texture of both materials can be described by a compn. of the four ideal orientations: (112) $[11\bar{1}]$, $(11\bar{1})$ [112], (110) [001], and (112) $[1\bar{3}1]$. The crystallographic planes of these various orientations are tangential to the surface, and the directions are ‖ to the longitudinal axis of the torsion specimen, thus showing that the texture does not appear to be influenced by the principal normal strain history. Consideration of the possible slip directions associated with the ideal orientations (112) $[11\bar{1}]$, $(11\bar{1})$ [112], and (110) [001] suggests that torsional plastic flow occurs primarily by slip along the direction of max. shear stress, which is tangential to the surface of the specimen and ⊥ to its longitudinal axis. Microstructure observations made in conjunction with reverse torsion experiments are in agreement with this mode of flow. It is concluded that the results indicate that the relationship between strain and the development of textures is, however, by no means a simple one. 12 ref.—E. N.

***Interfacial Free Energy of Coherent Twin Boundaries in Copper.** R. L. Fullman (*J. Appl. Physics*, 1951, **22**, (4), 448–455).—A new explanation is given of the occurrence of dihedral angles >180° at the intersection of a boundary between a grain *A* and a grain *B*, *A* and the twin of *B*, and the coherent twin boundary of the twin (i.e. the (111) plane of symmetry). By measurements of such dihedral angles, the ratio: energy of coherent twin boundary to energy of grain boundary, the latter being assumed invariant, is found to be 0·045 ± 0·003. This ratio was also estimated by measuring, by means of an oblique sectioning technique, the dihedral angles of the two kinds of boundary at a Cu surface etched by Pb vapour. The result here was 0·026 ± 0·002. The reason for the systematic error being unknown, a simple mean value of 0·035 ± 0·006 was adopted. An appendix gives a full discussion of the geometry of oblique sections.
—R. W. C.

***Crystallography and Interfacial Free Energy of Non-Coherent Twin Boundaries in Copper.** R. L. Fullman (*J. Appl. Physics*, 1951, **22**, (4), 456–460).—Annealing twins in Cu sometimes do not go right across a grain. The boundaries terminating such a parallel-sided twin are termed non-coherent. These boundaries were found all to be close to planes of particular indices, viz. (113) of the grain (which nearly coincides with (335) of the twin), or vice versa. This information was found by using an accessory, which is described, allowing the inclination of a plane running through a thin sheet to be determined from a microscopic examination of the two surfaces of the sheet. The ratio: energy of non-coherent boundary to energy of grain boundary was determined as 0·80 ± 0·015. Two measurements by the Pb etch method (see preceding abstract) gave results close to this.—R. W. C.

***Long-Range Order in Beta-Brass and Cu_3Au.** D. T. Keating and B. E. Warren (*J. Appl. Physics*, 1951, **22**, (3), 286–290).—An investigation of long-range order was made by means of a Geiger-counter spectrometer, using single-crystal samples of β-brass and Cu_3Au. Experiments were made with the brass at different temp., and with the Cu_3Au either at temp. or in the quenched condition. In spite of a resolution which would have sufficed to detect the phenomenon, no splitting of high-angle reflections was found for partly disordered samples of either kind. There was a *continuous* transition in the line positions from those characteristic of the ordered to those characteristic of the disordered alloy. From this it is concluded that below the critical temp., there is only one ordered phase of variable long-range order, and not a mixture of fully ordered and disordered domains. The long-range order parameter for Cu_3Au is found to be a single-valued function of the temp., from whichever side it is approached. If a two-phase region does exist, it must be very narrow. The parameter/temp. relation agrees well with recent theoretical predictions.—R. W. C.

***A Problem of Archæological Metallurgy: X-Ray Examination of an Egyptian Object in Electrum.** (Mme) A. R. Weill (*Rev. Mét.*, 1951, **48**, (2), 97–104; discussion, 104).—W. made an X-ray examination of an ancient Egyptian jewel from the Louvre Museum with the object of carrying out a non-destructive analysis. She discusses the general appn. of X-rays to the analysis of solid soln. according to Vegard's law (*Z. Physik*, 1921, **5**, 17; *J. Inst. Metals* (Abstracts), 1923, **29**, 679). This law is subject to exceptions in binary alloys, so that there would appear to be still greater difficulties in applying it to ternary alloys, such as Electrum. Ancient Egyptian objects have been found to have the compn. range: Au 67·0–80·1, Ag 20·3–25, Cu 0–8%. Au and Ag are miscible in all proportions and have similar parameters, but the parameters of the alloys contg. Cu decrease rapidly with the Cu content. The jewel in question was subjected to visual examination, which indicated some Cu; to a *d* detn., which, by means of an equal-*d* curve, fixed the Ag content if the Cu content were known; and to the measurement of the parameter by X-rays, which finally suggested the compn.: Au 75, Ag 22, and Cu 3%. 11 ref.—J. H. W.

***Crystal Structures and Transformations in Indium–Thallium Solid Solutions.** Lester Guttman (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1950, **188**, (12), 1472–1477).—A report of metallographic and X-ray diffraction studies of the In–Tl system between 0 and 75 at.-% Tl, the metals from which the alloys were prepared being of ~99·9% purity. The results show that: (1) a transformation f.c. tetragonal ⇌ f.c.c. occurs at ~105° C. at 18 at.-% Tl, and 25° C. at ~23 at.-% Tl, the tetragonal form being the one which is stable at low temp.; the transformation produces a distinctive lamellar structure (illustrated by photomicrographs), and the progress of the interface between the phases can be followed by the surface distortions produced; (2) the b.c.c. form found at room temp. is undoubtedly a terminal solid soln. derived from the high-temp. form of Tl; and (3) there is no evidence of ordered structures at any temp. A revised equilibrium diagram of the system is given, which is derived from the results of Valentiner (*Z. Metallkunde*, 1940, **32**, 244; *Met. Abs.*, 1941, **8**, 325) and the present work. 13 ref.—E. N.

***Crystallography of the Cubic–Tetragonal Transformation in the Indium–Thallium System.** J. S. Bowles, C. S. Barrett, and L. Guttman (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1950, **188**, (12), 1478–1485).—Cf. preceding abstract. The diffusionless transformation from the f.c.c. (*A*1) to the f.c. tetragonal (*A*6) structure in certain alloys of the In–Tl system was analysed by metallographic and X-ray measurements on single grains in polycryst. specimens contg. 20·75 at.-% Tl. The experimental results show that (1) the transformation produces lamellæ on (101) planes with sub-lamellæ also on selected (101) planes; (2) the lamellar structures and the tetragonality appear at the same temp.; (3) the transformation ceases if cooling is interrupted; (4) the lamellar structures are formed by the rapid (but observable) propagation of an interface || the main bands; each nucleus develops into a larger number of main bands tilted in opposite senses; (5) the first plates to form possess a finite degree of tetragonality, the axial ratio being 1·020; as the temp. is lowered the degree of tetragonality increases, and becomes 1·0356 at room temp.; (6) the range of temp. over which transformation occurs, and the hysteresis gap between the direct and inverse transformations, are quite small, ¾° and 4½° C., resp.; and (7) the tetragonal to cubic transformation apparently proceeds by a mechanism that is just the reverse of the direct transformation, since, except for occasional residual markings, the relief effects produced on cooling disappear on heating. A math. proof is given (in an appendix) which shows that the transformation cubic → tetragonal can be accomplished by two shears, and it is proposed, therefore, that the transformation occurs by a shear on the (101) plane in the $[\bar{1}01]$ direction, followed by a second shear on (011), $[0\bar{1}1]$. The main lamellæ result from oppositely directed shears of the first type, the sub-lamellæ from oppositely directed shears of the second type. Experimental observations of the orientation changes produced by the transformation are in good agreement with those postulated by this double-shear mechanism, but not with the frequently postulated twinning mechanism. 14 ref.—E. N.

***Formation and Behaviour of Sub-Boundaries in [3·3%] Silicon Iron Crystals.** C. G. Dunn and F. W. Daniels (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (2), 147–154).—A report of an X-ray and microscopic study of the structures formed in large single and two-grain crystals of commercial Fe–Si (3·3%) alloy which have been subjected to plastic deformation by cold rolling or by bending, and subsequently annealed. In single crystals the following was observed: (1) polygonization, which produces a relatively fine sub-grain structure, and (2) sub-grain growth of a fairly uniform kind like normal grain growth but selective in nature, removing the major spread in orientation and thereby producing a more nearly perfect single crystal of low sub-boundary energy. Annealing of two-grain specimens in which either one or both of the grains contained sub-grains disclosed that: (1) when one grain alone contains sub-boundaries, the other grain grows quite readily, apparently through interaction of the sub-boundaries with the ordinary boundary, and (2) when both grains contain sub-boundaries, the ordinary boundary sometimes fails to move, apparently from lack of driving energy; at the same time, however, the sub-grains grow larger; at other times the boundary does move, but in general the area swept is not free of sub-boundaries. The kinetics of the boundary and sub-boundary migration are discussed. 16 ref.—E. N.

***The Crystallographic Mechanism of the Martensite Reaction in Iron–Carbon Alloys.** J. S. Bowles (*Acta Cryst.*, 1951, **4**, (2), 162–171).—[In English]. By means of a closely reasoned analysis a crystallographic mechanism of the martensite reaction is derived which is fully consistent with: (1) the orientation relationship observed between austenite and martensite, (2) the martensite habit plane, and (3) the angle of surface tilt in a martensite needle and the behaviour of scratches traversing such a needle. The starting point is the theoretical treatment of the problem by Jaswon and Wheeler (*ibid.*, 1948, **1**, 216); in this work certain planes of the austenite lattice were shown to be unrotated in the transformation if the atomic displacements were to be kept as small as possible. Using these invariant planes, the index transformation matrix which defines the orientation relationship, and certain geometrical properties of a general type of homogeneous distortion (i.e. a distortion in which the direction of atom movement D does not lie in the undistorted plane P), B. derives by systematic trial and error a unique mechanism consisting of two successive distortions: (1) a homogeneous distortion (not a shear) in the austenite, with $P = (225)$ and $D = [\bar{1}\bar{1}2]$; (2) a shear, homogeneous only within minute regions of the plates formed by the first distortion, with $P = (112)$ referred to the martensite lattice, and $D = [1\bar{1}1]$. It is pointed out that both the distortions are in the resp. twinning directions of austenite and martensite. This analysis applies to the formation of cubic martensite; slight modifications, which are discussed, are required when tetragonal martensite is formed. The mechanism is found to satisfy the requirements (1)–(3) mentioned above, and certain peculiarities in connection with the multiplicity of the habit planes are also explained. The analysis here described applies strictly only to pure Fe and Fe–C alloys contg. $<1·4\%$ C; highly alloyed ferrites have slightly different orientation relationships and habit planes.—R. W. C.

***Metallographic Study of the Martensite Transformation in Lithium.** J. S. Bowles (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 44–46).—Single crystals of pure Li were produced by the Bridgman technique, and then given an excellent metallographic polish by gentle swabbing with cold methyl alcohol saturated with reaction product, Li alcoholate. The alcohol was then removed with dry benzene and, after the latter had been allowed to evaporate, the specimens were rapidly quenched in liq. N, when the spontaneous transformation to a martensitic structure—b.c.c. to h.c.p. with the hexagonal layers in imperfect stacking sequence—occurred with an audible click, i.e. at temp. above 77° K. As the relief effects produced did not disappear when the specimens were transferred into mineral oil at room temp., when, of course, the martensite phase was no longer present, orientations could be determined from back-reflection Laue patterns, and from these the habit plane was derived. This was found to be ~(441), and is closely similar to the habit planes utilized in other martensite transformations in b.c.c. metals. Although this (441) plane is not in agreement with Burgers' prediction (*Physica*, 1934, **1**, 561; *Met. Abs.*, 1934, 1, 580) of (112), the atom movements proposed by Burgers are probably correct, for they are consistent with the observed superlattice in Cu–Al martensite. To modify Burgers' mechanism so that it becomes consistent with the observed habit plane and relief effects, it will be necessary to find out how the " Burgers' unit distortion " is repeated throughout the lattice. 11 ref.—E. N.

***Crystallographic Angles for Magnesium, Zinc, and Cadmium.** Edward I. Salkovitz (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 64).—A note. The angles between the crystallographic planes in h.c.p. crystals of Mg, Zn, and Cd are tabulated. A typical (0001) standard projection of Zn ($c/a = 1{\cdot}856$) is given, which, for all practical purposes, serves for Cd ($c/a = 1{\cdot}886$) also. For Mg ($c/a = 1{\cdot}6235$), however, another net must be constructed.—E. N.

***Lattice Parameters of Magnesium Alloys.** Robert S. Busk (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1950, **188**, (12), 1460–1464).—Two groups of binary Mg alloys were prepared from high-purity elements; the first group consisted of the alloys of those elements—predominantly Group B—which are relatively soluble in Mg; the second group consisted of most of the remaining metallic elements—all Group A, transition or rare-earth elements—which are relatively insoluble in Mg. The lattice parameters of the alloys, as determined from X-ray back-reflection diffraction patterns of completely homogenized samples, together with additional data from the literature, are tabulated. These values show that: (1) the change in lattice parameter a_0 with addn. of solute $\propto$ the radius of the solute alone. Elements which cause a decrease in a_0 are Au, Mn, Rh, and Zr; all these have atomic radii smaller than that of Mg. Elements which increase are As, Ba, Ce, La, Ni, and Pd; Ba, Ce, and La all have atomic radii greater than that of Mg, and their effect is straightforward; As, Ni, and Pd, however, have atomic radii smaller than that of Mg, and the reason that these elements cause a decrease in a_0 values is not known; (2) the change in lattice parameter c_0 $\propto$ both the atomic radius and valence of the solute, the relative effect of each being such that the c/a ratio $\propto$ the electron concentration alone; elements decreasing c/a are Au, Ce, La, and Pd, and those increasing c/a are Ba, Ir, Pt, Rh, Ti, and Zr; (3) the solute elements Ca, Cu, Sb, Si, Te, and W show no change in either a_0 or c/a; of these Ca and Cu do have some solid solubility in Mg, but the change in parameter is apparently too small to be found; the other elements seem to have vanishingly small solid soly.; (4) the rare-earth elements Ce and La may act as electron receivers in Mg solid soln.; (5) the fact that enough Ti goes into solid soln. to raise the c/a ratio, coupled with its extremely low liq. solubility, probably means that the Mg-rich, Mg–Ti binary alloys, like the similar Mg–Zr alloys, constitute a peritectic system; (6) few, if any, solutes in Mg are fully ionized; (7) the effective atomic radius, and the relative effective valence shown by a solute in binary solid soln. are retained in ternary alloys. Empirical equations are derived from the data which allow the calculation of the lattice parameters of ternary solid soln. from binary data. 14 ref.—E. N.

Finishing Molybdenum. —— (*Product Finishing* (*Lond.*), 1951, **4**, (3), 62–63).—Notes on the cleaning, electropolishing, metallographic polishing, and etching of Mo are given.
—H. A. H.

***High-Temperature Structure and Thermal Expansion of Some Metals as Determined by X-Ray Diffraction Data. I.—Platinum, Tantalum, Niobium, and Molybdenum.** James W. Edwards, Rudolph Speiser, and Herrick L. Johnston (*J. Appl. Physics*, 1951, **22**, (4), 424–428).—Crystal structure and thermal expansion were studied for the four metals between 1100° and 2500° K. The coeff. of $(T - 291)$ and of $(T - 291)^2$ are given, and comparison with values calculated by the Grüneisen theory show that this gives fairly good extrapolation to high temp. No structure changes were observed, and a report in the literature postulating such a change in Ta between 2473° and 2773° K., on the basis of an observed permanent dilatation, is discounted as being due to mech. strain.—R. W. C.

***Grain Boundaries and Sintering [of Silver].** A. P. Greenough (*Nature*, 1950, **166**, (4230), 904–905).—Cold-rolled, high-purity Ag sheet contg. evenly distributed blowholes was annealed at 900° C. for 19 hr. in N. The blowholes were seen, in the microsections, to have accumulated in the planes of the grain boundaries and to have decreased in number but increased in size. The blowholes were swept up by the grain boundaries during the process of recrystn. and grain growth. The agglomeration of blowholes can be explained in terms of the free energy of the system. The polished surface of the metal showed grooves, sometimes associated with steps, after the first annealing process. This can occur by transference of blowholes from the outer layers of the metal to the surface, either bodily or as vacant lattice sites.
—B. R. T. F.

***X-Ray Measurement of Short-Range Order in Ag–Au [Alloys].** N. Norman and B. E. Warren (*J. Appl. Physics*, 1951, **22**, (4), 483–486).—Single crystals of AgAu and Ag_3Au were examined in monochromatic X-radiation on a Geiger-counter spectrometer, measurements being made at numerous points in the $(hk0)$ layer of reciprocal space. Corrections were made for diffuse scattering due to temp. and for other forms of extraneous scattering. The short-range order parameter for first and second nearest neighbours was obtained from the data by a Fourier summation method. The results for nearest neighbours (α_1) were: AgAu, $\alpha_1 = -0{\cdot}08$ $(-0{\cdot}33)$; Ag_3Au, $\alpha_1 = -0{\cdot}05$ $(-0{\cdot}33)$. The values in brackets would correspond to full order. The parameters, α_2, for second nearest neighbours were 0·01 in each case. The ratio of α_1 for AgAu and Ag_3Au is in agreement with a recent theoretical treatment (Cowley, *Phys. Rev.*, 1950, [ii], 77, 669; *Met. Abs.*, 1950–51, **18**, 180). The value for the critical temp. obtained from the data by applying this theory is 160° K., which accounts for the impossibility of establishing long-range order.—R. W. C.

***Accurate Determination of the Lattice of Beta-Titanium at 900° C.** Daniel S. Eppelsheimer and Robert R. Penman (*Nature*, 1950, **166**, (4231), 960).—The lattice const. of Ti at 900° ± 5° C. was determined using Cohen's method of least squares. A 19-cm. Unicam high-temp. camera was employed, a vacuum of 0·3 μ Hg being obtained in the vacuum chamber. The specimens (99·0+ % pure, the impurities being 0·3% C with small quantities of O, N, and Fe) were prepared by swaging followed by annealing for 1 hr. at 1000° C. The value of a_0 for the b.c.c. lattice at 900° ± 5° C. was determined as 3·3065 ± 0·00001 Å., and the corresponding atomic dia. 2·86 Å.—B. R. T. F.

The Structure of Electrodeposited Metals. (Hothersall). See col. 45.

*†**The Slip, Twinning, Cohesion, Growth, and Boundaries of Crystals.** H. Wilman (*Proc. Phys. Soc.*, 1951, [A], **64**, (4), 329–350).—The stability of two lattices, rotated relative to each other and in contact along a lattice plane, is discussed in terms of the *d* of sites on the two lattices which are in coincidence for particular angles of rotation. Numerous experimental observations of such relationships are then listed, and the angles of rotation shown to be associated with a high *d* of lattice sites in coincidence. The evidence is based on electron-diffraction photographs, made of crystals grown from soln., cleaved crystal flakes, single-crystal films floated off soluble substrates (e.g. Ag and Pd deposited on rock salt), beaten metal foils (Cu, Au, Ag, Pt, and Pd), and some of such foils after recrystn. It is considered that in most cases mentioned, there has been a phys. rotation of the lattices relative to each other, induced by stress. The occurrence of such rotation (called " rotational slip ") is proved conclusively by direct torsion applied to large cleavage fragments of $K_4Fe(CN)_6$, and of gypsum. It is considered that rotational slip plays an important role in connection with a number of met. phenomena, including deformation bands, secondary recrystn., and mech. twinning in layers of close-packed cubic metals deposited electrolytically or in epitaxy. It also plays a part in unidirectional abrasion. The relation of rotational slip to growth of crystals is fully discussed. 80 ref.
—R. W. C.

***Forces Exerted on Dislocations and the Stress Fields Produced by Them.** M. Peach and J. S. Koehler (*Phys. Rev.*, 1950, [ii], 80, (3), 436–439).—The force on an element of a Burgers dislocation in an external stress field is deduced, and an analogy is drawn between this expression and that for the force on a current element in a magnetic field. Math. expressions are then obtained for the behaviour of dislocations in a stress field and for the interaction of dislocations with one another.—P. C. L. P.

***The Widened Lattice Interval and Its Essential Role in the Behaviour of Crystalline Metal.** Donald P. Smith (*Science*, 1951, **113**, (2935), 348–352).—X-ray evidence of widened lattice intervals in work-hardened metal, as compared with the continuous and uniform lattice indicated for the fully annealed or dead-soft condition, is supplemented and discussed in the light of H solubilities. The widened intervals are evidently rifts in the crystal lattice, produced initially by plastic deformation, but capable of extension or healing under suitable conditions. 16 ref.—J. R.

" Oriented Growth " in Primary Recrystallization. Joseph J. Becker (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (2), 115).—A note. The extension of the " oriented growth " hypothesis of recrystn. textures to the primary recrystn. of single crystals and of polycryst. materials showing strong deformation textures is discussed. B. considers that if recrystn. textures are to be explained by selective growth from among nuclei of all orientations, it is difficult to see how the recrystn. texture : (1) can have any dependence whatever on the manner or the degree of deformation, and (2) can be sometimes very dependent on the annealing temp. 2 ref.—E. N.

***A Theory of Deformation Textures in Metals.** W. R. Hibbard, Jr. (*Rev. Mét.*, 1951, 48, (2), 131–134).—During severe cold working the orientation of most of the grains of a polycryst. metal approach a common orientation with respect to the axis of stress, which defines the deformation texture. This texture develops progressively ; but, once it is established, the grains can be subjected to further cold working without much change of orientation. H. puts forward a theory of deformation texture starting from considerations of the stable orientation of single crystals and of the slip and flow under stress particularly applied. This theory prognosticates the predominant textures found experimentally in f.c.c., b.c.c., and h.c.p. metals ; but it takes no account of the deviations from the ideal orientations that are found in polar figures. 22 ref.—J. H. W.

***Atomic Relationships in the Cubic Twinned State.** W. C. Ellis and R. G. Treuting (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, **191**, (1), 53–55).—It is shown that twinning in f.c.c., diamond, and b.c.c. metals can be the result of small atom displacements of a fraction of an interatomic distance on planes lying within a coincidence-site superlattice ; no extensive shear is required. In order to satisfy nearest-neighbour disturbance requirements across such boundaries, imperfections are required in a stable lateral twin boundary. The predisposition to twinning of materials having a diamond cubic structure is the result of this lattice having a double layer of coincidence sites along twin planes, and a favourable lateral (or incoherent) trans-boundary spacing. 7 ref. —E. N.

***A Metallographic Study of Diffusion Interfaces.** G. C. Kuczynski and B. H. Alexander (*J. Appl. Physics*, 1951, 22, (3), 344–349).—By sintering wires to flat metallic blocks, diffusion interfaces were obtained between the following pairs of metals : Cu–Ni, Au–Ag, Cu–Ag, Cu–Fe, Fe–Ni, Cu–Zn, Ni–Zn, and 70 : 30 brass–Cu. Typical photomicrographs are given of the sectioned interfaces. Grooves were observed in all cases, " digging " into the lower-m.p. component in the interface region. This is attributed principally to the faster self-diffusion in the lower-melting metal. The relevance of this (as yet imperfectly understood) phenomenon to practical powder metallurgy is touched upon.—R. W. C.

***The Mechanism of the Electrolytic Polishing of Metals.** W. J. McG. Tegart and R. G. Vines (*Rev. Mét.*, 1951, 48, (4), 245–250).—T. and V. carried out experiments on the electropolishing of Bi to study the parts played by the diffusion layer and the film of oxide and their importance in the process. The bath finally adopted was a saturated soln. of KI acidified with HCl at 7 V. and a c.d. of 20 amp./dm.^{2}. An adherent brown film formed on the surface of the metal, and it was found advantageous to polish for successive periods of 30 sec. each. The mechanism of electrolytic polishing was seen to be more complex than at first appeared. The viscous film round the specimens, chiefly caused by diffusion, is important to the extent that it contributes to the prodn. and the protection of a film of oxide adjacent to the metal. This oxide film in its turn protects the metal and ensures the uniform soln. of the surface. Although this theory was based on metals polished in relatively static conditions and at low c.d., it is also applicable to high c.d. where much gas is evolved. Chem. polishing is, perhaps, a manifestation of the same phenomenon of oxide formation and protection. 11 ref. —J. H. W.

***A Universal Electrolytic Polishing and " Rectification " [Controlled Dissolution] Solution.** Jaime Estruch Battle (*Galvano*, 1951, (169), 11–13).—Details are given of an electrolytic polishing soln. which is claimed to be applicable to ferrous metals, brasses, bronzes, light alloys, Cu, Pb, Zn, and Sn. This soln., which contains 25 g./l. H_2SO_4, 33 g./l. HF, 8·3 g./l. H_3BO_3, 372 g./l. H_2CrO_4, 12 g./l. HNO_3, 4·3 g./l. phthallic acid, and 328 g./l. H_3PO_4, is operated at 94° C. using either Cu or stainless-steel cathodes. Recommended c.d. ranges and treatment times for the various metals are listed. It is also claimed that " rectification " or controlled electrolytic dissolution can be performed in this bath.—H. A. H.

***The Study of Metal Surfaces by Means of Plastic Replicas, [Light Microscopy], and the Phase-Contrast Process.** W. Faasch, G. Hein, and L. Koch (*Metalloberfläche*, 1950, [A], 4, (9), 129–135).—A study is reported of the adaptation of the well-known plastic-replica techniques used in electron microscopy to the examination of metal surfaces—particularly electrodeposits—by microscopy with transmitted or reflected light, with and without the use of phase contrast. Of the replica methods which were tried, " Triafol "—a 2% soln. of cellulose acetobutyrate—and 8% soln. of lacquer in amyl acetate were the most satisfactory. Examination of the replicas by microscopy with transmitted light was found to be superior to examination by reflected light or examination of the actual metal surfaces by the met. microscope. Structural differences were accentuated, and further improvement was obtained by the use of phase contrast. Examination of the surfaces of electrodeposits of Zn, Cd, and of etched Al foil confirmed these findings, although no improvement over normal metallographic methods was found in the case of electrodeposits of Ni or anodically polished brass. The only defect in the method is the difficulty encountered in separating the plastic films from the metal surfaces, which often leads to tearing. 29 photomicrographs show the results obtained. —E. N.

***Replica Techniques in Electron Microscopy.** A. W. Agar and R. S. M. Revell (*Brit. J. Appl. Physics*, 1951, 2, (1), 8–11 ; erratum, (2), 60).—Basing themselves on studies of pearlite, bainite, and martensite in steel, A. and R. claim that it is possible to obtain electron micrographs of replicas giving consistent and reliable interpretation. A modified form of the Formvar replica technique was used, and shadow casting and pre-shadowing with a Au–Pd alloy were found useful.—R. W. C.

Preparing Rod Specimens for X-Ray Photography. D. Summers-Smith (*Metallurgia*, 1951, 43, (257), 154).—A small jig for preparing specimens ~0·015 in. sq. and at least 0·25 in. long is described and illustrated.—F. M. L.

Powder Diffraction Patterns from Microsamples. L. K. Frevel and H. C. Anderson (*Acta Cryst.*, 1951, 4, (2), 186).—[In English]. A note describing a modification to the usual Debye–Scherrer arrangement for obtaining good-quality powder patterns from samples weighing as little as 1 γ. The sample is picked up on the sharp apex of a lead wedge, which is oscillated during the exposure about the apex ridge as axis. Specimen patterns are shown.—R. W. C.

Using the Bragg–Brentano Method [of Powder Diffraction] in Conjunction with Fine-Focus Sealed-Off X-Ray Tubes. Henri Brasseur (*Acta Cryst.*, 1951, **4**, (2), 189).—[In French]. A note advocating a focusing diffraction arrangement in which the (fine) X-ray focus lies on the circumference of the cylinder to which the film is shaped. The method is stated to enable a great reduction of exposure time, and has the merit of simplicity.—R. W. C.

Bibliography of Electron Microscopy. (Cosslett). See col. 91.

5 — POWDER METALLURGY

***Hot Pressing Beryllium Powder.** A. U. Seybolt, J. P. Frandsen, and R. M. Linsmayer (*Steel*, 1950, **126**, (13), 71–74, 96).—The authors have reviewed the earlier work on hot-pressing of Be powder and have investigated the factors to be considered. By the use of the apparatus described, it is possible to make compacts up to 12 in. in dia. and 4 in. thick, and of theoretical *d*. The analyses and properties of the samples are given and compared with those of cast Be.—E. J.

Sintered Metal Bearings. [Reservoil]. —— (*Mass Prod.*, 1951, **27**, (2), 38–51).—The manufacture and appn. of "Reservoil" sintered metal bearings by the Morgan Crucible Co., Ltd., are described.—R. W. R.

New Self-Lubricating Bearings for Machinery. [Deva]. H. Winkelmann (*Metall*, 1950, **4**, (23/24), 504–505).—A review of the prodn., properties, appn., and advantages of "Deva" metal for self-lubricating bearings. It is made by powder-metallurgy methods, and its main feature is the high proportion (>5%) of graphite which it contains. The usual difficulty of obtaining satisfactory diffusion of the alloying components in the presence of such large amounts of graphite is overcome by carrying out the sintering by H.F. (500–2000 c./s.) inductive heating of the compact (contained in a graphite crucible), whereby the intense H.F. eddy currents set up within the compact facilitate the diffusion process and carry it to completion.—E. N.

***Hot-Dip Galvanizing of Powder-Iron Compacts.** (Bablik, Götzl, and Kukaczka). See col. 40.

The Manufacture of Cemented Tungsten Carbide. B. E. Berry (*Murex Rev.*, 1951, **1**, (8), 165–183).—B. considers some of the problems involved in the manufacturing processes concerned. The prodn. of tungstic acid from ore is described, followed by some consideration of the important properties of W powder. The manufacture of WC is then discussed. Hot pressing, which is used for large pieces of hard metal or for pellets of intricate bore shape, is also described, and B. concludes with a short description of dies and tools.—R. J.

Hard-Metal Bits for Rock Drills. Hans Steiner (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (11), 205–217).—The development, mech. properties, and performance characteristics of hard metals—based on mixtures of the carbides of W, Ti, and Ta, +8–13·5% Co—used for bits for rock drills, are described.—E. N.

The Production of Metal Powders. H. W. Greenwood (*Mass Prod.*, 1951, **27**, (3), 68–72).—A short account of the prodn. and properties of carbonyl Fe, Cu, and other powdered metals.—R. W. R.

***Research on the Mechanism of Fritting.** P. Laurent and M. Eudier (*Rev. Mét.*, 1951, **48**, (4), 271–275; discussion, 275).—L. and E. investigated the effect of time and temp. on metal powder compressed at elevated temp. using Fe and Cu as examples. The mechanisms already put forward are: (1) sublimation of the metal from surfaces of large radius of curvature and deposition on surfaces of smaller radius of curvature; (2) surface diffusion of the metal by thermal agitation; (3) diffusion from laminæ under the effect of surface tension; (4) flow under the effect of surface tension or of the attraction of surfaces near points of contact. L. and E. critically review these hypotheses and suggest that the stages of fritting at temp. of the order of 0·8 times the abs. m.p. of the metal and in a reducing atmosphere are as follows: as the temp. rises, the first important operation is the reduction of the surface oxides by the atmosphere. Under the increasing effect of internal stress and surface tension of the grains, a rapid agglomeration of the particles takes place and is momentarily impeded by the reduction of the internal oxides. Later, and at higher temp., the internal stress is released by recrystn. and increase in grain-size; pores are closed and tend towards a spherical form. 11 ref.—J. H. W.

***Grain Boundaries and Sintering [of Silver].** (Greenough). See col. 30.

***Diffusion Phenomena Occurring During Sintering.** Ben H. Alexander (*J. Metals*, 1951, **191**, (2), 95).—Cf. *ibid.*, 1950, **188**, 1219; *Met. Abs.* 1950–51, **18**, 528. An A.I.M.M.E. Research in Progress summary. When compacts are made by simultaneously winding fine wires of two metals (*A* and *B*) on a spool and then sintering, such compacts exhibit all possible types of junctions (*A–A*, *B–B*, *A–B*), although the junctions of dissimilar metals (*A–B*) are present in the greatest number. On measuring the change in size and shape of the interstices between the wires, it is found that although a perfectly uniform bond or "neck" is formed between two wires of the same metal, this is not true when the adjacent wires are dissimilar. In the latter case the bond appears to form by the preferential movement of one kind of atom, with the result that the bond or neck is asymmetric, i.e. it is wider on one side than the other. An overall increase in vol. of the compacts accompanies this peculiar bond formation, the expansion being greater the more asymmetric the bond. Attempts are being made to correlate the vol. changes, which are considerable in some cases (15–20%), with the extent of diffusion or homogenization. Further, the rate at which a bond or neck forms between two unlike metals is found to be about equal to the faster of the rates of bond formation between the metals in like pairs. The systems being studied are Ag–Cu, Cu–Ni, Ag–Au, Cu–Fe, Cu–Zn, Ag–Ni, Cu–W, Ag–Zn, and Au–Ni, all of which give a wide range of mutual solubilities, and consequently greatly differing amounts of diffusion upon sintering.—E. N.

***Routine Measurement of Paint Pigments and Other Fine Powders.—II.** P. C. Carman and P. le R. Malherbe (*J. Appl. Chem.*, 1951, **1**, (3), 105–108).—[Part I. is of no metallurgical interest]. Extension of a previous study on the measurement of surface dia. of various types of fine powders (*J. Soc. Chem. Ind.*, 1950, **69**, 134) includes consideration of two samples of W powders used by Burden and Barker (*J. Inst. Metals*, 1948–49, **75**, 51; *Met. Abs.*, 1949–50, **17**, 817). For both samples air permeabilities at the lower porosities (i.e. tightest compression in a vice as compared with tightest hand-packing) gave similar sp.-surface results, equivalent to a surface mean dia. of $\sim$0·3 μ. Detn. by the N-adsorption method were in reasonable agreement with results obtained from air-permeability measurements on one of the W-powder samples, but no such agreement was found in the case of the other sample, which, it is considered, may be an example of a powder with porous particles.—J. R.

Material Calculations in Works Engaged in Powder-Metallurgy Processes. Hermann Franssen (*Metall*, 1950, **4**, (21/22), 484–486).—A discussion of the prodn. and use of balance sheets relating to the input and output of raw materials and finished products, resp.; they are not only of financial interest, since powder-metallurgy processes consume and produce large quantities of expensive materials, but are also of great tech. importance.—E. N.

Treatise on Powder Metallurgy. Vol. II. (Goetzel). See col. 93.

6 — CORROSION AND RELATED PHENOMENA

***Corrosion-Resistance of Electropolished and Anodized Aluminium Reflectors.** W. Kerth (*Metalloberfläche*, 1950, [A], 4, (12), 190).—Al reflectors which had been electrolytically polished and then anodized (Eloxal process) were exposed to various corrosive media with the following results: (1) flue gases, at 70°–110° C., contg. up to 0·4% SO_2 together with a trace of SO_3 showed little or no effect at the end of 4 months, reflectivity being unimpaired after washing off the accumulated dirt and soot; (2) dil. H_2SO_4 (1 : 3) vapour had no effect after 3 months; and (3) vapour of conc. H_2SO_4 produced a few isolated pin-points of corrosion after 4 weeks, which had doubled in number after a further period of 4 weeks. It is concluded that such Al reflectors are excellent for searchlights and floodlighting, as they are virtually incorrodible under service conditions.—E. N.

***Weathering of Some Aluminium Alloys.** —— (*Light Metals*, 1950, 13, (147), 215–221).—An account is given of the behaviour of 2S, 3S, 50S, and 51S Al alloys after long exposure to rural, indust., and maritime atmospheres, and the depth of penetration to be expected per annum under various conditions of exposure is estimated. Methods of cleaning weathered surfaces to restore or improve the surface appearance are suggested: rubbing with wet steel wool, followed by washing with soap and water, is usually effective. A number of case histories are given.—J. E. T.

Durability of Aluminium and Its Alloys.—I.–II. —— (*Light Metals*, 1950, 13, (147), 195–203; (149), 344–349).—[I.—] The problem of durability is considered in a general manner, and a comparison is made between the properties of Al alloys and the heavy metals. The durability of the various classes of light alloy is described, and attention is paid to the types of corrosion that may be encountered. [II.—] The susceptibility of various types of alloy to intercryst. corrosion is considered, and galvanic corrosion is discussed. —J. E. T.

†The Nature and Mechanism of Stress-Corrosion in Wrought Aluminium Alloys.—I.–IV. H. Vosskühler (*Werkstoffe u. Korrosion*, 1950, 1, (4), 143–153; (5), 179–184; (8), 310–320; (9), 357–366).—A critical survey of stress-corrosion in Al alloys contg.: (*a*) Mg 3–9 and Zn $\not>$1%, (*b*) Mg and Zn in various ratios, with a total content of $\not>$10%, and (*c*) Cu 3–4·5% and Mg and Mn ~1% each. Stress-corrosion is one form of grain-boundary corrosion, of which the other two forms are stress-free intercryst. corrosion and layer corrosion, and various combinations of these three forms are usually present together. For stress-corrosion to occur the alloy must: (*a*) be exposed to attack by a suitable corrosive medium, (*b*) have mech. stresses present within it, and (*c*) be in a condition which renders it susceptible to stress-corrosion. These factors are discussed in relation to: (1) type of stress-corrosion tests—(i) the shape of the specimen, (ii) the nature and magnitude of the stresses—whether causing a const. stress or a const. deformation, and (iii) the nature and condition of the corrosive medium—the compn. of the atmosphere or soln., and whether exposure to it is continuous or intermittent; (2) the cryst. structure of the material—(i) the solid soln. concentration, and (ii) the grain-size and texture; (3) the method and degree of hot or cold working—rolling, forging, extrusion, &c.; and (4) the heat-treatment—(i) the time and temp. of soln.-treatment, (ii) the nature of the quench, and (iii) the time and temp. of cold or warm ageing and whether or not the material has been given a small degree ($\not>$10%) of cold work immediately after soln.-treatment and quenching. There are numerous photomicrographs. 51 ref.—E. N.

***Corrosion and Stress-Corrosion Properties of a High-Strength Aluminium–Zinc–Copper–Magnesium Casting Alloy.** R. A. Quadt and E. C. Reichard (*Trans. Amer. Found. Soc.*, 1950, 58, 525–529; discussion, 529).—Mech. properties of 0·25-in.-dia. test-bars of Al casting alloy contg. Zn 7·4, Cu 0·8, and Mg 0·4% were determined after exposure to 3% aerated NaCl soln. Under these conditions of testing the specimens are unaffected by loads up to 80% of their yield strength. Above this stress, cracking occurs as a result of stress-corrosion.—V. K.

Some Examples of Chemical Damage to Building Construction [Roof Coverings of Aluminium Foil]. A. Leon (*Werkstoffe u. Korrosion*, 1950, 1, (12), 473–476).—Includes some examples of damage to roof coverings made of Al foil, due to the metal being in direct contact with concrete, bitumen, or with Fe or galvanized Fe nails.—E. N.

Corrosion-Resistance of Copper and Copper Alloys. —— (*Chem. Eng.*, 1951, 58, (1), 108–112).—Typical indust. uses of Cu alloys and a corrosion-rating chart are given.—D. M. D.

***The Rate of Solution of Copper in Various Salt Solutions, and the Mechanism of the Corrosion of Copper.** W. Katz (*Werkstoffe u. Korrosion*, 1950, 1, (10), 393–399).—In an attempt to find an explanation for the various rates at which Cu dissolves in different types of salt soln., a study was made—by means of measurements of the polarization and cathodic currents, and of the thickness of the layers of O atoms on the surface of the metal—of its rate of soln. in: ammoniacal 0·5*N*-NH_4Cl, 0·5*N*-NaCl (*p*H = 2), 0·5*N*-KNO_3 (*p*H = 2), 0·5*N*-Na_2SO_4 (*p*H = 3), *M*/15-Na_3PO_4 (*p*H = 4·5), and 0·5*N*-CH_3COONa. The results show that the corrosion of Cu is a cathodically controlled process, and that this premise affords an explanation for the increasing rate of attack (with time or Cu content of the soln.) which occurs, for example, in acid-buffered chloride soln. As the Cu content of the soln. increases, the surface films of O decrease in thickness by amounts which vary with the nature of the soln.: thus, in the chloride soln. the decrease is from 36 atoms, initially, to a fraction (0·17) of a layer at a $Cu^{\cdot\cdot}$ content of 10 mg./50 ml. soln., while in the acetate soln. the decrease is from 26 to 6·2 at 20 mg. $Cu^{\cdot\cdot}$/50 ml. soln. This decrease in film thickness is an indication of release from the normal O depolarization process, resulting from an increase in the rate of discharge of $Cu^{\cdot\cdot}$ ions to $Cu^{\cdot}$ ions at the cathode. This reaction predominates when high rates of corrosion occur; it is most noticeable when the soln. contains complex-forming anions such as those of Cl and NH_3, and is less noticeable in Na_2SO_4 soln.; it does not occur in phosphate or acetate soln., owing to the probable formation of protective films of a cathodic nature on the surface of the Cu. 6 ref.—E. N.

***Studies in the Corrosion of Metals Occasioned by Aqueous Solutions of Some Surface-Active Agents. I.—Copper.** H. Holness and T. K. Ross (*J. Appl. Chem.*, 1951, 1, (4), 158–169).—The action of dil. aq. soln. of representative anionic, cationic, and non-ionic wetting agents on sheet Cu has been studied, using London tap-water, distilled water, and tap-water softened by the Zeolite base-exchange process. The effects of (1) varying the *p*H of the soln. from 2 to 12 and (2) temp. changes were also investigated. Most of the anion-active agents studied first coated the Cu with a water-repellent film affording protection to the metal in the cases of primary and secondary Na alkyl sulphates and Na alkyl aryl sulphonate, but not with sulphonated castor oil or triethanolamine lauryl sulphate; with the cationic agents no water-repellent film was formed, and only triethanolamine attacked the Cu to any great extent. There was no obvious action with the non-ionic agent. In general, attack increased with increasing temp., but changes in soln. *p*H had no marked effect on the course of the corrosion. The results are graphically expressed and a mechanism is suggested. 32 ref.—J. R.

***Corrosion Phenomena in Very Dilute Aqueous Solutions. IV.—Copper.** L. W. Haase (*Werkstoff u. Korrosion*, 1950, 1, (10), 390–393).—Cf. *ibid.*, (4), 129; *Met. Abs.*, 1950–51, 18, 533. H. discusses the solubility of Cu in conc. and dil. aq. acid and salt soln., and in hot and cold H_2O, and the effects of small alloying addn. of Ag, Sn, As, and Sb which result in further decreases of solubility in very dil. soln. The mechanism of the corrosion process is noted, and it is pointed out that although local-cell effects do occur in very weak soln., they are extremely small with Cu, and soon cease, owing to the formation of thin, continuous films of corrosion products whose solubility is so low that they render the Cu virtually incorrodible; only under extremely unfavourable conditions do local cells persist long enough to cause pitting of the metal surface. 9 ref.—E. N.

Corrosion in [Copper Alloy] Preheater Tubes. W. Katz (*Metalloberfläche*, 1950, [A], **4**, (11), 161–168).—A description of sp. examples of electrochem. corrosion encountered under service conditions in tubes used for preheating hot salt (chlorides and sulphates) soln.: (1) local cells set up by the motor-electric effect and dissolved O, inside tubes of a Cu–Sn (5%) alloy, (2) intercryst. corrosion inside tubes of a Cu–Al (4%) alloy, and (3) stress-corrosion, on both the soln. and steam sides, of tubes made from a Cu–Al (4–5%) alloy. Such corrosion often arises from the use of materials in which the working stresses have not been fully relieved. Suggestions are made as to the remedies in materials, construction, and operation which are necessary to minimize damage.—E. N.

Stressed Brass [Lip-Stick] Bases. H. H. Symonds. J. C. Chaston (*Metal Ind.*, 1951, **78**, (17), 350; (19), 381).—The cracking of brass lip-stick bases was due to stress cracking brought about by conditions that included the hardness of the metal, the circumferential tensile stress in the walls of the pressings, and the pitted nature of the surface. The absence of any significant corrosion showed that the failure could not be due to season-cracking. In correspondence, C. suggests that on the evidence presented, the failures described are typical examples of season-cracking.—J. H. W.

Cracked [Brass] Kettle Bodies. H. H. Symonds (*Metal Ind.*, 1951, **78**, (19), 389).—Samples of cracked brass kettle bodies showed tarnishing both inside and out, and cracking was found to be due to the period of time during which the spinnings were kept in stock under unsatisfactory conditions. —J. H. W.

Blemished [Nickel Silver] Tableware. H. H. Symonds (*Metal Ind.*, 1951, **78**, (18), 367–368).—Ag-plated nickel silver fish knives exhibited surface pits persisting through the deposit. These pits were considered to have their origin in defects in the basis metal.—J. H. W.

***The Oxidation of Lithium and the Alkaline Earth Metals in Liquid Ammonia.** Joseph K. Thompson and Jacob Kleinberg (*J. Amer. Chem. Soc.*, 1951, **73**, (3), 1243–1245).—A study of the oxidation behaviour of Li and the alkaline-earth metals in liq. NH_3, under conditions inhibiting amide formation in the case of Na, is reported. All the metals behave nearly alike under the conditions described, the product in every case contg. chiefly monoxide and a relatively low proportion of peroxide. Dissoln. data for the metals and absorption-spectra studies of the oxidized soln. are discussed.—J. R.

[Nickel-Base] Alloys Resistant to Hydrochloric Acid. B. Waeser (*Werkstoffe u. Korrosion*, 1950, **1**, (9), 345–349).—A review of the resistance of many metals and alloys (especially those with a Ni base) to corrosion by conc. or dil. HCl. The effects of inhibitors are noted. 36 ref.—E. N.

***Periodic Phenomena in the Attack of Metals by Vapour: Application to the Determination of Rates of Corrosion [of Silver and Copper].** René Dubrisay and François Arlet (*Rev. Mét.*, 1951, **48**, (4), 267–270; discussion, 270).—D. and A. investigated the phenomena wherein, if a sheet of Ag or Cu pierced with a hole is placed on a tube contg. a piece of I, a series of concentric coloured rings is formed with the hole as centre. They measured the dia. of the rings under monochromatic (Na) light and applied the formula: $e = kp\lambda$, where e is the thickness of the layer, λ the wave-length (of Na light), p the order of the ring from the centre, and k a coeff. depending solely on the refractive index of the iodide. They tabulated and plotted the results at room temp. and at 24°, 46°, and 60° C. At all these temp. the curves are similar and indicate a complex mechanism, for which various theories have already been put forward. It appears certain that two factors are the rate of attack of the metal by the vapour and the time that the vapour is at the hole. Curves of the thickness of the iodide against the time show that the rate of corrosion is very high at first and diminishes rapidly with the time, and that it increases with the temp. up to a certain point, after which it diminishes—an unexpected result. Above 100° C. the appearance changes completely; some rings are seen at once, and these increase in size, but not in number. With Br no rings are obtained at room temp., but some may appear at 0° C. With electrolytically polished Ag or Cu the results are the same as at temp. above 100° C. and no rings appear. The rate of formation of rings increases with the decrease in size of the tube.—J. H. W.

***Corrosion in Amine Gas-Treating Plants.** F. C. Riesenfeld and C. L. Hughes (*Petroleum Refiner*, 1951, **30**, (2), 97–106).—Results are reported of experiments which have been conducted with pilot-scale and commercial heat-exchangers handling amine soln. Most of the materials tested were ferrous, but a Mn-bearing Al alloy, Monel metal, and Inconel were also used. The experiments indicate that the Mn-bearing Al alloy is a suitable material for use in heat-exchangers in glycol-amine plants, and that Monel and Inconel show good resistance to corrosion in aq. amine systems operating under a variety of conditions. 8 ref.—G. T. C.

Corrosion and Field Lining of Sulphate Digesters. T. T. Collins, Jr., S. J. Baisch, and G. T. Vander Velden (*Paper Trade J.*, 1951, **132**, (6), 24, 26–27; (7), 18, 20, 22, 24, 26).—A general survey is made of corrosion in sulphate digesters used in the pulp and paper industry. After giving a review of the problem with special ref. to the preservation of existing units, the authors describe how 630 ft.^{2} of a digester was lined with welded Inconel strips; details and illustrations of the procedure are given. 62 ref.—J. R.

Corrosion Control for Cooling Systems of Electric Apparatus. Emil J. Remscheid (*Gen. Elect. Rev.*, 1951, **54**, (3), 36–40).—The corrosion problems associated with water-cooled elect. gear, and the methods available for overcoming them, are briefly described. Special ref. is made to water-cooled Hg-arc rectifiers; Na_2CrO_4 addn. to the coolant eliminates pitting corrosion of the Fe cooling system encountered when using certain types of H_2O.—D. K. W.

†The Influence of Internal Stresses on Corrosion. (A Review). E. Franke (*Werkstoffe u. Korrosion*, 1950, **1**, (10), 404–412).—A review of the literature dealing with: (1) origin and nature of internal stresses, (2) the effects of stresses on electrode potential and rate of corrosion, (3) the effects of the nature of the surfaces, (4) stress-corrosion in non-ferrous metals, (5) stress-corrosion in Fe and steel, and (6) corrosion-fatigue. 202 ref.—E. N.

Contact Corrosion in Industrial Plant. J. F. Stirling (*Pract. Eng.*, 1951, **23**, (576), 294–296).—A general review of the cause of contact corrosion and of the ways of preventing it.—E. J.

Electrolytic Corrosion of Metals. A. E. Durkin (*Gen. Elect. Rev.*, 1951, **54**, (2), 45–48).—The more important elements and some of the commonly used alloys are placed in an electromotive series, and a brief explanation is given of the manner in which metals that are to be used in contact with one another under corrosive conditions should be selected, in order to reduce electrolytic corrosion. 10 ref.—D. K. W.

The Origin and Nature of Galvanic Currents [Electrochemical Corrosion]. L. W. Haase (*Werkstoffe u. Korrosion*, 1950, **1**, (8), 297–299).—A brief review.—E. N.

***A New Instrument for the Measurement of the Galvanic Currents of Corrosion.** H. P. Godard (*Rev. Mét.*, 1951, **48**, (4), 267–282).—Galvanic corrosion is defined as the accelerated corrosion that a metal undergoes when in contact with another and the two are exposed to a corrosive medium. G. briefly describes the factors influencing galvanic corrosion, viz. soln. potential, polarization, relative surfaces of anode and cathode, diffusion, elect. contact, and the condition of the electrolyte. He investigated this phenomenon by: (1) measuring the potential of a metal couple in open circuit, (2) measuring the potential of a couple compared with a ref. couple, (3) measuring the direction and intensity of the const. current of corrosion, (4) comparing the pinholes, loss in weight, diminution of mech. properties of specimens, either coupled or single, and (5) determining the polarization characteristics of the specimens in the test soln. He discusses the practical appn. of this investigation to such problems as the corrosion of Alclad and other metal-protected alloys.—J. H. W.

The Use of Potential Measurements in Corrosion Research. G. Masing (*Werkstoffe u. Korrosion*, 1950, **1**, (11), 433–437).—Theoretical and practical aspects of the use of potential measurements in studying corrosion phenomena are discussed, and illustrated with ref. to the results obtained with Al, an Al–Cu–Mg alloy, Fe, and Zn in various buffered aq. salt soln. 4 ref.—E. N.

Practical and Relatively Simple Corrosion Tests Found Effective. F. L. LaQue (*Petroleum Refiner*, 1951, **30**, (2), 120–122).—Details are given of the design of various types of holder for corrosion specimens which may be inserted in the actual plant at the corrosive location.—G. T. C.

New Corrosion-Resistant Materials [in Use] Abroad [Outside Germany].—III.-V. Erich Franke (*Werkstoffe u. Korrosion*, 1950, **1**, (8), 305–308; (9), 349–357; (12), 497–506).—Cf. *ibid.*, (5), 171; *Met. Abs.*, 1950–51, **18**, 533. The prodn., the mech. and corrosion-resistance properties, and the appn. of the following materials are described in some detail: [III.—] British Ni–Mo and Ni–Cr–Mo sand-casting alloys, viz. Langalloys, contg. Ni 56–85, Mo 0–30, Cr 0–15, Si 0·75–10, and Cu 0–3%, which correspond to the American Hastelloy and Chlorimet alloys. 2 ref. [IV.—] Cu–Ni alloys contg. small amounts of Fe, for condenser tubes, i.e. alloys of Cu with Ni 10–30, Fe ≯1·8, and Mn ≯1·2%. 35 ref. [V.—] Organic polymers of the ethylene tetrafluoride type, known as Teflon in the U.S.A., and Fluon in Britain. 36 ref.—E. N.

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7 — PROTECTION

(Other than by Electrodeposition.)

Sulphuric Acid Anodizing: Further Patents Available to Industry. —— (*Electroplating*, 1951, **4**, (3), 93).—Short summaries are presented of 11 British patents relating to anodizing processes or the treatment of anodic films, which have expired recently.—H. A. H.

Notes on the Production of Aluminium Reflector Surfaces. —— (*Electroplating*, 1951, **4**, (4), 121–122).—Hints are given on points to observe in the manufacture of high-quality Al reflectors. The factors considered include: material of manufacture, polishing, cleaning, and degreasing, electrolytic brightening, anodizing, and rinsing.—H. A. H.

Surface-Treated Aluminium in Architecture. H. v. Uslar (*Metalloberfläche*, 1951, [B], **2**, (11), 161–165; (12), 183–186).—U. discusses the choice of: (1) suitable alloys, in relation to structural considerations, and (2) mech. polishing, and chem. and anodic oxidation procedures for obtaining the desired surface finish.—E. N.

Chemical Surface-Treatment Extends the Durability of Kool-Vent All-Weather Aluminium Awnings. Norman P. Gentieu (*Products Finishing*, 1951, **15**, (7), 38–43).—Brief information is given on the design of the awnings. They are treated by the Alodizing process, details of which are not included.—G. T. C.

Surface Treatment and Finishing of Light Metals.—VII. S. Wernick and R. Pinner (*Sheet Metal Ind.*, 1951, **28**, (288), 373–379, 383).—Cf. *ibid.*, (285), 79; *Met. Abs.*, 1950–51, **18**, 666. Before considering details of indust. anodizing plants, W. and P. discuss some of the general principles applicable to all, such as pretreatment, racking for anodizing, equipment, &c. The chromic acid process is then dealt with in detail, including the American accelerated chromic acid process, and the modified process developed by Slunder and Pray (*Indust. and Eng. Chem.*, 1946, **38**, 592; *Met. Abs.*, 1946, **13**, 420). 20 ref.—R. J.

Alloyed Chromized Coatings. —— (*Product Finishing (Lond.)*, 1951, **4**, (4), 58–59).—A note on the prodn. of alloyed coatings by modifications of the chromizing process. Typical coatings considered include: Cr–Al, Cr–Zr, Cr–Si, &c. —H. A. H.

Methods of Testing Lacquer Coatings on Tinplate. W. E. Hoare (*Sheet Metal Ind.*, 1951, **28**, (287), 231–233, 244).—Methods of testing stoved or air-dried lacquer films on tinplate are described. The main uses of such lacquers are: (1) protection of exterior surfaces, (2) protection of interior surfaces, (3) protection of contents of a container from metal contamination, and (4) protection of decoration or legend. The types of lacquer and the required properties are briefly discussed. The main tests are: (*a*) continuity tests, (*b*) film weight or thickness-of-coating tests, (*c*) adhesion tests, (*d*) bend tests, and (*e*) abrasion and hardness tests. 20 ref. —R. J.

***A Survey of Coating-Thickness Variation on Commercially Galvanized Sheets.** M. L. Hughes (*J. Iron Steel Inst.*, 1951, **167**, (1), 48–65).—H. describes the results of a study of the distribution of coating thickness on 13 representative samples of galvanized sheet made at 9 different works. The results of measurements made along, across, and on both sides of the sheets are presented graphically. The thickness detn. were made by a stripping method using a 2% soln. of Sb_2O_3 in HCl. It is shown that the "oversheet" average coating thickness is surprisingly const. when sheets from a single batch are considered; the individual sheets themselves may, however, show marked longitudinal and transverse variations in coating thickness. In some cases there is a tendency for the lower face of a sheet to be poorer in Zn. The possible causes of the observed coating-thickness variations are discussed.—R. W. R.

Revamped Production Facilities Boost Pipe Galvanizing 30%. Dan Reebel (*Steel*, 1950, **126**, (8), 84–86).—Describes, with illustrations, a rebuilt pickling and galvanizing plant to treat all sizes of pipes from ½ to 3 in. in dia., in both standard and extra heavy grades. The consumption of Zn (with the addn. of 0·2–0·3 lb. of Al/ton) is considerably reduced as a result of the modernization.—E. J.

***Hot-Dip Galvanizing of Powder-Iron Compacts.** H. Bablik, F. Götzl, and R. Kukaczka (*Sheet Metal Ind.*, 1951, **28**, (287), 277–281).—The fundamental differences between powder Fe and low-C mild steel are the lack of scale, the porosity, and the large surface area available for reaction. Pickling treatments must therefore be short. No difficulty is experienced from entrapped pickle liquor, as this is completely evaporated out by immersion in the molten Zn bath. The main conclusions drawn from this work were: (1) hot-dip galvanizing of powder-Fe compacts is practicable and the corrosion-resistance is equal to that of hot-dip galvanized rolled Fe; (2) a short pickling operation is employed and the dry-galvanizing process is used; (3) addn. of Al are ineffective; and (4) U.T.S. is slightly reduced.—R. J.

Construction of a Small-Scale Sheet-Galvanizing Unit. M. L. Hughes and I. E. Thomas (*J. Iron Steel Inst.*, 1951, **167**, (1), 46–47).—A fairly detailed account is given of the construction of and methods of operation of a small-scale galvanizing unit (capacity 6 cwt. of Zn) for the experimental galvanizing of sheet 8 in. wide, in a manner similar to that employed in industry. The sheet is passed by means of a system of rolls, through the gas-fired pot at six speeds in the range 6–40 ft./min.—R. W. R.

The Metallurgy of Metal Spraying [Especially Zinc]. K. Krekeler (*Metalloberfläche*, 1950, [A], **4**, (10), 151–153).—After a brief outline of the oxy-acetylene wire process, K. discusses the properties—adhesion, flexibility, and corrosion resistance—of the coatings obtained in relation to: (1) the nature of the basis metal, (2) the particle size of the sand used for sand-blasting preparatory to spraying, (3) the thickness of the coating, and (4) the operating conditions of the pistol, namely (*a*) wire dia. and rate of feed, and (*b*) the nature of the

flame—whether oxidizing, neutral, or reducing. Photomicrographs and graphs illustrate the effects of these variables in relation to the properties of sprayed Zn coatings.
—E. N.

The Metallurgy of Metal Spraying. Karl Krekeler (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (12), 340–342).—A brief review of the processes, and of the properties—adhesion, ductility, and porosity—of the coatings produced.—E. N.

Recent Developments in Metal Spraying by the Powder Process. W. G. McDermott (*Electroplating*, 1951, **4**, (4), 114–116, 131).—Details are given of a new metal-spraying pistol for use with the powder process. This pressure-fed gun is claimed to have many advantages over the standard-type induction-fed pistol. These include the prodn. of denser coatings possessing greater adhesion to the base metal and a very low oxide content, and increased speed of appn., resulting in a lower cost for better coating.—H. A. H.

Metal Spraying in France: A Survey of Recent Developments. J. Cauchetier (*Electroplating*, 1951, **4**, (5), 163–164).
—H. A. H.

Joint Cathodic-Protection Systems. —— (*Petroleum Eng.*, 1951, **23**, (2), D53–D55).—Bulletin No. 4 issued by the Correlating Committee on Cathodic Protection, sponsored by a number of organizations. Information is presented on techniques for the design and installation of systems to protect neighbouring structures of two or more operators. The possible division of responsibilities for construction, operation, and maintenance is discussed, and so is the sharing of costs.
—G. T. C.

The Protection of Electric Equipment from Corrosion in Pulp and Paper Mills. G. W. Knapp (*Paper Trade J.*, 1951, **132**, (3), 18, 20, 22, 24, 26).—J. R.

Corrosion Protection in Breweries. H. Netz (*Werkstoffe u. Korrosion*, 1950, **1**, (11), 444–446).—N. discusses recommended procedures and materials (paints, varnishes, enamels, &c.) for the most efficient protection of machinery, apparatus, and pipes used in breweries.—E. N.

Soft Plastic Materials for Corrosion Protection. Hj. Sæchtlin (*Werkstoffe u. Korrosion*, 1950, **1**, (9), 367).—Cf. *ibid.*, (6/7), 251; *Met. Abs.*, 1950–51, **18**, 538. Corrigenda and some supplementary observations.—E. N.

The Influence of [the Nature of] the Pigment on the Corrosion Protection Afforded by Paints and Lacquers. C. Kalauch (*Werkstoffe u. Korrosion*, 1950, **1**, (10), 400–404).—For the successful corrosion protection of metal surfaces by paints and lacquers, the nature of the pigment and its phys.-chem. properties are all-important. K. discusses the influence of these factors in relation to the painting and lacquering of steel sheet. 11 ref.—E. N.

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8 — ELECTRODEPOSITION

Corrosion-Resistant Chromium Deposits. H. Silman (*Metal Ind.*, 1951, **78**, (17), 327–336).—S. considers the factors that influence corrosion-resistance of Cr deposits and, in particular, discusses the effects of the structure, hardness, gas content, degree of polishing, and the c.d. of the bath on the resistance.
—J. H. W.

Faulty Chromium Deposits. H. H. Symonds (*Metal Ind.*, 1951, **78**, (15), 293).—Micro-examination revealed a cracked condition of the Cr on Cr-plated jack-spindles having irregularities in the Cr deposit, but no defects in the basis metal were found.—J. H. W.

***Experiments in Chromium Plating from Non-Aqueous Media.** A. L. Hanson, D. Frokjer, and D. Mitchell (*Metal Finishing*, 1951, **49**, (4), 48–49, 69).—Experimental details are presented of attempts to deposit Cr from a number of organic materials. The most satisfactory were: (i) formamide, (ii) acetamide, and (iii) acetamide–urea mixture. Less satisfactory were acetic acid and ethanolamine, and very little success was obtained with acetonitrile and dimethyl formamide. It is shown that Cr^{3+} will deposit more effectively from non-aq. soln. than will Cr^{6+}. C.d. of 5–15 amp./dm.^{2} were required to produce satisfactory deposits. The addn. of a little H_2SO_4 to the baths increased the conductivity, but it had no appreciable affect on the plating when Cr^{3+} was used, though it improved it when Cr^{6+} was used. 4 ref.
—G. T. C.

Copper as a Finishing Material: Review of 1950. (——). See col. 70.

***The Electrodeposition of Copper–Lead Alloys.** A. L. Ferguson and Nelson W. Hovey (*J. Electrochem. Soc.*, 1951, **98**, (4), 146–154).—An investigation is described of a number of soln. from which Cu–Pb alloys might be deposited. A cyanide–tartrate soln. appeared to be the most promising of these and was investigated in detail. The preferred compn. of the soln. was: NaCN 147, CuCN 150·6, NaOH 40·0, $NaKC_4H_4O_6.4H_2O$ 211·7, $Pb(C_2H_3O_2)_2.3H_2O$ 75·6 g./l. Using a rotating cathode, the effects were studied of variations in concentration of NaCN, tartrate, acetate, and CuCN, the replacement of tartrate by citrate, *p*H, NH_3, dilution, certain addn. agents, the use of Cu–Pb alloy anodes, c.d., temp., and cathode rotation, in so far as these factors affected alloy compn., single-metal potentials, and alloy potentials. Only the ratio of NaCN : CuCN appears to be critical as far as deposit compn. is concerned. Under optimum conditions, bright smooth deposits may be obtained, but these deposits are hard and brittle unless the Pb content is relatively high. The soln. is not stable over a long period of time unless it is stored out of contact with air. Although it is stated that the bath described is not suitable for commercial use, especially for thick deposits and for the higher Cu contents, it is claimed that fundamental information has been obtained which may be useful for developing a commercial bath. 11 ref.
—G. T. C.

***Ageing Effects in Copper–Lead Alloy Plating Solutions of the Cyanide–Tartrate Type.** Nelson W. Hovey, Albertine Krohn, and A. L. Ferguson (*J. Electrochem. Soc.*, 1951, **98**, (4), 155–159).—Cf. preceding abstract. The effects of ageing in cyanide–tartrate plating soln. for Cu–Pb alloys were studied during a 50-day period by measurements of single-metal potentials, static and dynamic alloy potentials, oxidation–reduction potentials, alloy compn., and absorption spectra of the plating soln. Similar measurements on a portion of the soln. stored under N showed that the cause of ageing was atmospheric oxidation. Although the fact that ageing takes place suggests that the soln. might not be satisfactory for commercial appn., the potential measurements indicate that in a continuous plating process, the Cu would be kept in the reduced state, so that no great change would occur in the compn. It is also suggested that by diluting the soln. and adding more free cyanide, the ageing effect might be minimized or completely eliminated. The dilution would increase the difference between the static potentials of Cu and Pb, leading to deposits of higher Cu content, but increase in the free cyanide would counterbalance this effect. Excess cyanide might help to prevent air oxidation because of its reducing properties and the common-ion effect which it exerts on the cuprocyanide complex. Work is proceeding.—G. T. C.

The Corrosion-Resistance of Nickel and Chromium Electrodeposits. III.—How to Produce a Satisfactory Coating. J. Liger (*Galvano*, 1951, (168), 11–14).—Cf. *ibid.*, 1950, (167), 11; *Met. Abs.*, 1950–51, **18**, 613. The factors contributing to the prodn. of a satisfactory Ni electrodeposit are considered. They include coating thickness, plating soln. operation and control, Cu plating before Ni deposition, and surface prepn. and polishing.—H. A. H.

***Plating Aluminium with Nickel.** J. M. Bryan (*Metal Ind.*, 1951, **78**, (20), 405).—A satisfactory laboratory etching process for preparing the surface of Al for plating was im-

practicable for commercial appn., because the high concentration of trichloroacetic acid rendered it too costly and the use of acetone for washing involved high fire-risk. The addn. of 25% H_3PO_4 allowed the concentration of the trichloroacetic acid to be reduced to 5%, and a brief dip in 1 : 4 HNO_3 enabled water to be used instead of acetone for washing. The compn. of the Ni-plating soln. was that devised by Bullough and Gardam (*J. Electrodepositors' Tech. Soc.*, 1947, **22**, 169; *Met. Abs.*, 1947–48, **15**, 418) containing: $NiSO_4.7H_2O$ 120, $(NH_4)_2SO_4$ 21, KCl 8 g., H_2O to 1 l. The bath is worked at $p\mathrm{H} = 5{\cdot}6$–$5{\cdot}8$ and 35° C. for 80 min., with a c.d. of 15 amp./ft.^{2}.
—J. H. W.

Industrial Nickel Plating in the United States. W. H. Prine (*Metal Ind.*, 1951, **78**, (19), 383–388).—P. tabulates bath compn. and plating conditions, phys. properties of the deposits, and recommended pretreatment cycles, and describes the methods of plating adopted in the United States for : (1) decorative plating for appearance and corrosion-resistance, (2) plating for salvage by building-up, (3) heavy indust. deposits for corrosion-resistance, (4) plating non-conductors, and (5) electroforming.—J. H. W.

What Platers Can Do About the Nickel Shortage. —— (*Steel*, 1951, **128**, (14), 81–83).—By the use of synthetic lacquers, more Cu, Zn, and other coatings, it is shown to be possible to eliminate or reduce Ni consumption in conventional Cu–Ni–Cr finishes.—E. J.

Silver Plating [on Brass and Other Copper-Base Alloys] with an Intermediate Nickel[-Silver] Plating. A. Pollack (*Metalloberfläche*, 1950, [B], **2**, (10), 151).—When economic considerations necessitate the use of thin Ag electrodeposits on Cu-base alloy articles, the following procedure is recommended in order to obtain white, adherent deposits: (1) bright-Ni plate in the usual type of bath; (2) treat in a Ag–Ni bath contg. $NiSO_4.7H_2O$ 50, NaCN 50, and AgCN 1·2 g./l.; the operating conditions are: distance between electrodes 15 cm., 4–5 V. for the first 30 sec., dropping to 1·5–2 V. during the next 2–3 min., and final holding at this voltage for 2–3 min.; and (3) Ag plate in the usual way for 5–15 min. For very cheap articles the final pure Ag coating can be omitted.—E. N.

***Service Behaviour of Electrodeposited Silver–Lead Bearings in Internal-Combustion Engines.** (Rossenbeck). See col. 82.

Tin Plating [Practice in the U.S.A.]. Frederick A. Lowenheim (*Metal Ind.*, 1951, **78**, (21), 423–425).—L. discusses the nature of the problem and the aims and scope of Sn plating as it is carried out in the U.S.A., as regards the electrotinning of cold-rolled strip and the general jobbing and manufacturing. In addn. to the electrotinning process, he describes immersion methods, "liq. finishing", the tinning of Al-alloy pistons, and the use of Sn alloys for plating.—J. H. W.

Anode Maintenance in the Alkaline Stannate Tin-Plating Bath. F. A. Lowenheim (*Metal Finishing*, 1951, **49**, (3), 60–64).—The importance of proper anode control and maintenance in ensuring good work from the alkaline stannate bath is emphasized. The differences between Sn anodes and anodes of other metals used in plating are described, and means of recognizing when an anode is working properly are explained. The use of oxidizing agents is recommended only as a first-aid measure.—G. T. C.

Automatic Barrel Plating [of Zinc] : New Canning Plant at the British Screw Co., Ltd. —— (*Metal Ind.*, 1951, **78**, (13), 247–250).—Describes an automatic barrel Zn-plating plant for the treatment of screws, bolts, and nuts.—J. H. W.

***Determination of Impurities in Electroplating Solutions. XXI.—Traces of Lead in Zinc-Plating Baths.** Earl J. Serfass and Mary H. Perry (*Plating*, 1951, **38**, (5), 473–476).—Cf. *ibid.*, 1950, **37**, 1057; *Met. Abs.*, 1950–51, **18**, 378. Work is described which has been carried out under Research Project No. 2 of the American Electroplaters' Society. In the method evolved, Pb is separated from interfering ions by extraction with a CCl_4 soln. of dithizone in the presence of KCN. After removal of excess dithizone with alkaline KCN soln., the transmittance of the CCl_4 soln. is measured and the quantity of Pb determined from a calibration curve. Full details of the experimental procedure are included, and the method of obtaining the calibration curve is described. A "stop-and-go" method is included, which allows an approx. detn. of the Pb to be made by visual comparison of colours without the need for a colorimeter. The method can be used to determine the Pb in both cyanide and sulphate baths.
—G. T. C.

***Attempts To Electrodeposit Zirconium.** M. L. Holt (*J. Electrochem. Soc.*, 1951, **98**, (3), 33C–35C). Work is described which was directed at establishing whether or not Zr can be electrodeposited from a soln. A number of aq. and non-aq. soln. were used, and attempts were made to deposit Zr and, in some cases, Zr alloys from them. No weighable deposit of pure Zr was obtained from any of the soln. tried, though in a few cases very thin, slimy, metallic-appearing deposits were obtained. These, however, faded or oxidized a few sec. after removal from the plating soln. Weighable deposits contg. Zr and Fe were obtained from some soln. It is pointed out that when the activity of Zr is considered it is not surprising that the metal has not been electrodeposited from aq. soln. It is suggested that organic soln. should offer the most promise, since Al has been deposited from several such soln. No investigation was made with fused baths. 14 ref.—G. T. C.

Review of Electrodeposition of Metals and Alloys. J. W. Cuthbertson (*Product Finishing* (*Lond.*), 1951, **4**, (1), 67–74).—Abridged from *Proc. Inst. Elect. Eng.*, 1950, [I], **97**, (104), 43; *Met. Abs.*, 1949–50, **17**, 947.—H. A. H.

Plating Stainless Steel. H. H. Head (*Metal Ind.*, 1951, **78**, (15), 287–290).—H. gives details of Cd (over Ni strike), Cu, Cr, Au, Ni, and Ag plating of stainless steel.—J. H. W.

The Industrial Applications of Electrodeposited Coatings. P. Morisset (*Usine nouvelle*, **1951**, (Numéro spécial), 73, 75, 77, 79–80).—A comprehensive survey of the characteristics and indust. appn. of electrodeposited coatings. A number of particularly useful tables are included. In the first of these, the min. and average coating thickness is specified when the metal coating is plated on to steel, Zn, Cu, and Al and their resp. alloys intended for indust. or decorative purposes. The electrodeposited metals include Ni (with a Cu undercoat), Cr after Ni plating, Cu, Zn, Cd, Pb, and Sn (on Cu deposits), Zn–Sn (80 : 20), Ag and Au (on Cu deposits), Rh, Pt, and Pd (on Ni or Au). Ref. are given for these specifications. Further tables show recommended Ni coating thicknesses for different corrosive conditions relative to housing and motor cars and when applied to Fe, Cu, Zn, and Al and their resp. alloys. The properties of the electrodeposited coatings are briefly described; hard Ni and Cr plating receive detailed attention.—H. A. H.

Manufacturing Parts by Electrodeposition. A. G. Arend (*Mass Prod.*, 1951, **27**, (3), 80–81).—A short account.
—R. W. R.

Making Electrotypes. —— (*Plating*, 1951, **38**, (5), 465–468).—A series of photographs with descriptive captions is presented which illustrate the various stages in the prodn. of electrotypes.—G. T. C.

Rotogravure Roll Making. —— (*Plating*, 1951, **38**, (3), 248–252).—A series of photographs with descriptive captions describing the various electroplating operations used for this purpose.—G. T. C.

Plating Plastics. J. F. Stirling (*Pract. Eng.*, 1951, **23**, (573), 224–225).—After reviewing other methods of coating plastics, such as metal spraying, vacuum evaporation, and sputtering, S. considers the electroplating process. The plastic component is first cleaned and degreased, and then roughened and sealed, after which a conducting layer is deposited on the part. This may be graphitic or composed of bronzing lacquers. The article is then Cu plated, and subsequently any of the usual metals may be electrodeposited.
—E. J.

***Electrodeposits as Bases for Paints.** (Silman and Wernick). See col. 71.

Plating Improved with Reverse-Current Techniques. G. W. Slomin (*Iron Age*, 1951, **167**, (6), 94–96).—Periodic reverse-current plating makes possible greater plating speed and smoother, finer-grained deposits. The process combines plating and electrolytic polishing in one operation, the current reversal reducing surface imperfections, holding edge build-up to a min., and improving brightness.—J. H. W.

Formation of Stains on Electrodeposits. Walter Burkart (*Metalloberfläche*, 1950, [B], 2, (9), 134–135).—Inefficient degreasing and the use of unsuitable polishing powders for buffing, preliminary to plating, are fruitful sources of the stains often found on the final electrodeposits. For the best results: (1) degreasing should be carried out (*a*) in organic solvents or (*b*) electrolytically, and (2) the polishing powder should (i) be almost neutral, with a *p*H $\sim$7; if it is excessively acid or alkaline it reacts (except in the case of the noble metals) with the metal, forming, in the presence of air, relatively thick oxide films which are only partly dissolved during plating, and (ii) be S-free. Methods are described for the detn. of the *p*H of the paste, and for the detection of S therein. —E. N.

Drag-Out Losses in Wire Plating. I.—Determination of Drag-Out and Liquid Film Thickness. Joseph B. Kushner (*Wire and Wire Products*, 1950, **25**, (5), 414–415, 418).—A math. method is given for determining drag-out loss and liq. film thickness. Two practical examples are calculated: (1) the amount of Au lost during the Au plating of wire, and (2) the thickness of the liq. film during the Zn plating of 0·025-in.-dia. wire.—W. A. M. P.

Drag-Out Losses in Wire Plating.—II.–III. Joseph B. Kushner (*Wire and Wire Products*, 1950, **25**, (8), 646–648, 695–697).—Methods of reducing drag-out losses and the use of recovery rinse baths are discussed.—W. A. M. P.

Strike Baths. J. B. Mohler (*Metal Finishing*, 1951, **49**, (3), 57–59, 69).—The purpose and function of strike baths are described, and information is presented on the compn. and operating conditions for Sn, Cu, and Ag strike baths.—G. T. C.

The Function of Sulphites and Bisulphites in Cyanide Plating Solutions. Joseph Haas (*Metal Finishing*, 1951, **49**, (4), 64–65).—Two years' use of sulphites in cyanide soln. have indicated that they reduce wastage of cyanide, promote anode efficiency, and result in the prodn. of a lustrous deposit. Bisulphite is recommended for use in place of dil. H_2SO_4 or $NaHCO_3$ as a means of lowering the *p*H of the baths when this is necessary. The addn. also reduces the excess cyanide content of the bath.—G. T. C.

The Structure of Electrodeposited Metals. A. W. Hothersall (*J. Electrodepositors' Tech. Soc.*, 1950, 25, 203–210).—A simple qual. picture of the structure of electrodeposits is presented, with particular ref. to the manner in which it is affected by the conditions of electrodeposition and by the structure and surface geometry of the basis metal. Among the conclusions reached is one that the structure of electrodeposited metals is often influenced by the structure of the basis metal but, except for certain coarse-grained deposits formed on a coarsely cryst. base which is not too dissimilar in crystal form and atomic spacing, the effect of the structure of the basis metal is usually confined to a thickness of deposit too small to be visible under the microscope. Beyond this thickness, the structure of the deposit is governed solely by the conditions of electrodeposition. On the other hand, effects produced by the geometry of the surface persist for a much greater thickness through the deposit, and may lead to misinterpretation of the structure of the deposit. The interference theory of metal deposition is briefly described, namely that the size of the crystals of an electrodeposit, which determines its hardness, is controlled by the degree to which molecules or ions obstruct the free movement of the metal cations near or on the cathode surface. If the interference is small, large crystals are formed; if great, a fine-grained, hard deposit is obtained. —G. T. C.

***A Microhardness Tester for Metals at Elevated Temperatures. [Hardness of Electrodeposits].** (Brenner). See col. 55.

Practical Implications of Throwing Power. S. Wernick (*J. Electrodepositors' Tech. Soc.*, 1950, 25, 83–92; discussion, 93–94; and (abridged) *Metal Ind.*, 1950, **76**, (2), 28–29; (4), 65–67).—The importance of throwing power when endeavouring to comply with thickness specifications is emphasized. A number of simple tests for the throwing power of a Ni soln. are described, using specimens of various shapes. In order to avoid troubles due to poor throwing power, it is suggested that attention be given to: (1) the shaping of anodes and the contour of the cathode; (2) the use of "robbers" and/or insulators which are placed at strategic points in the soln.; and (3) planned racking of work, so that the individual components on a rack help one another in "spreading" the Ni to the best advantage. Further research work is advocated with a view to establishing methods of achieving max. throwing power from electrolytes at present in use and in order to develop new or modified electrolytes with increased throwing power.—G. T. C.

The Measurement of Throwing Power and the Prediction of Thickness Distribution. G. E. Gardam (*J. Electrodepositors' Tech. Soc.*, 1950, 25, 77–81; discussion, 93–94; and (abridged) *Metal Ind.*, 1950, **76**, (2), 28–29; (4), 65–67).—The Haring and Blum method of determining throwing power is criticized as being difficult to carry out and as producing in certain soln. results which are worthless to the practical plater for the following reasons: (1) the results do not depend on the linear ratio but do depend on the scale of the experiment; (2) the test involves current going along two entirely separate paths between the anode and the two cathodes, whereas in commercial practice the path of the current is more usually common as far as a point very near the cathode; (3) the scale of magnitude of the current distribution on the objects commercially plated varies in a ratio which may largely exceed Haring and Blum's linear ratio of 5 : 1, but is usually quite unknown and incalculable. The results of G.'s investigations suggest that it is much more important to persuade designers that the articles provided for plating should be simple and free from recessed portions than that the plater should endeavour to undo the harm of a badly designed article by using a soln. of excessive throwing power. He also suggests that the results obtained by the Haring and Blum cell are no more than indicative in a practical case, and much theoretical work would need to be done before the distribution of deposit could be predicted on an actual article. —G. T. C.

Continuous Electroplating Effects Operational Economy. Dan Reebel (*Steel*, 1951, **128**, (8), 64–66).—A description, with illustrations, is given of a continuous automatic plant which cleans, plates with Zn or Cd, rinses, and dries large quantities of small components.—E. J.

Maintenance in the Plating Shop. J. M. Beskine (*Product Finishing* (*Lond.*), 1951, **4**, (3), 92–98, 100, 102, 104).—Recommendations are made for the maintenance of the elect. equipment in electroplating shops.—H. A. H.

Bath Temperatures and Bath Heating in Electroplating. Heinz W. Dettner (*Metalloberfläche*, 1950, [B], 2, (12), 181–182).—The introduction of higher temp. for electroplating baths has resulted in the use of high c.d., and the prodn. of deposits which are more adherent and ductile. Recommended bath temp. and, consequently, c.d., voltage, and compn. are given for the deposition of Cu and its alloys, Zn, Cd, Ni, Cr, and Au. Details are given of the types of the thermostatically controlled immersion heaters which are normally used for warming the vats.—E. N.

Organic Tank Linings: An Outline of Care and Maintenance. Manson Glover and Gunnar Lindh (*Plating*, 1951, **38**, (5), 471–472).—General information is presented on tanks suitable for lining, tank installation, on means of preventing damage by heat or mech. causes, and on controlling corrosion. —G. T. C.

Recovery of Chromic Acid from Plating Rinse Waters. E. W. Neben and W. F. Swanton (*Plating*, 1951, **38**, (5), 457–460, 470).—A unit composed of glass-lined and stainless-steel parts for the recovery of chromic acid is described. The indications are that the life of this equipment is likely to be satisfactory, even if handling material from baths in which a fluoride catalyst is used. The process is essentially one of vacuum evaporation, and the concentrate may be used for making up the plating bath. Data are presented on the selection of a suitable size of unit and of the economics of operating such a unit. 3 ref.—G. T. C.

Cyanide Waste Disposal and Control. C. F. Hauri (*Plating*, 1951, **38**, (3), 240–242, 265).—Steps taken in one particular plant for disposal of cyanide and other wastes are described. These include: (1) reclamation of cyanide immediately after operations where it was used; (2) a collection system for all waste paint, grease, oil, solvent, &c.; previously discharged into the sewer, they are now burned; (3) non-cyanide alkaline wastes were disposed of in conjunction with spent pickling liquors, thus avoiding undue influence on the *p*H of the sewage; (4) all cyanide cleaning soln., dips, &c., were discontinued; (5) destruction of all solid or conc. plating wastes and cyanide-bearing heat-treating salts was initiated.—G. T. C.

Plating Wastes: Treatment and Disposal. J. E. Cooper (*Plating*, 1951, **38**, (4), 346–352, 357).—Sources of plating wastes are briefly outlined, and methods of disposal are classified as: (1) modifications in design or operation or both within the manufacturing process to minimize or eliminate the waste problem, and (2) installation of a chem. treatment to destroy toxic and objectionable wastes in plating-shop effluent. Various suggestions are made for the modification of plating shop design, including provision of arrangements whereby all spillage, over-flow, &c., is collected into emergency holding tanks, of which each shop would require at least two, one for acidic and the other for alkaline wastes. Reclaim rinse-tanks should follow all plating baths. Various methods for the treatment of cyanide-contg. wastes are briefly described and alkaline chlorination is the one preferred. Cr wastes are most readily treated by reduction and pptn. by lime. SO_2 and Na metabisulphite are preferred reducing agents, but if there is a local source of waste acid pickling liquor, $FeSO_4$ may prove advantageous. A large indust. waste-treatment plant is described. 27 ref.—G. T. C.

Methods of Determining *p*H in the Plating Shop. L. J. Howells (*Australasian Plating and Finishing*, 1950, **1**, (1), 11–13, 15).—A brief review of colorimetric and electrometric methods.—T. A. H.

Water in the Plating Shop. —— (*Product Finishing* (*Lond.*), 1951, **4**, (3), 40–50).—A detailed survey of all aspects of the use of water in the plating shop. 8 ref.—H. A. H.

Review of Australian Standard Specifications for Plating. I. E. Miller (*Australasian Plating and Finishing*, 1950, **1**, (1), 19–20).—A brief review.—T. A. H.

The Possible Adaptability of the German Metal Working [Electroplating] Industry to Modern Methods. Karl Sommer (*Metalloberfläche*, 1950, [B], **2**, (10), 147–149).—The various factors which have led to stagnation in the German electroplating industry are noted, and the advantages of adopting modern methods—illustrated by a comparison of old and new techniques in Ni plating—are discussed.—E. N.

The Electroplating Industry in Atlanta.—I.–II. Victor J. Baran (*Plating*, 1951, **38**, (3), 243–247; (5), 464, 469–470).—[I.—] A detailed survey of the general nature and scope of the electroplating industry in Atlanta is presented. [II.—] Comments are made regarding anodizing, Cd, Cr, Ni, Ag, and Au plating as carried out in Atlanta.—G. T. C.

A Modern Electroplating Shop. —— (*Sheet Metal Ind.*, 1951, **28**, (288), 380–383).—A description of the new electroplating shop of the Westinghouse Brake and Signal Co., Ltd. —R. J.

Time and Motion Studies as a Basis for Estimating Costs [in Electroplating, Cleaning, and Finishing Processes]. W. Schmalz (*Metalloberfläche*, 1950, [B], **2**, (11), 169–171). —E. N.

A Study of Job Plating and Its Future. O. Altschul (*Electroplating*, 1951, **4**, (3), 81–83, 97).—A short history of jobbing plating with special ref. to economic factors. Recommendations are made for ensuring a satisfactory future for the industry.—H. A. H.

Estimating and Quoting Plating Prices. Jerome L. Bleiweis (*Metal Finishing*, 1951, **49**, (3), 67–69).—A costing system for use in metal finishing is described.—G. T. C.

Plating Costs in a War Economy. W. O. Zinn (*Products Finishing*, 1951, **15**, (6), 48–49).—A brief general discussion. —G. T. C.

The History of Electrodeposition. R. C. Davies (*Sheet Metal Ind.*, 1951, **28**, (289), 477–482).—R. J.

9—ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

***Contribution to the Understanding of the Electrometallurgy of Aluminium.** J. E. Boner (*Helv. Chim. Acta*, 1950, **33**, (5), 1137–1153).—[In German]. The reactions which probably occur in the electrolytic prodn. of Al are critically considered, and an attempt is made, based on recent theories of complex formation and thermal dissociation, to correlate the conflicting information in the literature of the subject. 31 ref.—E. N.

Electrothermal Processes in Aluminium Metallurgy. [—II.] Arturo Paoloni (*J. Four Elect.*, 1951, **60**, (1), 13–15).—Cf. *ibid.*, 1950, **59**, 140; *Met. Abs.*, 1950–51, **18**, 615. A summary of patents covering (*a*) some indirect processes for the thermal prodn. of Al; (*b*) methods for the electrolytic recovery of Al from electrothermal alloys.—H. A. H.

***Metallurgical Investigation of the Electrothermal Reduction of Cobalt[-Iron-Manganese] Mixed Oxide and the Refining of Cobalt.** Willy Hilgers (*Z. Erzberg. u. Metallhüttenwesen*, 1950, **3**, (10), 334–341).—A report of a study of the optimum conditions for the electrothermal prodn. of crude Co contg. Fe 2–3, S 0·045–0·08, Cu 0·4–0·5, Ni 0·04–0·06%, and traces of Si, from Co–Mn–Fe mixed oxide. A ferro (15–25% Fe)–Co alloy is recovered from the various slags produced in the process. 13 ref.—E. N.

***The Electrolytic Precipitation of Metallic Copper Powders: Influence of Mechanical Obstacles Very Near the Cathode.** N. Ibl and G. Trümpler (*Helv. Chim. Acta*, 1950, **33**, (5), 1370–1378).—[In German]. A study is reported of the electrolytic pptn. (at 0·2–0·5 V.) of Cu from $CuSO_4$ soln. (⅓ g.-mol./l.) on to a horizontal cathode which has a mech. obstacle in its immediate neighbourhood—in this case a thick sludge of $BaSO_4$ on the cathode itself. The results show that under these conditions a powder precipitate of Cu is obtained at c.d. 0·7–0·2 m.amp./cm². after some 4–6 hr. which are some 100 times smaller than those necessary for the formation of powder under normal conditions. The phenomenon is attributed to the $BaSO_4$ sludge suppressing the normal convection currents in the soln., thereby producing only a very thin film of Cu^{++} ions on the cathode. 12 ref. —E. N.

***Electrokinetics of Hydrogen Evolution. II.—Deuterium Overvoltage on Mercury.** B. Post and C. F. Hiskey (*J. Amer. Chem. Soc.*, 1951, **73**, (1), 161–164).—Measurements of D overvoltage on Hg cathodes at c.d. of 10^{-6}–>10^{-3} amp./cm.² and temp. from 4° to 70° C. are compared with previous detn. of H overvoltage made under similar conditions (*ibid.*, 1950, **72**, 4203; *Met. Abs.*, 1950–51, **18**, 381); the D overvoltages were 50–70 mV. higher than the comparable values for H overvoltage. Electrolytic-sepn. factors are computed from the data, and agree very well with experimental results. The overvoltage and sepn. data derived appear to preclude the possibility that the slow step of the H-discharge process involves more than one ion or atom.—J. R.

***Phase-Boundary Potentials of Nickel in Foreign-Ion Solutions.** D. MacGillavry, J. J. Singer, Jr., and J. H. Rosenbaum (*J. Amer. Chem. Soc.*, 1951, **73**, (3), 1388).—A letter. Stable potentials of Ni are plotted as a function of *p*H in (1) KOH and (2) 0·100*M*-phosphate buffer soln. Curves for the phosphate buffers show analogy with those obtained by El Wakkad and Salem (*J. Phys. Colloid Chem.*, 1950, **54,** 1371) for Hg, and by Tourky and El W. (*J. Chem. Soc.*, **1948,** 740 and 749; *Met. Abs.*, 1948–49, **16,** 569 and 570) for Cu. Tests of the soln. for Ni were always negative, except when O was not excluded.—J. R.

***The Kinetics of the Decomposition of Sodium Amalgam [in Aqueous Electrolytes].—I.** G. Trümpler and K. Gut (*Helv. Chim. Acta*, 1950, **33,** (6), 1922–1933).—[In German]. A report of a study of the cathodic decompn., at various c.d., of pure Na amalgam in electrolytes with borate buffers of pH = 8·2–9·2 at 25° C. 7 ref.—E. N.

***Polymerization Induced by Hydrogen in Metals.** Giuseppe Parravano (*J. Amer. Chem. Soc.*, 1951, **73,** (2), 628–630).—H liberated during electrolysis or adsorbed at metal surfaces is shown to cause initiation of polymerization in soln. of methyl methacrylate, the efficiency of the process increasing with increasing overvoltage of the cathode.—J. R.

***Theoretical Analysis of the Current-Density Distribution in Electrolytic Cells.** Carl Wagner (*J. Electrochem. Soc.*, 1951, **98,** (3), 116–128).—The current distribution at plane electrodes of finite breadth is calculated under the assumption of a linear polarization curve. The current distribution at triangular-wave profile electrodes is calculated for zero polarization and an approx. soln. is given for high polarization. Results of the theoretical analysis are compared to experimental data reported by Mantzell. It is shown how measurements with the Haring cell can be used to predict c.d. distributions for other geometrical conditions. 14 ref.—G. T. C.

***On the Thermodynamics of Leclanché-Type Cells.** C. Drotschmann (*Metall*, 1950, **4,** (23/24), 507–510).—The theoretical e.m.f. of fresh Leclanché-type cells is calculated from the thermochem. data contained in the literature dealing with the various stages of the reactions which occur within the cell. The results are in good agreement with values determined experimentally.—E. N.

10 — REFINING

***The Separation of [Aluminium] Alloy [Constituents] by Centrifuging.** Walter Roth (*Z. Erzberg. u. Metallhüttenwesen*, 1950, **3,** (10), 328–334).—During the solidification of a melt, the d of the primary crystals separating out at any given temp. is usually different from that of the residual melt and segregation occurs. It should be possible, therefore, theoretically, to increase the degree of segregation by centrifuging, although during the course of solidification: (1) in hypoeutectic alloys, the main constituent of which is less dense than the alloying metal, the differences in d fluctuate and may even become reversed, i.e. compared with the residual melt the primary crystals may be less dense at high temp. and more dense at low temp.; (2) the shape and size of the primary crystals may alter; and (3) the viscosity of the melt changes. Small-scale experiments on 70-g. melts of binary Al alloys show that in hypereutectic alloys, where the primary crystals consist of intermetallic compounds, it is possible to accomplish, with little difficulty, separations which seem feasible from the equilibrium diagram. In hypoeutectic alloys, however, little separation is possible, owing to the dendritic structure of the pasty, Al-rich solid-soln. primary crystals formed. The binary alloys studied contained Cu (5 and 20), Mg (5 and 20), Zn (4 and 20), Fe (5), and Si (25%). It is considered that the development of suitable large metal centrifuges would lead to the initiation of new segregation processes. —E. N.

11 — ANALYSIS

***Spectrochemical Analysis of Bismuth Matrices: Porous-Electrode Technique.** Joseph C. Delaney and Louis E. Owen (*Analyt. Chem.*, 1951, **23,** (4), 577–580).—For the determination of 0·0001–0·5% of Be, Co, Cr, Fe, Mn, Mo, Nb, Ni, Ti, V, and/or W in Bi or Bi–Pb eutectic, the sample is dissolved in HNO_3 and the soln. spectrographed by the porous-cup technique after addn. of Pt to serve as an internal standard. —F. M. L.

***Polarographic Determination of Cadmium in Aluminium Alloys.** William Stross (*Metallurgia*, 1951, **43,** (257), 145–147).—The sample is dissolved by first attacking with NaOH soln. and then boiling with HNO_3. $(NH_4)_2SO_4$ is added to the soln., which is then electrolysed to remove Cu and Pb. Any Fe in the soln. is reduced with hydroxylamine at pH = 3, gelatin is added, and, after expelling O with H or N, the Cd is determined polarographically.—F. M. L.

***Determination of Microgramme Amounts of Calcium and Magnesium in Vanadium Metal.** Joseph Rynasiewicz, Ruth Guenther, M. E. Sleeper, and R. H. Gale (*Analyt. Chem.*, 1951, **23,** (4), 668–669).—Up to 1 g. V is dissolved in HNO_3 and the soln. evaporated to dryness several times with HCl. The V is removed by adding cupferron to the chloride soln. and extracting the V cupferride into chloroform. Ca and Mg are determined in the residual soln. by normal procedures. —F. M. L.

***Spectrophotometric Determination of Ce(IV).** A. I. Medalia and B. J. Byrne (*Analyt. Chem.*, 1951, **23,** (3), 453–456).—0·025 γ Ce(IV) can be detected by measuring the absorbency at 320 mμ. Care must be taken not to use too large an amount of persulphate, since NH_4', NO_3', and persulphate ions absorb appreciably at this wave-length. 10 ref.—F. M. L.

***Copper in White Metals: A Photometric Method of Determination.** W. C. Coppins and J. W. Price (*Metal Ind.*, 1951, **78,** (11), 203–204).—Dissolve the sample in Br–HBr, evaporate to dryness, cool, and dilute to 50 ml. with HBr in a stoppered graduated bottle. Measure the extinction on a Spekker absorptiometer and calculate the Cu content by a comparison with calibration curves. Pb-base alloys require stirring and Sn-base alloys give a little residue. When the Fe content in the Pb-base alloys exceeds that of the Cu, correct on the basis of 0·1% Fe ≡ 0·005% Cu; no correction for Fe is required in the Cu-base alloys.—J. H. W.

***Electrolytic Separation of Copper from Arsenic.** George Norwitz (*Analyst*, 1951, **76,** (901), 236–237).—0·1–0·5 g. Cu can be separated from as much as 0·75 g. As by electrolysis from a HNO_3 soln. contg. 50 g. NH_4NO_3 to which is added H_2O_2 to oxidize As to the pentavalent state.—F. M. L.

***Volumetric Determination of Iron in Aluminium Alloys.** George Norwitz (*Metallurgia*, 1951, **43,** (257), 154).—The sample of Al alloy is dissolved in HCl, Zn pellets are added, and after 2 min. the soln. is filtered. 3–4 drops of saturated $HgCl_2$ soln. are added, and air is bubbled through the soln. to oxidize Ti. Fe is then titrated, after addn. of H_3PO_4, with standard $K_2Cr_2O_7$, using diphenylamine as indicator.—F. M. L.

***Potentiometric Method for Microdetermination of Iron: Extension of Use of Silver Reductor.** Ibert C. Wells (*Analyt. Chem.*, 1951, **23,** (3), 511–514).—0·27–0·5 mg. Fe may be determined with an accuracy of 2‰ by reducing $FeCl_3$ in a scaled-down Walden Ag reductor and then potentiometrically titrating with standard $Ce(SO_4)_2$ soln.—F. M. L.

Colorimetric Determination of Iron: A Review of Known Methods.—I. T. S. West (*Metallurgia*, 1951, **43,** (258), 204–206).—The detn. of Fe with *o*-phenanthroline and with α,α′-dipyridyl is discussed. 21 ref.—F. M. L.

***Separation of Iridium from Lead on a Micro Scale: Gravimetric Determination of Iridium.** R. R. Barefoot, W. J. McDonnell, and F. E. Beamish (*Analyt. Chem.*, 1951, **23,** (3), 514–516).—Soln., resulting from fire assays, contg.

small amounts of Ir and Pb may be treated in two ways for the recovery of Ir. The Pb may be removed as phosphate and the Ir then determined by hydrolysis to $Ir(OH)_4$, filtering, ignition, and reduction in H. Alternatively, Ir may be precipitated in an acetic acid–NH_4 acetate soln. with 2-mercaptobenzothiazole and the precipitate filtered off and treated as above.—F. M. L.

***Determination of Lead as Basic Lead Chromate and Its Application to the Separation of Lead from Silver.** F. C. Guthrie and J. T. Nance (*J. Appl. Chem.*, 1951, **1**, (3), 109–111).—Detn. of Pb by pptn. of the basic chromate $PbO.PbCrO_4$ provides a ready means for the quant. analysis of Ag–Pb mixtures. To the neutral boiling soln. contg. Pb and Ag a soln. of K_2CrO_4 is gradually added; the Pb initially precipitates as the pale yellow chromate, the colour changing owing to formation of dark red Ag_2CrO_4 when sufficient K_2CrO_4 has been added. On addn. of sufficient NH_3 the Ag_2CrO_4 dissolves and orange-coloured $PbO.PbCrO_4$ is formed, settling readily to leave a clear yellow filterable liquid from which the Ag can be precipitated as bromide. Results of detn. by the method on Pb alone and with increasing proportions of Ag are tabulated.—J. R.

***Determination of Impurities in Electroplating Solutions. XXI.—Traces of Lead in Zinc-Plating Baths.** (Serfass and Perry). See col. 43.

***The Determination of Lead Oxide in the Presence of Lead.** R. M. Black (*Analyst*, 1951, **76**, (901), 208–210; discussion, 210–211).—For the detn. of PbO in Pb dross, a weighed sample of the dross is disintegrated by gently heating with Hg. The mixture is cooled, aq. NH_4Cl added, and the soln. distilled for 15 min. into 2% H_3BO_3 soln. The NH_4 borate formed from the NH_3 evolved by the decompn. of NH_4Cl by PbO is titrated with 0·1N-HCl, using a mixture of bromocresol green and methyl red as indicator.—F. M. L.

***Precision Colorimetry: Application to Determination of Manganese.** Irving G. Young and C. F. Hiskey (*Analyt. Chem.*, 1951, **23**, (3), 506–508).—The transmittancy-ratio approach to colorimetric measurements has been investigated with ref. to the detn. of Mn by oxidation with periodate followed by a comparison of the unknown with a known soln. of high absorbency. A precision of 1‰ is shown to be readily obtainable.—F. M. L.

***Colorimetric Determination of Nickel with Alpha-Furildioxime.** A. R. Gahler and A. M. Mitchell, with M. G. Mellon (*Analyt. Chem.*, 1951, **23**, (3), 500–503).—At $pH = 7{\cdot}5$–$8{\cdot}3$ the Ni complex of α-furildioxime can be extracted into 1 : 2-dichlorobenzene and the colour determined spectrophotometrically.—F. M. L.

***Differential Spectrophotometric Determination of High Percentages of Nickel.** Robert Bastian (*Analyt. Chem.*, 1951, **23**, (4), 580–586).—Ni is determined in materials contg. 98–100% Ni with an accuracy of ±0·05% by comparing on a spectrophotometer the absorbency of a perchlorate soln. of the sample and standards contg. similar known amounts of Ni.—F. M. L.

***Spectrophotometric Determination of Nickel in Calcium Metal Using 1,2-Cyclohexanedionedioxime [Nioxime].** Raymond C. Ferguson and Charles V. Banks (*Analyt. Chem.*, 1951, **23**, (3), 448–453).—2 g. Ca is added to 50 c.c. H_2O and when the reaction is complete 5 c.c. HCl and 1 c.c. HNO_3 are added and the soln. heated to dissolve $Ca(OH)_2$ and oxidize any Fe. The soln. is cooled, NH_4 citrate added and the pH adjusted with aq. NH_3 to 4–6. To an aliquot part are then added 1 c.c. each of 10% gum arabic soln. and 0·8% nioxime soln. After standing 1 hr., the absorbency is measured at 550 mμ. The red inner complex of Ni with nioxime is stabilized by gum arabic and is insensitive to pH changes. 24 ref.—F. M. L.

***Colorimetric Determination of Rhenium.** Emil E. Malouf and Merwin G. White (*Analyt. Chem.*, 1951, **23**, (3), 497–499).—0·1 γ to 2·5 mg. Re may be determined in the presence of large amounts of Mo by extracting the complex of Mo with Na ethyl xanthate into a 1 : 1 mixture of CCl_4 and C_6H_6 and then extracting the Re-thiocyanate complex into ether and measuring the colour electrophotometrically.—F. M. L.

***The Determination of Sodium in Aluminium and Its Alloys by Vacuum Distillation.** W. McCamley, T. E. L. Scott, and R. Smart (*Analyst*, 1951, **76**, (901), 200–207; discussion, 207).—Na is quantitatively separated from Al by distillation *in vacuo* at 900° C. and is determined in the distillate by the Zn uranyl acetate method. Preferably a vacuum of $<10^{-1}$ mm. Hg is used for the distillation, although for alloys contg. ~0·1% Na a vacuum of 10^{-1} mm. Hg will give reasonable results.—F. M. L.

***Determination of Sodium and Potassium in Lithium Metal by Flame Photometer.** W. R. Inman, R. A. Rogers, and J. A. Fournier (*Analyt. Chem.*, 1951, **23**, (3), 482–483).—As little as 0·001% K and Na may be determined in Li, using a flame photometer.—F. M. L.

***Spectrophotometric Determination of Tellurium as the Iodotellurite Complex.** Ralph A. Johnson and Francis P. Kwan (*Analyt. Chem.*, 1951, **23**, (4), 651–653).—The reddish-yellow colour of the iodotellurite complex is used for the spectrophotometric detn. of small amounts of Te. Se, Bi, Cu, and Fe interfere and must be removed.—F. M. L.

***Volumetric Determination of Thorium by High-Frequency Titrimetry.** W. J. Blaedel and H. V. Malmstadt (*Analyt. Chem.*, 1951, **23**, (3), 471–475).—H.F. titrimetry may be used for the detn. of Th by titration with standard oxalic acid soln.—F. M. L.

***The Determination of Tin, Lead, Copper, Iron, Aluminium, and Cadmium in Samples of Zinc.** Fritz Bertram (*Z. Metallkunde*, 1950, **41**, (7), 213).—A new, rapid, accurate method is described, whereby all the usual impurities in commercial and high-purity Zn can be determined on a single sample. Attack 30 g. turnings by carefully controlled addn. of ~150 ml. HCl (2HCl : $1H_2O$)—taking care that the reaction does not become violent—until only a trace of the Zn remains. Under these conditions the more noble impurities are undissolved and remain as a slime. Filter off the residue and discard the filtrate. Dissolve the precipitate in 700 ml. HNO_3 and 300 ml. H_2O, and evaporate until syrupy; add hot H_2O, boil, and filter off the precipitated $SnO(OH)_2$; dry, ignite, and weigh as SnO_2. To the filtrate add 10 ml. conc. H_2SO_4, fume, add 300 c.c. cold H_2O, boil, and cool to ~60° C.; add 25 ml. C_2H_5OH, stand for 2 hr., filter off the precipitated $PbSO_4$, dry, and weigh. Electrolytically determine the Cu in the filtrate. Add 20 g. NH_4Cl to the residual soln., boil for 5 min., cool, add excess NH_4OH to precipitate Fe and Al as hydroxides, filter off, and determine the Fe and Al grav. or by the Reinhardt–Zimmermann method. To the filtrate add sufficient 2% Na_2S soln. to precipitate CdS and any residual Zn as ZnS, stand for 30 min., and filter. Treat the precipitate with warm 15% H_2SO_4 to dissolve the ZnS, and filter. Dissolve the residual precipitate of CdS in the least possible quantity of slightly diluted HNO_3, add 3 c.c. conc. H_2SO_4, evaporate to dryness, calcine, and weigh the $CdSO_4$.—E. N.

***Polarography of Tin(IV) in Presence of Traces of Tetraphenylarsonium Chloride.** I. M. Kolthoff and Ralph A. Johnson (*Analyt. Chem.*, 1951, **23**, (4), 574–576).—Tetraphenylarsonium chloride catalyses the reduction of Sn^{IV} to Sn^{II} at the dropping Hg cathode and the polarographic wave is appreciably improved. In the presence of a small amount of this reagent, Sn may be readily determined even in the presence of a concentration of Pb up to 0·01M.—F. M. L.

***The Determination of Zinc by Dithizone.** H. Barnes (*Analyst*, 1951, **76**, (901), 220–223).—Up to 5 γ Zn can be separated from up to 15 γ Cu and then determined absorptiometrically by first extracting the bulk of the Cu from a soln. buffered at $pH = 5{\cdot}32$ into a soln. of dithizone in $CHCl_3$. $Na_2S_2O_3$ is added to the aq. layer to suppress further Cu extraction and the Zn is then extracted into a soln. of dithizone in CCl_4.—F. M. L.

***Nephelometric Determination of Very Small Quantities of Zinc.** L. Bertiaux (*Chim. Analyt.*, 1951, **33**, (2), 59).—After reviewing various methods for the detn. of small quantities of Zn, B. describes a nephelometric method suitable for its estn. when in quantities of the order of $\frac{1}{10}$ mg. The Zn soln.

(free from foreign metallic radicles other than those of the alkalies or alkaline earths) is neutralized (using methyl orange as indicator) if acid by NH$_4$OH, if alkaline by HNO$_3$, and then slightly acidified with one drop of dil. HNO$_3$; under these conditions the necessary NH$_4$ salts are then always present in the soln. Add 10 c.c. Na$_3$H(CO$_3$)$_2$.H$_2$O (100 g./l.), dilute to 500 c.c., add 5 c.c. 8-hydroxyquinoline (20 g./l., in C$_2$H$_5$OH), stir, and stand for 15 min. Compare the turbidity obtained with standards. If the Zn is present in quantities $>$0·0001 g., boil the soln. gently for 5 min., filter off the precipitated oxinate on a sintered glass (or paper) filter, and wash with very hot H$_2$O, followed by C$_2$H$_5$OH, until the precipitate is colourless. Dry at 130° C. (or 125° C. in the case of a paper filter) until a const. weight, cool, and weigh. Precipitate × 0·1848 = Zn.—E. N.

***m-Cresoxyacetic Acid: A Selective Reagent for Zirconium.** M. Venkataramaniah and Bh. S. V. Raghava Rao (*Analyt. Chem.*, 1951, **23**, (3), 539–540).—From 0·2–0·25N-HCl Zr can be quant. precipitated and separated from many elements by m-cresoxyacetic acid.—F. M. L.

Colorimetric Determination of Metals in Industrial Wastes. P. G. Butts, A. R. Gahler, and M. G. Mellon (*Metal Finishing*, 1951, **49**, (4), 50–60, 63).—Methods are described for the detn. of Cd, Cr, Cu, Fe, Pb, Mn, Ni, and Zn in sewage and indust. waste waters. The methods permit the detn. of as little as 0·05 mg. of the ion in question in a 1-l. sample and they are applicable in the presence of other heavy metals up to at least 200 times the concentration of the desired constituent. Allowance is also made for the presence of organic matter and such anions as may commonly be found in polluted water. Detailed procedures are given for preliminary treatment of the sample, qual. identification of the heavy metals present, and of their quant. detn.—G. T. C.

***Spectrophotometric Studies of Dithizone and Some Dithizonates: Molecular Extinction Coefficient of Dithizone in Carbon Tetrachloride.** Stancil S. Cooper and (Sister) Mary Louise Sullivan (*Analyt. Chem.*, 1951, **23**, (4), 613–618).—The molar extinction coeff. of dithizone and its compounds with Pb, Zn, Ag, and Hg^{II} have been determined. A method for determining the concentration of a dithizone soln. in CCl$_4$ from its opt., d, and molar extinction coeff. is described.

—F. M. L.

Paper Chromatography. I.—Instruments and Techniques. II.—Physical and Geometric Factors. III.—Kinetic Studies. Ralph H. Müller and Doris L. Clegg (*Analyt. Chem.*, 1951, **23**, (3), 396–403; 403–408; 408–411).—Reviews.

—F. M. L.

Automatic Fraction Collector for Chromatographic Separations. Donald F. Durso, Elwyn D. Schall, and Roy L. Whistler (*Analyt. Chem.*, 1951, **23**, (3), 425–427).—An electronically controlled apparatus for carrying out chromatographic sepn. on a large scale is described. Different developers are fed to a column and the effluents collected in separate receivers.—F. M. L.

***Analysis of Solids with the Mass Spectrometer.** J. G. Gorman, E. J. Jones, and J. A. Hipple (*Analyt. Chem.*, 1951, **23**, (3), 438–440).—A new technique is described, with particular ref. to the detn. of Ni and Cr in stainless steels, which, it is suggested, may find wide appn. in the routine, rapid analysis of solids with the mass spectrometer.—F. M. L.

***Tris(hydroxymethyl)aminomethane as an Acidimetric Standard.** John H. Fossum, Peter C. Markunas, and John A. Riddick (*Analyt. Chem.*, 1951, **23**, (3), 491–493).—Tris(hydroxymethyl)aminomethane has most of the properties required for an ideal primary acidimetric standard.

—F. M. L.

***Further Examination of Reproducibility in Chemical Analysis.** J. O. Lay (*J. Iron Steel Inst.*, 1951, **167**, (1), 42–43).—A statistical comparison was made of analyses of a number of Mn-contg. samples. Both standard chem. and absorptiometric methods were used, and the results were obtained by a number of different operators. It is concluded that the attainment of a within-operator standard deviation of 0·010 and a total standard deviation of 0·015 is a sufficient indication of the satisfactoriness of a method of met. analysis.

—R. W. R.

12—LABORATORY APPARATUS, INSTRUMENTS, &c.

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

***A Simple Torsion Pendulum for Measuring Internal Friction.** M. E. Fine (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1950, **188**, (11), 1322).—A description of the design and construction of a modified form of the apparatus for measuring internal friction described by Kê (*ibid.*, 1948, **176**, 448). It has a low parasitic energy loss or background, and deals with specimens 0·020–0·075 in. in dia., at 1–12 c./s., and max. strains of 0·5–10 × 10^{-5}; with appropriate modification the sample may be placed in a thermostat. The apparatus may also be used to measure the shear modulus.—E. N.

Reduction-of-Area Gauge Helps Determine Mechanical Properties of Metals at Low Temperatures. —— (*Steel*, 1950, **126**, (11), 92–93).—As the simultaneous load and dia. measurements are essential for determining the true stress/strain curve in tension, a new gauge has been developed, and full details of its construction and use are described. It is suitable for use in temp.-controlled baths ranging from −196° to +100° C. with an accuracy of 0·0001 in.

—E. J.

A Simple Instrument for Evaluating Polished or Buffed Surfaces. G. E. Gardam (*Metal Finishing*, 1951, **49**, (4), 61–63).—Cf. *Electrodepositors' Tech. Soc.*, **1950**, (Preprint No. 4); *Met. Abs.*, 1949–50, **17**, 959.—G. T. C.

***Apparatus and Technique for Measuring Thermal Conductivity of Powdered Material.** W. Z. Girton and J. H. Potter (*Amer. Soc. Test. Mat. Bull.*, 1951, (172), 47–52).—Describes the development of apparatus and technique for measuring the thermal conductivity of powdered material. The material was placed in the annular space between two concentric cylinders. The flow of heat was radially outwards, and the quantity of heat was measured by the temp. rise of the cooling water surrounding the outer cylinder. Guard rings were used to minimize end effects. Heat-transfer is a combination of conduction through the solid particles, convection in the gas pockets surrounding the particles, and radiation between adjacent particles, and the results, expressed as a single coeff. designated the " apparent thermal conductivity ", are thus characteristic of the particular sample of powder used. Results are given for powdered Zn, carbonyl Fe, Cu–CuO eutectic, and CaCO$_3$, and it is concluded that they are satisfactorily reproducible.—P. T. G.

Film-Thickness Measurements. E. J. Dunn, Jr. (*Amer. Soc. Test. Mat. Bull.*, 1951, (172), 35–39).—D. describes a number of instruments for measuring the thickness of films of paint, varnish, lacquer, &c. Some of these are based on magnetic principles and are for measuring thicknesses of films on steel, but others can be used also for films on non-magnetic bases. The results are given of several series of comparisons of these various instruments, carried out in different laboratories. The recommended method for films $>$0·5 in. thick involves the use of a dial micrometer, with one of the magnetic methods as an alternative. For films $<$0·5 mil thick the difficulties of measurement become much greater. It is tentatively concluded that one of the magnetic methods is the most satisfactory for the very thin films.

—P. T. G.

***New Techniques and Results in the Measurement of Magnetostriction [of Alloys].** (Goldman). See col. 24.

The Electron Microscope and Its Metallurgical Applications. —— (*Alluminio*, 1947, **16**, (3/4), 117–120).—The principles and uses of the electron microscope are briefly described, a diagram of the circuit involved given, and a comparison is made with the light microscope. Details of its use in investigating pure Al and 99·5% Al and the methods by which the surface replicas are made are given. A description of its use in the laboratories of the (French) Compagnie Générale de T.S.F. is incorporated.—H. J. W.

A Review of Some Electronic Instruments for Use in Industry. S. F. Smith (*Sheet Metal Ind.*, 1951, 28, (287), 247–255; discussion, 255–256).—S. discusses the need for new instruments in industry and describes the various electronic devices available. These include strain-measuring devices, extensometers, equipment for vibration testing, and stroboscopic equipment.—R. J.

13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

***Measurement of Low-Order Ductility.** Norman P. Pinto (*J. Metals*, 1950, **188**, (12), 1444).—A simple apparatus for the measurement of low orders of ductility (1–10% elongation on 2 in.) in specimens of brittle metals is described. The test-bar—round, sq., or rectangular in section—is bent over a former in which the radius of curvature decreases progressively, e.g. from 20 to 2 in., and the elongation is read from a graph relating the radius of curvature at the point of fracture to bar thickness. The test is valid through both the elastic and plastic ranges, and notch effect is minimized; it does not, however, yield precise values of elongation but approximations useful particularly in comparing similar materials.—E. N.

***The Application of the Resonance-Vibration Method to the Fatigue Testing of Spot-Welded Light Alloy Structures.** R. Week (*Weld. Research*, 1950, 4, (2), 33r–38r).—Fatigue tests by the resonance-vibration method on spot-welded light-alloy beams are described, the beams consisting of two top-hat sections of 16-gauge D.T.D. 390 alloy sheet (4·9% Cu–1% Mg–Si–Fe–Mn alloy). The work was exploratory and designed to investigate difficulties arising from the use of this method, and fracture was generally attributable to changes in stiffness, bolt holes, clamping stresses, &c. Experimental procedure was eventually perfected sufficiently to cause fracture in consequence of the alternating bending stresses. The experimental procedure and possibilities for future work are discussed.—W. D. B.

***A Microhardness Tester for Metals at Elevated Temperatures. [Hardness of Electrodeposits.]** Abner Brenner (*Plating*, 1951, 38, (4), 363–366).—The difficulties of designing a hardness tester of this type are briefly outlined, and a description is given of a new type of instrument. This is an indenter type of instrument in which a capillary tube of fused silica is used as an indenter shaft and moves in an inert atmosphere through a dry bearing lubricated with graphite. The clearance between the shaft and the bearing is $\sim$0·001 in., and inert gas escapes slowly through this space during the test. The use of an inert atmosphere is, of course, necessary to avoid oxidation of the metal surface at the elevated temp. The graphite bearing enables the indenter to move with very little friction, which is essential if a light load is to be accurately applied. The apparatus is suitable for determining the hardness of electrodeposits. The procedure for making measurements is described in detail, and some of the information obtained with the instrument is described. It is shown, for instance, that the hardness of soft Cr and hard Cr at 800° C. has dropped in each case to $\sim$100 D.P.N. in contrast to initial values of 600 and 1000 D.P.N., resp. On cooling to room temp. neither electrodeposited Ni nor Cr regains its original hardness. On the other hand, Co and Co alloys regain a large part of their initial hardness and may even increase in hardness if the max. temp. has not been too high. This latter fact is not, however, evidence of good hot hardness. For example, Co–P alloys have a hardness of only 5 D.P.N. at 800° C. The hot hardness of Co–W alloys appears to increase with the W content. The alloy with 31% W has outstanding hot hardness. At 800° C. it is as hard as at room temp. 7 ref.
—G. T. C.

Six Advantages of Scratch-Hardness Testing. John H. Hruska (*Weld. Eng.*, 1950, 35, (4), 17–19, 22).—A description is given of scratch-hardness testing, with particular ref. to its employment on Co–Cr–W facing on valves. The advantages of this process for evaluating hot hardness and sub-zero hardness are outlined.—W. D. B.

***Hooke's Law: Is It a Rigid Law?** Adrien Jaquerod (*Rev. Mét.*, 1951, 48, (2), 85–90).—J. has already shown (*J. Suisse Horlogerie*, 1946, 418, *et al.*) that Hooke's law is never obeyed in torsional oscillations and that the divergencies may be serious. The period always varies with the amplitude and, unexpectedly, this variation is particularly rapid with small amplitudes. The experiments were carried out on a large number of metals and alloys, using a vibrograph having a pendulum wire 60–70 cm. long and 0·2–1·5 mm. in dia. The periods were recorded by means of a photoelectric cell and a beam of light reflected by a mirror on the torsion pendulum. J. extended these experiments to bending oscillations with steel and brass, and confirmed the former observations: the oscillations are never isochronous; their periods in all the cases studied diminish with the amplitude, and this diminution is particularly rapid with small amplitudes. The specimens were thin and relatively long tongues of the metal, carrying an Al point, vibrating across a spark gap, the circuit being connected elect. through a transformer. Curves showing the relationship between $(T/T_0 - 1)10^5$ and the amplitude were drawn and extrapolated to zero, T being the period and T_0 the period for a nul amplitude. Extrapolation is somewhat uncertain, but this is immaterial for medium and large amplitudes. In the case of steel, annealing at successively higher temp. raises the value of the $(T/T_0 - 1)$ curve correspondingly. The oscillations themselves caused cold working, as was indicated by the shape of the curves and as had been found in the torsion experiments. Damping is affected by annealing in the same way as the period. If the decrement is calculated as a function of the amplitude, a curve is nearly always obtained whose angular coeff. becomes smaller as the amplitude diminishes; the decrement is virtually const. for small amplitudes, and this is precisely the region in which the divergencies from Hooke's law vary most rapidly. In the case of many alloys there is no longer any parallelism between the curve of isochronism and that of decrement. There appears to be no relation between cause and effect; all that can be said is that, in certain cases the discrepancies and the internal friction vary in a like manner. No explanation has yet been put forward to account for these facts.—J. H. W.

Testing the Wear-Resistance of Surfaces. H. Meincke (*Metalloberfläche*, 1950, [A], 4, (10), 145–151).—A review, dealing with: (1) the construction and use of several types of German wear-testing machines which are based on the principle of measuring wear as a function of the loss in weight of the specimen caused by the abrasive action of a loaded hard-metal slider traversing its surface; and (2) the results obtained from tests carried out with such machines in relation to: (*a*) the load on, the speed of, and the time or distance traversed by the slider; (*b*) the structure of the surface to be tested, e.g. type and size of crystals, nature of any surface films; and (*c*) the hardness of the specimen. Although no relationship can be found between wear-resistance and Brinell hardness, the former can often be correlated with scratch hardness. It is concluded that wear tests are best carried out at loads of 0·5–5 kg./cm.^{2}, velocities of $\sim$1·45 m./sec., and distances of 2000–10,000 m. 9 ref.—E. N.

***A Mechanical Wear Test Using Fission Fragments.** David H. Frisch and Jules S. Levin (*Amer. Soc. Test. Mat. Bull.*, 1951, (172), 64–65).—Radioactivity was induced into the surface of stainless-steel cylinders by exposure to fission fragments, produced by bombarding with neutrons a thin foil contg. a little U^{235}. The activity of the surfaces was measured over a period of 105 hr., and then one of the surfaces was abraded and the activity after abrasion compared with that of the unabraded control specimen. This process was repeated several times. From the results it is concluded that the method is adaptable to measuring wear depths from a few millionths to a few ten-thousandths of an in. of dense materials and up to one-thousandth of an in. of lighter materials such as plastics. Wear may be measured at times from min. to months after irradiation of the surface. —P. T. G.

A Method of Determining the Thickness of Silver Deposited on Glass. Lloyd S. Nelson (*J. Chem. Education*, 1951, **28**, (2), 103).—The method described by Gardner and Case (*U.S. Nat. Bur. Stand. Circ.*, 1931, (389)) for determining the thickness of Ag deposits on glass by observation of the coloured rings formed on placing a crystal of I on or close to the surface is outlined. The relationship between the number of rings, formed by conversion of the Ag to AgI, and the Ag-film thickness is tabulated.—J. R.

Measuring the Thickness of Thin Coatings with Radiation Back Scattering. Eric Clarke, J. R. Carlin, and W. E. Barbour, Jr. (*Elect. Eng.*, 1951, **70**, (1), 35–37).—In β gauges for the detn. of sheet thickness, the normal arrangements of placing the radioactive source on one side of the material and the detecting head on the other in effect limits the gauge to estn. based on mass/unit weight; measurement of coating thickness on a base material is thus not practicable with this type of instrument. Provided that the atomic numbers of the coating and base materials are different, however, coating thickness can be determined non-destructively by back scattering of β particles originating from a shielded source located in the same assembly as the detector. Principles and theory of the equipment are discussed and illustrated.—J. R.

The Defective Part Stays on Receiving Dock Due to New Testing Plan at Byron-Jackson Plant. —— (*Western Metals*, 1950, **8**, (7), 31–32).—A short description of the appn. of the Dy-Chek inspection process to gas-turbine parts. Cracks or flaws are detected by a red dye after appn. of a white "developer." The red lines or spots emerge with excellent clarity against the white background. No indication is given of the compn. of the two liquids.—T. G.

Symposium on Non-Destructive Methods of Testing, Held at Constance, 8 Sept. 1950. —— (*Metall*, 1950, **4**, (21/22), 469–470).—A short review of the contributions on: (1) radiographic, elect.-conductivity, inductive, and supersonic methods of crack detection; (2) the radiographic method of stress detn.; and (3) recent advances in the various types of induction instruments developed by Förster.—E. N.

Symposium on the Role of Non-Destructive Testing in the Economics of Production. (——). See col. 93.

RADIOLOGY

X-Ray Examination of Small Parts. —— (*J. Metals*, 1950, **188**, (12), 1443).—Thin-walled tubes, failures, and problems involving instruments, tools, and fine mechanisms can be successfully analysed by radiographic methods by using the proper techniques.—E. N.

Modern Non-Destructive Methods for Detecting Faults in Castings. — Bastien (*Usine nouvelle*, 1951, **7**, (6), 21).—A brief survey of modern methods for the X-ray radiographic examination of metals, with particular ref. to the examination of welds and castings for faults.—H. A. H.

An Evaluation of Radiography, with Particular Emphasis on Detection Methods. Carlton H. Hastings (*Amer. Soc. Test. Mat. Bull.*, 1951, (172), 66–70).—Past trends of development, the present status, and future objectives of the radiographic method of non-destructive testing are discussed. It is considered that improvements are desirable in the quality of radiographs and in the speed with which inspection data are obtained. It is suggested that techniques employing television or other electronic principles show promise for attaining these ends, and could eventually lead to automatic or semi-automatic inspection and control of products.—P. T. G.

Progress in Gamma Radiography. —— (*Atomics*, 1951, **2**, (4), 112–118).—Supplies of Co^{60}, Ta^{182}, and Ir^{192} are now available, which are suitable for the non-destructive inspection of castings and welds. Co^{60} and Ta^{182} are suitable for the penetration of steel 2–6 in. thick, and Ir^{192} for steel of thickness 0·5–2·5 in., and for short-focus work with pipes, &c., and the radiography of small metallic objects. Supply details and an exposure-times chart are given. A new X-ray film, 2 to 3 times as fast as films hitherto supplied is now available. New apparatus for the storage of active materials, for exposure of sources for radiography, and for radiation monitoring are briefly described.—J. S. G. T.

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14 — TEMPERATURE MEASUREMENT AND CONTROL

A General Survey of Automatic Temperature Control. —II. J. Bernot and P. Lambert (*J. Four Elect.*, 1951, **60**, (2), 48–50).—Cf. *ibid.*, 1950, **59**, (6), 147; *Met. Abs.*, 1950–51, **18**, 628.—H. A. H.

Pyrometry in the Non-Ferrous Foundry. [—I.–II.] Robert A. Colton (*Foundry*, 1951, **79**, (2), 98–100, 243–245; (3), 88–89, 234–237).—[I.—] C. stresses the importance of measuring the temp. at which non-ferrous castings are poured and discusses the effects of pouring temp. on the soundness, properties, and surface finish of castings. [II.—] The different types of pyrometer commercially available are described and their use discussed. Methods of protecting the thermocouple tip are considered.—R. W. R.

Metal-Decorating Ovens Require Precise Temperature Control. —— (*Indust. Heating*, 1950, **17**, (3), 481–482, 484, 486, 488).—A description of continuous-feed patented metal-decorating ovens, using a zone type of construction. The controls for gas-fired ovens, temp. control for the separate zones, and combustion safeguards are discussed. —D. M. L.

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15 — FOUNDRY PRACTICE AND APPLIANCES

Aluminium Alloy Castings: A Review of British Achievement. Frank Hudson (*Trans. Amer. Found. Soc.*, 1950, **58**, 20–36; discussion, 37; and (abridged) *Found. Trade J.*, 1950, **89**, (1773), 201–205; (1775/76), 257–262).—Official Exchange Paper from the Institute of British Foundrymen to the American Foundrymen's Society. The current melting, moulding, prodn., and inspection practices in the manufacture of Al alloy castings are reviewed. 15 ref.—V. K.

***Solidification of [Duralumin] Ingots.** B. H. Alexander (*Trans. Amer. Found. Soc.*, 1950, **58**, 270–273; discussion, 273–276).—The progress of solidification of Duralumin ingots in a 6 × 6-in. mould cavity with 2·25-in. wall thickness, and in a 2·4 × 2·4-in. mould cavity with 0·7-in.-thick cast-iron moulds was studied by the slushing-out method. A straight-line relationship between $\sqrt{\text{time}}$ and ingot thickness was found to be obeyed for increasing pouring temp., which shift the line to the right. Slight deviations from this law were observed towards the end of solidification. 15 ref. —V. K.

***Segregation of Iron and Manganese in Some Aluminium Casting Alloys.** (Glaisher). See col. 16.

***Precision Casting Aluminium in Moist Investment Moulds.** H. Rosenthal and S. Lipson (*Trans. Amer. Found. Soc.*, 1950, **58**, 460–464; discussion, 464).—Cf. *Amer. Foundryman*, 1950, **17**, (4), 122; *Met. Abs.*, 1949–50, **17**, 848.—V. K.

Technique of Producing Pressure-Cast Aluminium Matchplates. Kurt A. Miericke (*Trans. Amer. Found. Soc.*, 1950, **58**, 198–201).—The technique of casting matchplate patterns in Al alloys is described. The gypsum plaster used for the mould contains 75–80 parts of gypsum plaster and 20–25 parts of fibrous talc, by weight. The alloy is forced into the mould cavity under a pressure of 5 lb./in.^{2}.—V. K.

Develops New-Type [Aluminium Alloy] Matchplate Frame. Franz Schumacher (*Foundry*, 1951, **79**, (2), 194, 196).—S. describes a new type of cast Al alloy matchplate frame, with which changes of the matchplates may be made in 20 sec. The frame can accommodate one to four different patterns. —R. W. R.

An [Aluminium] Alloy Designed for Pattern Shops. S. Zuckor (*Trans. Amer. Found. Soc.*, 1950, **58**, 718–719; discussion, 719).—The advantages of the Al-base alloy contg. Cu 4, Si 4–6, and Mg 0·2% for the prodn. of patterns are briefly outlined.—V. K.

Metal Specifications for the Brass and Bronze Foundry. James G. Dick (*Trans. Amer. Found. Soc.*, 1950, **58**, 208–214; discussion, 214–216).—Tentative specifications for ingotting brasses and bronzes for foundry melting purposes are outlined. The items include: size of the melt/ingot heat, general characteristics of the ingot, and methods of marking, loading, sampling, and testing.—V. K.

***Effect of Superheating and Casting Temperatures on Physical Properties and Solidification Characteristics of Tin Bronzes.** Bernard N. Ames and Noah A. Kahn (*Trans. Amer. Found. Soc.*, 1950, **58**, 229–258; discussion, 258–260).—The standard, U.S. Navy, horizontal, cast-to-size, fully-fed test-bar was used to study the effect of melting practice, tapping and pouring temp., and deoxidation with 0·01% P on the mech. properties, micro- and macro-structure, *d*, and pressure-tightness of 88 : 8 : 4 gun-metal. Comprehensive tables, graphs, and macro- and micro-graphs are given to show these relationships. Pouring temp. was found to be the most important variable necessary to control in test-bar casting practice. High pouring temp. gives a better indication of metal quality than low pouring temp. Small addn. of P are generally beneficial. 24 ref.—V. K.

***Melt Quality and Fracture Characteristics of 85 : 5 : 5 : 5 Red Brass.** F. M. Baker, C. Upthegrove, and F. B. Rote (*Trans. Amer. Found. Soc.*, 1950, **58**, 122–131; discussion, 131–132).—Coloured fracture photographs of test specimens obtained from melts of different quality in 85 : 5 : 5 : 5 bronze are given. Cf. also *Amer. Foundryman*, 1950, **18**, (4), 30; *Met. Abs.*, 1950–51, **18**, 553.—V. K.

Equipment for Degassing Magnesium Alloy Melts. Alex J. Juroff (*Trans. Amer. Found. Soc.*, 1950, **58**, 715–716; discussion, 717).—A simple unit for degassing Mg alloys with Cl is described. The gas is passed directly from a cylinder via a rubber connector tube and a C tube immersed in the metal at a rate of 0·04 lb. Cl/lb. metal, for 7 min., at temp. between 1330° and 1360° F. (721° and 738° C.).—V. K.

New [Magnesium] Alloy Castings for [U.S.] Air Force Result in Major Expansion. F. L. Church (*Modern Metals*, 1951, **7**, (1), 31–34).—Describes the planned expansion in an American foundry as a result of the commercial exploitation of the post-war Mg–Zr–rare-earth alloys. The founding of a large compressor case is quoted as an example of the prodn. methods used.—R. J.

***Casting of Magnesium–Rare-Earth–Zirconium Alloys in Sand Moulds.** K. E. Nelson and F. P. Strieter (*Trans. Amer. Found. Soc.*, 1950, **58**, 400–407; discussion, 407–410).—Mg-base alloy, EK30, contg. 2–4% rare-earth metals and 0·1–0·4% Zr, has been found to possess mech. (static) and creep-resistant properties equal to those of other alloys in this class, with the advantage of simplified foundry handling. Contamination with Al, $<$0·1%, results in grain coarsening. 12 ref.—V. K.

***Development of High-Strength Magnesium Casting Alloy ZK61.** (Meier and Martinson). See col. 21.

Magnesium Foundry Practice in Canada. M. W. Martinson and J. W. Meier (*Trans. Amer. Found. Soc.*, 1950, **58**, 729–736; discussion, 736–738).—Prodn. statistics, moulding, melting, and casting practice of Mg-base alloys in Canada are reviewed. 17 ref.—V. K.

***An Investigation of Melting and Casting Procedures for High-Purity Nickel.** Douglas W. Grobecker (*Trans. Amer. Found. Soc.*, 1950, **58**, 720–727; discussion, 727–728).—The practice recommended for melting and casting pure Ni includes: use of sillimanite-lined induction furnaces, oxidizing the melt by addn. of 0·5% NiO, boiling with 0·2% C addn. around the superheating temp., deoxidizing with 0·05% Mn, 0·1% Al, and 0·1% Mg, and pouring at the lowest pouring temp. consistent with avoiding cold shuts, into green, skin- or oven-dried sand moulds. 12 ref.—V. K.

Metal Melting: Application of Thermodynamic Principles to Melting Non-Ferrous Metals. Robert I. Moore (*Trans. Amer. Found. Soc.*, 1950, **58**, 465–469; discussion, 469–470).—Elementary thermodynamic principles are applied to the problem of economic melting of metals.—V. K.

***Principles of Gating Design: Factors Influencing Molten Steel Flow from Finger-Gating Systems.** W. H. Johnson, W. O. Baker, and W. S. Pellini (*Trans. Amer. Found. Soc.*, 1950, **58**, 661–668).—Cf. *Amer. Foundryman*, 1950, **17**, (4), 106; *Met. Abs.*, 1949–50, **17**, 852.—V. K.

***A Study of the Principles of Gating.** K. Grube and L. W. Eastwood (*Trans. Amer. Found. Soc.*, 1950, **58**, 76–99; discussion, 100–107).—The flow of water and Wood's metal in transparent (Lucite) moulds was studied by means of cinephotography. Different designs of pouring basins, down-runners, wells, runner bars, and ingates were related to the appearance of drossing (turbulent flow) or gas aspiration into the mould. Completely quiescent and non-aspirating mould gating systems cannot be obtained for practical foundry purposes, but they can be approached by observing certain rules in the design of the gating system: (*a*) a runner well helps in dissipating the kinetic energy at the entrance to the runner bar; (*b*) this energy is further reduced by doubling or trebling the total area of runner bars to the area of down-runner at the exit; (*c*) total area of ingates should be equal to or slightly larger than the area of the runner bars between the down-runner and the first ingate; (*d*) the area of the runner bar should be progressively reduced after the first ingate by the amount of the preceding ingate area; and (*e*) the runner bar should be in the drag and the ingates in the cope part of the mould. By applying the above rules a complete filling of the gating system is possible at the beginning of pouring. This rule is of great importance in casting dross-sensitive alloys.—V. K.

Methods of Gating Sand Moulds. Pat Dwyer (*Foundry*, 1951, **79**, (1), 124, 126).—D. gives a short account of the factors to be considered when selecting the gating method to be employed with a casting and gives diagrams of a number of commonly used gates.—R. W. R.

Casting Tools for Moulding Non-Metallic Materials. Thomas A. Dickinson (*Foundry*, 1951, **79**, (3), 190–192).—Procedures for casting the tools used for moulding rubber, plastic, and similar articles are described.—R. W. R.

Dimensional Tolerances in Casting. Gabriel Joly (*Usine nouvelle*, 1951, **7**, (3), 25).—The French railway and marine specifications for the dimensional tolerances of metal castings are considered with a view to suggesting a basis for their unification. This is not regarded as possible, and a new system is detailed.—H. A. H.

Definition and Classification of Foundry Defects of Metallurgical Origin. Albert Portevin (*Fonderie*, 1951, (62), 2345–2357; discussion, 2357–2358).—P. gives a broad classification of foundry defects of a met. nature from chem. and phys. causes, considers controlling the liability of alloys to suffer from defects and the characteristics of defects and their causes, and gives examples of different explanations of the same defect. In discussion, *M. Ballay* briefly described the work being done in France in identifying foundry defects, investigating their causes, and seeking their remedies.—J. H. W.

Developments in the Parlanti Casting Process. —— (*Machinery* (*Lond.*), 1951, 78, (1990), 3–8).—This is a gravity die-casting process, using dies made of anodized Al alloy (L 33). By varying the thickness of the mould walls, their chilling effect can be altered and the progress of solidification placed under control. When casting ferrous materials, cores can be made from Al and may be hollow or so proportioned as to melt out. Vacuum castings are made by enclosing the mould in an evacuated container sealed by an unanodized Al plate. The metal is poured into a runner box clamped to the top of the plate; the plate melts after a short time and allows the metal to flow into the die. Examples of castings made by the process in Al alloys, cast Fe, and steels are described and illustrated.—J. C. C.

Production of Complex Precision Die-Castings Speeded with Less Rejects by Western Plant. Joseph L. Haver (*Western Metals*, 1950, 8, (11), 21–23).—Describes the improvement of die-castings by use of new and modified die-casting machines.—T. G.

The Control of Porosity and Shrinkage Voids : Sound Die-Castings. E. N. Field (*Metal Ind.*, 1951, 78, (17), 323–326).—F. discusses the factors necessary for the prodn. of sound die-castings, such as the use of low injection temp., the maintenance of a high average die temp., correct positioning of the gate, the disposition of the runners, and the provision of a sufficiently high injection pressure. By due attention to these points the incidence of porosity, shrinkage voids, and surface imperfections is reduced to negligible proportions. —J. H. W.

Forming Processes for Metals. [III.—] Die-Casting. W. M. Halliday (*Mass Prod.*, 1951, 27, (2), 52–63).—Cf. *ibid.*, 1950, 26, (11), 54; *Met. Abs.*, 1950–51, 18, 559. The advantages and limitations of the pressure and gravity die-casting processes are discussed and compared.—R. W. R.

Economic Moulds for Pressure Casting. Maurice Billing (*Fonderie*, 1951, (62), 2368–2373).—Continuing a discussion on the equipment used in pressure die-casting (*ibid.*, 1950, (53), 2059; *Met. Abs.*, 1950–51, 18, 128), B. describes moulds suitable for the process.—J. H. W.

Mechanized Handling Reduces Die-Casting Production Costs. E. F. Ross (*Steel*, 1950, 126, (15), 82–84).—A modern foundry utilizing conveyors and other handling devices is described, for prodn. of Al and Zn alloy die-castings.—E. J.

Non-Ferrous Investment Casting. Hiram Brown (*Proc. Inst. Brit. Found.*, 1950, 43, A87–A103; and *Trans. Amer. Found. Soc.*, 1950, 58, 38–51).—See *Met. Abs.*, 1950–51, 18, 554.—V. K.

Glass Solves Delicate Coring Problems [in Investment Castings]. D. D. Malcomb (*Iron Age*, 1951, 167, (15), 107–109).—Thin fragile cores necessary for the prodn. of Al radar tube-feed antenna units by investment casting are reinforced with Pyrex glass tubes, thereby enabling the cores to withstand the turbulence and shock of the metal's entering the cavity.—J. H. W.

Design for Economy Demands Consideration of Quantities [Met by Investment and Die-Casting]. W. T. Minech (*Machine Design*, 1951, 23, (3), 139–141).—M. discusses the problem of designing components to allow for smooth prodn. The items discussed include purchased parts and materials cost, tooling and labour costs, and the relation between total number to be produced and the prodn. costs. Examples are given showing how the adoption of investment and die-casting techniques results in very great savings in total prodn. costs.—D. M. L.

Frozen-Mercury Patterns Add Flexibility to Precision Casting [Mercast Process]. A. H. Allen (*Steel*, 1951, 128, (12), 66–69).—Cf. *Met. Abs.*, 1950–51, 18, 678. Describes the use of frozen Hg as a pattern material instead of the conventional wax, for the precision casting of intricate shapes in a variety of metals and alloys. A number of advantages are claimed, and the article is illustrated.—E. J.

Pattern Materials and Production in Precision Investment Casting. E. I. Valyi (*Trans. Amer. Found. Soc.*, 1950, 58, 429–438; discussion, 438).—The properties of different types of pattern equipment used in precision investment founding are reviewed.—V. K.

Use of Portable Air Tools in Foundry Cleaning Rooms. A. G. Ringer (*Trans. Amer. Found. Soc.*, 1950, 58, 510–513; discussion, 514).—The working characteristics of pneumatic chipping hammers and grinders for cleaning castings are briefly described.—V. K.

Bandsawing in Foundries. Geo. H. Sheppard (*Trans. Amer. Found. Soc.*, 1950, 58, 621–625).—The working characteristics of different types of bandsaws used for fettling ferrous and non-ferrous castings are described.—V. K.

Foundry Grinding Practices. L. I. Smith (*Grits and Grinds*, 1950, 41, (10), 1–7).—S. describes the grinding plant and technique available for the fettling of castings.—R. W. R.

Foundry Cleaning-Room Abrasive Operations. Boyd H. Work (*Trans. Amer. Found. Soc.*, 1950, 58, 685–691).—Characteristics and appn. of different types of grinding machines for foundry purposes are described.—V. K.

Material Transport in the Cleaning Room. N. L. Smith and R. J. Wolf (*Trans. Amer. Found. Soc.*, 1950, 58, 550–556).—Mech. devices for use in foundry fettling transport operations are described.—V. K.

New Tool Removes Casting Defects Quickly. Robert H. Herrmann (*Foundry*, 1951, 79, (3), 164, 166–167).—H. describes the use of an O-jet elect. arc torch for gouging out defects in castings.—R. W. R.

Cold-Formed Flexible "Precision" Patterns and Core Boxes. R. B. Wagner and J. E. Wiss (*Trans. Amer. Found. Soc.*, 1950, 58, 675–679).—The manufacture of accurate patterns and core boxes in gypsum plaster and in thermoplastic polyvinyl chloride is described. 18 ref.—V. K.

Production Patterns and the Matchplate. Robert F. Dalton (*Trans. Amer. Found. Soc.*, 1950, 58, 699–704).—Different types of patterns are compared, with special ref. to the prodn. of pressure-cast matchplates. 18 ref.—V. K.

Pattern Engineering and Foundry Practice. A. Pirrie (*Canad. Metals*, 1950, 13, (3), 24–27, 49).—The need for co-ordinating the design and prodn. of castings is discussed, and serious errors in pattern design are emphasized. —W. A. M. P.

***A Study of Factors Affecting Moulding-Sand Density, Shrinkage, Expansion, and Workability.** R. P. Schauss, R. F. Baley, and E. E. Woodliff (*Trans. Amer. Found. Soc.*, 1950, 58, 217–225; discussion, 225–228).—Tables showing ramming *d* and hot expansion–contraction characteristics of foundry sands as a function of clay type and seacoal addn. are given. The lowest hot shrinkage at 2500° F. (1370° C.) is obtained with the lowest clay content. The addn. of seacoal (2–7%) reduces appreciably the shrinkage of sands of high clay contents.—V. K.

***A Study of the Effects of Various Binders and Additives on Hot Strength of Moulding Sands.** Robert E. Morey and Carl G. Ackerlind (*Trans. Amer. Found. Soc.*, 1950, 58, 411–421; discussion, 421–422).—The effects of moisture content, type of clay, and organic and inorganic addn. on the hot compression strength of moulding sands was determined over the temp. range 200°–2500° F. (95°–1370° C.). Clay binders produce relatively high hot strength which is further raised with an increasing moisture content. The presence of an organic bond, together with clay, increases the hot strength in the range 400°–800° F. (200°–425° C.); at still higher temp. the effect is dependent on the nature of the organic bond. Si flour or Fe oxide increases the hot strength, while wood flour cause its lowering. 59 ref.—V. K.

Flowability of Moulding Sands. William H. Moore (*Trans. Amer. Found. Soc.*, 1950, 58, 650–658; discussion, 658–660).—It is suggested that the flowability, *F*, of a foundry sand be defined as the combined effect of the ventability, V, and the bondability, B, so that $F = (V + B)/2$. On the basis of this definition, the flowability of different types and conditions of sand was obtained by the measurement of permeability and green compression strength.—V. K.

***Effect of Moisture on Core-Sand Mixtures.** O. Jay Myers (*Trans. Amer. Found. Soc.*, 1950, 58, 557–562; discussion, 562–563).—Increasing amounts of H_2O (0–9%) are related to the properties of core sands contg. 1% core oil and 1% cereal binder added separately or together. Moisture reduces the permeability of the sand. The optimum green compression strength is obtained with 0·5–1% moisture content with cereal or with combined binders, and it continuously increases with the oil binder. Tensile strength in the baked condition reaches a max. at 1% moisture with oil binder, 5% moisture with cereal binder, and 3% moisture with combined binders. —V. K.

Treatment of Bond Clays for Foundry Sand. A. E. Pavlish (*Trans. Amer. Found. Soc.*, 1950, 58, 492–501; discussion, 501–502).—The base-exchange reaction of a number of types of clay was studied with both inorganic and organic reagents. Bonding properties of clays can be widely modified in this way; tables showing the changes in permeability, and green and dry compression strength are given.—V. K.

Controlled Sand Produces Quality Castings. Bradley H. Booth (*Trans. Amer. Found. Soc.*, 1950, 58, 52–62).—Basic properties of foundry sands and their relation to moulding behaviour, sand selection, and foundry control testing are discussed.—V. K.

Reproducibility of Foundry-Sand Tests. B. H. Booth, P. C. Rosenthal, and H. W. Dietert (*Trans. Amer. Found. Soc.*, 1950, 58, 611–619; discussion, 619–620).—A statistical analysis was made of data obtained in a standard procedure sand testing of the same sample by 10 different laboratories. The scatter of results is fairly wide and can be reduced only by careful maintenance of the equipment, compliance with the detailed testing instructions, and special training of the laboratory personnel.—V. K.

Standard Data for Bench Core Making. H. R. Williams (*Trans. Amer. Found. Soc.*, 1950, 58, 503–509; discussion, 509).—Time operation studies of core-making process are analysed.—V. K.

Phenolic Resin Core Binders: Effects of Core Composition on Physical Properties of Cores. J. E. McMillan and J. A. Wickett (*Trans. Amer. Found. Soc.*, 1950, 58, 439–447; discussion, 447–449).—The combined effects of cereal, solid phenolic resin, and water on the standard properties of core sands are presented in the form of tables and graphs. The results show that the correlation which exists may eventually allow a general theory of core bonding to be obtained, but considerably more work remains to be done.—V. K.

Electronic Core Baking. Boyd Hopkins (*Foundry*, 1951, 79, (3), 154, 212).—H. describes the use of H.F. core-baking equipment in an American foundry.—R. W. R.

Modern Core Ovens. Charles H. Barnett (*Indust. Heating*, 1950, 17, (7), 1231–1236).—The great importance of correct core prepn. is stressed, and it is shown that the elimination of variables is essential. The prepn. of sand cores is outlined, and their behaviour in the mould during the entry of the liquid is described. Inefficient baking methods will not give consistently good cores. The factors affecting the quality of cores (temp., baking time, the oven atmosphere, and the gas circulation in the oven), are discussed in detail. The use of recirculating heating systems in core ovens is now being extended as automatic firing and temp. control then becomes simpler. B. discusses the use of fuel oils, coal, electricity, and gas for the firing of core ovens.—D. M. L.

Instruments Keep Accurate Records [in the Foundry]. Maurice F. Degley (*Foundry*, 1951, 79, (3), 108–111).—Stresses the advantages accruing from the use of recording instruments for measuring temp., recording the operation of sand millers, moulding lines, &c.—R. W. R.

Pyrometry in the Non-Ferrous Foundry. [—I.–II]. (Colton). See col. 57.

Modernization Trends in the Small Foundry. Lester B. Knight (*Canad. Metals*, 1950, 13, (2), 26–29).—K. shows how economics can be effected in small foundries by modernizing plant layout, materials handling, prodn. equipment, and methods. Case histories of three foundries—two grey Fe foundries and a brass foundry—are given to illustrate what can be done with modest capital expenditures.—W. A. M. P.

The [French] Foundry Industry. —— (*Usine nouvelle*, **1951**, (Numéro spécial), 21, 23, 24).—A general survey of the French foundry industry (ferrous and non-ferrous), with special ref. to its development since 1945 and comparison with conditions existing in America.—H. A. H.

Formation of Technical Management Groups for French Foundry Industry. M. Jean Lain (*Trans. Amer. Found. Soc.*, 1950, 58, 739–741).—The activities of the École Supérieure de Fonderie (French Foundry School) are outlined.—V. K.

Spanish Foundries. F. R. Morral (*Foundry*, 1951, 79, (3), 138, 141).—A short account of operations at one or two Spanish foundries.—R. W. R.

Foundry Apprentice Training at Caterpillar Tractor Co. F. W. Shipley and B. L. Bevis (*Trans. Amer. Found. Soc.*, 1950, 58, 515–524).—The syllabus for foundry apprentice training is given.—V. K.

The Engineering Student and Summer Foundry Work. N. J. Stickney (*Trans. Amer. Found. Soc.*, 1950, 58, 680–684). —A scheme for preparing vacation courses for engineering students in foundries is outlined.—V. K.

[Labour-]Fatigue Data Summary: Report No. 1. M. E. Annich (*Trans. Amer. Found. Soc.*, 1950, 58, 169–172; discussion, 172–173).—Labour-fatigue allowances in different foundry operations are statistically analysed on the basis of a questionnaire of the American Foundrymen's Society.—V. K.

Statistical Quality Control [in the Foundry]. Harry A. Schwartz (*Foundry*, 1951, 79, (1), 94–95, 223–224).—S. briefly discusses the advantages accruing from the use of quality-control methods in foundries.—R. W. R.

Wage Incentive Administration. Charles T. Hassell (*Trans. Amer. Found. Soc.*, 1950, 58, 604–609; discussion, 610).—The appn. of time study in foundries as a means of improving the output is discussed.—V. K.

Management Views Costs [in the Foundry]. J. A. Wagner (*Trans. Amer. Found. Soc.*, 1950, 58, 710–713; discussion, 713–714).—A brief review of foundry costing.—V. K.

Casting of Brass and Bronze: Some Practical Aspects of Brass and Bronze Casting in America, 1900–1950. (Hull). See col. 94.

Kurs Odlewnictwa. [Foundry Course]. (Gierdziejewski). See col. 95.

16 — SECONDARY METALS: SCRAP, RESIDUES, &c.

Secondary-Aluminium Mission to the U.S.A. —— (*Metallurgia*, 1951, 43, (256), 67–68).—A summary of the main findings of an O.E.E.C. mission to the U.S.A. to study secondary Al plants.—F. M. L.

[Current] Practice in the American Secondary [Aluminium] Metal Industry. —— (*Metall*, 1951, 4, (23/24), 530–531).—A brief summary of the report of a European working party which made a 7 weeks' tour of the U.S.A. Deals with melting, refining, casting, rolling, and uses of the secondary metal.—E. N.

The Removal of Zinc from Galvanized Iron and Steel Scrap. Edmund R. Thews and Martin Stromeyer (*Metalloberfläche*, 1950, [A], 4, (8), 121–123; (11), 169–172; and (abridged) *Electroplating*, 1951, 4, (4), 133–134).—After discussing the economic and met. reasons for the removal of Zn from galvanized Fe and steel scrap, T. and S. briefly describe the various chem., electrochem., and thermal methods employed.—E. N.

17 — FURNACES, FUELS, AND REFRACTORY MATERIALS

World's Largest Induction Furnace for Aluminium Melting at Vanadium Corp. Plant. Manuel Tama (*Indust. Heating*, 1950, **17**, (7), 1150–1152, 1154, 1156, 1272).—The 1000-kW. Al melting furnace at the Vanadium Corp. is described. The furnace melts assorted Al scrap at the rate of 5000 lb./hr. In conjunction with the melting equipment there is a 30-kW. holding furnace arranged for electromagnetic pumping of the liq. metal via a heated launder to the ingot moulds. The main furnace, of the Ajax–Scomet drum induction type, is powered by four twin-coil inductor units inserted into the lower part of the drum. The special arrangements for temp. control of the equipment are described in detail.—D. M. L.

Induction Furnaces [for Aluminium Melting] Cut Foundry Melting Costs. Vernon Swan (*Modern Metals*, 1951, **7**, (1), 38–40).—An analysis of the advantages and disadvantages of I.F. melting as compared with more conventional melting methods. S. concludes that where a continuous supply of metal is required, i.e. in die-casting, permanent-mould casting, or continuous pouring in sand moulds on a conveyor, I.F. melting pays. The induction furnace can reduce melting costs by up to 40%, in addn. to improving working conditions and casting quality and making possible better utilization of floor space.—R. J.

Flash-Annealing Furnace for Light-Alloy Sheet. —— (*Light Metals*, 1950, **13**, (146), 146–147).—A description is given of a conveyor-type, air-circulation furnace heated by elect. resistance elements. Single sheets up to 5 ft. 6 in. wide and circles down to 5 in. in dia. can be annealed at the rate of 1 ton/hr. The conveyors consist of tensioned ropes and are designed to avoid damaging the work. Two independent, balanced, air-circulating systems are employed, and the total weight of air circulated is substantially greater than usual, with the result that heat transfer is very rapid. —J. E. T.

New Furnaces Heat Copper Billets and Anneal Strip Continuously. —— (*Indust. Heating*, 1950, **17**, (6), 998, 1000).—A brief description of the continuous billet heater and continuous roller-hearth bright-annealing furnace installed at the plant of Revere Copper and Brass, Inc. The former has a capacity of 18,000 lb./hr. and is designed for heating billets for extrusion. The furnace is gas fired, using over- and underfiring burners. The bright-annealing furnace is designed for the treatment of Cu tubing in coils and straight lengths. The conveyor has a 20 : 1 speed range. Heating is by gas-fired radiant tubes located above and below the working hearth. The atmosphere generator is specially equipped for removal of S from the furnace gas.—D. M. L.

Large Melting and Annealing Furnaces Featured in Scovill's New Strip Mill. —— (*Indust. Heating*, 1950, **17**, (3), 404–408, 410, 412, 414–415, 418, 420–422, 544, 546–548, 550, 552).—Cf. *Met. Abs.*, 1949–50, **17**, 979; 1950–51, **18**, 129, 307, 632. A detailed illustrated account of the continuous-casting and strip-mill equipment installed by the Scovill Manufacturing Co. at Waterbury, Conn. Special three-inductor-unit drum-type furnaces of 1000 kW. capacity are arranged to melt 10,000 lb./hr. each. The construction of these brass-melting furnaces is described, and details are given of the arrangements for feeding the holding furnaces and the continuous-casting machines. This machine, of the Junghans–Rossi type, has a capacity of several million pounds of metal/week. Each slab emerging from the machine weighs over 2000 lb. The slabs are broken down in a cold-rolling mill and the bars passed into a continuous annealing furnace having a capacity of 37,000 lb./hr. The furnace length has been reduced by the use of a new pre-heat chamber. Automatic temp. control is maintained, but there is no full atmosphere control. Another large annealing furnace is described in some detail. Bars leaving the annealing furnaces go on a roller table, from where they are machined and fed to the cold run-down mill. Details are given of the coil-annealing equipment, the high-speed continuous spray-pickling machines, the cold-finishing mill, and the finishing operations.—D. M. L.

The Copper Reverberatory Furnace. W. H. Dennis (*Metallurgia*, 1951, **43**, (257), 107–111).—Factors affecting the design, construction, and operation of modern high-prodn. Cu reverberatory furnaces are reviewed.—F. M. L.

Better Magnesium Melting Method Cuts Die-Casting Rejects to 5%. Gerald C. Robechaud (*Western Metals*, 1951, **9**, (2), 26–27).—Deals with the preheating and melting of Mg ingots. Preheating is by gas, and melting is carried out in an 80-kW., 1000-lb. capacity elect.-resistance type breakdown furnace controlled within temp. limits of ±2° F. (±1° C.).—T. G.

Electric-Resistance Furnaces as a Means for Improving Quality [in Heat-Treatment Operations]. H. Friedrich (*Metall*, 1951, **5**, (1), 14–16).—A brief review of the construction, appn., and advantages of modern types of low-, medium-, and high-temp. elect. resistance furnaces, with and without controlled atmospheres, for the heat-treatment—annealing, hardening, &c.—of various metals and alloys.—E. N.

Vacuum Electric Furnaces for the [Bright] Annealing of Blanks. H. Kalpers (*Metall*, 1950, **4**, (21/22), 471).—A brief description of the design and capabilities of such furnaces, which may be vertical or horizontal. They operate, normally, at a vacuum of 10^{-1} mm. Hg, although small furnaces for special purposes can be worked at 10^{-2} mm. Hg when appropriate pumps are installed; temp. up to 1100° C. are attainable. Such furnaces are eminently suitable not only for the bright annealing of large numbers of small stampings, but also for tubes, rods, wire, strip, and foil of Cu and its alloys, Ni and its alloys, and stainless steels; they have the advantage that subsequent pickling is unnecessary, thus dispensing with the space, equipment, materials, and labour required for this operation. They can also be adapted to the sintering of hard metals and the degassing of metals for use in the radio industry.—E. N.

A New-Design Vacuum-Melting Furnace. J. M. Taub and D. T. Doll (*U.S. Atomic Energy Commission Unclassified Rep.*, 1950, (**LAMS-1130**), 24 pp.).—The furnace described can be used for melting and casting 5–50-lb. charges under a pressure of $\sim 1\ \mu$ Hg. It consists of two stainless-steel water-cooled shells, the upper contg. the melting and pouring crucible, and the lower section contg. the mould, which is either of graphite or a graphite–ceramic combination. The lower section is heated by a W resistance heater, and the upper section either by resistance or induction heating. —B. W. M.

Reheating Furnaces with a Rotating Floor. A. Bailly (*Usine nouvelle*, **1951**, (Numéro spécial), 39–40).—The advantages of reheating furnaces with a rotating floor are discussed generally, and a short description is included of a French furnace capable of dealing with 20 tons/hr. of steel billets or ingots.—H. A. H.

Western Metal-Processing Methods Stepped Up by Gas-Equipment Improvements. Ralph W. Marshall (*Western Metals*, 1950, 8, (6), 26–29).—A survey of the use of natural gas in California for various appn. in foundry practice, heat-treatment, and joining.—T. G.

Control of Refractory Materials. C. E. Sumpter (*J. Metals*, 1950, 188, (12), 1442–1443).—A brief review which, although dealing with refractories used in the construction and maintenance of elect. furnaces for steel, contains interesting details of the methods of checking and controlling the quality of such materials, and of the correlation of laboratory and service data.—E. N.

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18 — HEAT-TREATMENT

Brass Strip: Annealed and Cleaned Continuously. —— (*Steel*, 1951, **128**, (13), 76–77).—Describes, with a flow diagram, a plant for the continuous annealing and pickling of brass strip of thickness from 0·003 to 0·040 in.—E. J.

Electric-Resistance Furnaces as a Means for Improving Quality [in Heat-Treatment Operations]. (Friedrich). See col. 66.

Vacuum Electric Furnaces for the [Bright] Annealing of Blanks. (Kalpers). See col. 66.

Thin Metal Parts Successfully Heated by Induction. Joseph F. Libsch and Anthony Vescuso (*Wire and Wire Products*, 1950, **25**, (6), 498–499, 519–521).—A technique for the induction heating of small wires and metal parts using a standard commercial convertor operating at 500 kc./s. or less and providing flexibility for the heat-treatment of both small and large sections is described.—W. A. M. P.

Safety Procedures for Protection Against Flammable Atmosphere Hazards of Special-Atmosphere Furnaces.—I.–II. J. B. Smith (*Indust. Heating*, 1950, **17**, (4), 592–596, 598, 600. 743; (5), 818, 820, 822, 824).—[I.—] S. discusses the explosion hazards present in heat-treatment furnaces where the special atmospheres are composed of inflammable gases. The various types of atmospheres and the manufacturing processes involved are discussed, and details are given of the procedures for avoiding explosive mixtures in furnaces of different forms. [II.—] In cases where explosive mixtures cannot always be avoided special arrangements must be made to prevent explosions, and safeguards must be arranged against interruption of the flow of special atmosphere gases. Details of these products are given.

—D. M. L.

19 — WORKING

Aluminium Rolling in Chile. —— (*Light Metals*, 1950, **13**, (145), 78–79).—Data are given on a projected new rolling mill in Chile. The basic equipment is enumerated, and the estimated prodn. costs are given.—J. E. T.

Processing of Aluminium Alloys with Tungsten-Carbide Drawing Dies. C. E. Hughes and E. T. Miller (*Wire and Wire Products*, 1950, **25**, (10), 885, 886, 902–904).—A brief review, dealing with the cold working and drawing of Al alloys.—W. A. M. P.

Aluminium Aircraft Parts Hot-Formed Successfully. Gilbert C. Close (*Steel*, 1951, **128**, (10), 84–86).—The use of oil-immersion and of elect.-resistance methods of heating is described for the successful hot pressing of Al alloy aircraft components. By careful temp. control at 300° F. (150° C.) overageing is prevented.—E. J.

Aluminium Barrel Production. Hugh J. Jarman (*Mech. Handling*, 1951, **38**, (4), 137–139).—Prodn. of Al beer barrels at the Benson Manufacturing Co., Kansas City, is described and illustrated. The barrels are deep-drawn in two sections, which, after annealing, &c., are circumferentially joined by means of automatically revolving welding jigs. Output is at the rate of 1200 15½-gal. barrels/day.—J. R.

[Manufacture of] Aluminium Beer Barrels. —— (*Metal Ind.*, 1951, **78**, (17), 348–349).—Describes the manufacture of Al beer barrels in Canada.—J. H. W.

Aluminium Wire: Its Fabrication and Uses. C. J. Beneke (*Wire and Wire Products*, 1950, **25**, (7), 571–573).—A brief review dealing with the selection of Al alloys for various wire appn., fabrication methods and equipment, annealing and soln. heat-treatment, and indust. uses.—W. A. M. P.

***Investigations to Determine the Suitability of Light Alloys and of Alloys of the Zamak Type for Purposes of Screw-Cutting [on the Lathe].** Raymond Farrenq (*Métaux et Ind.*, 1950, (5), 107–129).—The characteristics of 9 alloys, viz. 4 types of Al alloy (normal Duralumin DN, leaded Duralumin DE, Duralinox H5, and Almasilium A.S.G.), 3 Zamak-type alloys, an Al alloy contg. Cu, Mg, Si, Pb, and Zn, and an Al–Mg alloy, when turned with screw-threads on a lathe were studied. Threads of 3 different pitches, viz. 0·1, 0·2, and 0·3 mm., were cut on the alloys to a depth of 2 mm. The features studied included: (1) the form of the turnings produced, (2) the wear of the cutting tools, piercing, and screw-cutting, and (3) pressure and temp. attained in cutting. The cutting angles of the tools used were 55°, 62°, 72°, and 82°; the front cutting edge was set back at 8°, and the back edge was set at 0°, 10°, 20°, or 27°. The principal results were as follows: (i) the best form of turnings are, in general, obtained by using a wide cutting angle, viz. 82°, cutting at high speed and cutting screws of greatest pitch. (ii) Duralumins, either with or without Pb, give the best turnings, when the alloys are either quenched or work-hardened without quenching. (iii) The alloys, so far as concerns the form of the turnings, are arranged in the following order of merit: brass, leaded Duralumin (treated), 2 of the Zamak-type alloys, untreated leaded Duralumin, treated Pb-free Duralumin. (iv) Treatment of Pb-free Duralumin has no effect upon tool wear, and tool wear, in the case of all the Duralumins of this class, is identical. (v) Leaded Duralumins on the other hand, behave quite differently as regards tool wear, the quenched and cold-worked but unquenched Duralumins causing 5 or 6 times the wear produced by the drawn alloys. This effect is not shown by the other alloys used, although one Zamak-type alloy caused slightly less tool wear when used as-extruded compared with the result with the drawn alloy. (vi) A study of the mode of detachment of the turnings shows that all the alloys behave in much the same way, except in the case of the untreated Duralumins, which exhibit a markedly inferior behaviour in this respect. (vii) The general process of cutting is sensibly the same for all the alloys. (viii) The effect of cutting angle, i.e. the decrease of pressure accompanying a decrease of cutting angle is more marked in the case of Duralumin in the drawn condition than for treated Duralumins. (ix) Normal Duralumins and Almasilium give higher cutting pressures than Duralumins contg. 1·5–2% Pb, while the Zamak-type alloys show the lowest cutting pressures.—J. S. G. T.

Conditions for Working Some Light Alloys Using Cutting Tools. — Schweyckart (*Usine nouvelle*, 1951, **7**, (13), 25).—The free-cutting properties of the most important light alloys are partially determined by Si content. Ranges of cutting rates using high-speed steel and carbide tools are given for alloys contg.: (*a*) <1% Si, (*b*) 9–13% Si, and (*c*) >18% Si. The influence of phys. and mech. properties of the alloys on cutting conditions is also discussed.—H. A. H.

High-Speed Milling of Light Metals. —— (*Modern Metals*, 1951, **7**, (1), 36).—As the high machinability of light metals is rarely fully exploited because of the speed limitations of commercial machines, new milling machines have been designed which operate at 3600 or 7200 r.p.m.—R. J.

Correct Machining of Copper-Base Alloys. Malcolm Buell and J. J. McGuiness (*Steel*, 1950, **126**, (16), 78–81).—The classification of Cu alloys into groups when considering machinability is described, and data are given for turning, milling, drilling, boring, tapping, &c., as well as coolants recommended.—E. J.

The Production and Applications of Copper-Base Alloy Wire. W. D. France, W. L. Wells, and W. E. Johnson (*Wire and Wire Products*, 1950, **25**, (10), 836–839, 936).—Some specific appn. of certain typical alloys are discussed and wire prodn. is briefly reviewed.—W. A. M. P.

The Production and Applications of "Doublé" [Gold Plate]. W. Radecker (*Metall*, 1950, **4**, (23/24), 516–517).—A brief review.—E. N.

[Lead] Accumulator Plate Manufacture. —— (*Metal Ind.*, 1951, **78**, (18), 364–365).—Briefly describes and illustrates the manufacture by casting and rolling of a pure-Pb accumulator plate weighing nearly 1 cwt.—J. H. W.

Electrical Wire and Cable Insulation by the Extrusion Method. B. H. Davis (*Wire and Wire Products*, 1950, **25**, (8), 641–644).—An improved method of manufacture of insulated elect. conductors by the use of the screw-type extruding machine is described.—W. A. M. P.

***Influence of Structural Inhomogeneities on the Hot-Working Properties of Magnesium–Aluminium Alloys.** (Lott). See col. 20.

***An Analysis of Some Roll-Pass Designs for Round Sections.** D. A. Winton (*J. Iron Steel Inst.*, 1951, **167**, (1), 31–41).—W. compares a number of different schedules for reducing (steel) billets to round sections. It is shown that use of the sq.–diamond and oval–sq. sequences necessitates fewer passes than other sequences, for the reduction of a given billet to any particular size of round. It is shown that the area A_p of any pass p is given by $A_p = A_0(1 - r)^p$ where A_0 is the billet area and r is the fractional reduction in area/pass and equals $\{1 - (A_p/A_{p-1})\}$. The design of the leader and finishing passes is discussed.—R. W. R.

How To Spin Metals. John W. Lengbridge (*Machinist* (*Eur. Edn.*), 1951, **95**, (16), 569–584).—L. describes a method of estimating costs of spun versus drawn material and gives basic information on the design of spun products. He outlines the actual technique of spinning low-alloy steels, Cu-base, Ni-base, and stainless alloys, the tools and equipment required, and the various prodn. methods used.—J. H. W.

Why Not Spin It? L. F. Spencer (*Iron Age*, 1951, **167**, (9), 105–109).—Spinning is an ideal forming operation for small prodn. quantities or for larger prodn. runs of parts that would require complex press tooling. Many metals and alloys, both ferrous and non-ferrous, can be spun, and the method is often used as an operation subsequent to deep drawing. The selection of a suitable lubricant is important. 9 ref. —J. H. W.

15,000,000 Stampings per Grind on Carbide Dies. John E. Spofford, Jr. (*Steel*, 1951, **128**, (10), 80–82).—Describes the use of cemented WC for inserts and punches of dies for high-speed presses.—E. J.

The Problem of Die Wear, with Special Reference to the Performance of Sintered Carbide Dies. I.—General Considerations. J. G. Wistreich (*Wire Ind.*, 1950, **17**, (203), 889–892, 895–896; discussion, 896–897, 899).—W. discusses the economic importance of the die problem and gives an account of some investigations into the nature of wear.—W. A. M. P.

Carbide Wire-Drawing Dies—Bearing vs. No Bearing. J. O. Tompkins (*Wire and Wire Products*, 1950, **25**, (7), 576–578).—The performance of six "no-bearing" and six "bearing" dies was compared by using the dies on continuous drawing machines as finishing dies during the regular prodn. of 0·038-in.-dia. tyre bead wire (0·74–0·80% C, 0·55–0·65% Mn). Bakelite castings of the die contours were made before and after drawing 500,000 ft. of wire through each die, and mech. tests were carried out on all wire produced by the two types of die. The "no-bearing" dies started to form bearings on one side of the die, and the scratch lines were deeper on the "no-bearing" die surface. These dies produced more out-of-round and scratched wire than the "bearing" dies, but mech. test results were virtually the same.—W. A. M. P.

The Development of Wire-Rod Production. I.—From the 17th Century to the Present Day. II.—Evolution of the Rod Mill. III.—U.K. and U.S.A. Rod Mills Compared. Future Trends. Leslie Thomas (*Wire Ind.*, 1950, **17**, (201), 736–739, 743; (202), 823, 825, 827; (203), 903–905, 907–908).—[I.—] An historical review. [II.—] A review of the development of the rod mill. [III.—] Statistics are given for U.K. and U.S.A. wire-rod prodn.—W. A. M. P.

Material Handling at the Rod and Wire Mills of the Western Electric Company. J. E. Wiltrakis and A. A. Bureau (*Wire and Wire Products*, 1950, **25**, (10), 870–875, 878, 879, 882, 883).—A description of the material-handling methods of the Western Electric Co. at their various wire plants.—W. A. M. P.

Application and Use of Lubricant Coatings. Floyd M. Hauger (*Wire and Wire Products*, 1950, **25**, (7), 579–581, 610).—H. reviews the appn. and uses of lubricant coatings and describes a new coating compound which eliminates the use of both lime and lubricants in the die box.—W. A. M. P.

20 — CLEANING AND FINISHING

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

Aluminium Bright-Dip Process Gives Low-Cost Reflectivity. F. H. Hesch and C. A. Rosellen (*Western Metals*, 1950, **8**, (12), 19–20).—Deals with the results obtained by a method of chem. polishing developed by Kaiser Aluminum and Chemical Corp. and termed Kaiser aluminium bright-dip. No indication of the compn. is given.—T. G.

High-Reflection Eloxal Polishing of [Aluminium] Mirrors. Otto Niedermeyer (*Metalloberfläche*, 1950, [B], **2**, (7), 102).—A brief but detailed outline of the process for producing a durable mirror finish on Al (99·5 and 99·99%) or its alloys with 0·5–1% Mg. It consists of chem. cleaning, preliminary buffing, electrolytic degreasing, followed by electrolytic polishing in an Eloxal-type bath for 10–20 min., cold H_2O rinsing, and a final hardening treatment in hot (90°–100° C.) distilled H_2O for 30 min.—E. N.

Photography and Printing on Aluminium. Karl Voss (*Australasian Plating and Finishing*, 1951, **1**, (2), 4–7).—The oxide layers produced on Al and its alloys by anodizing are finely porous, and may be impregnated with light-sensitive and other materials. For the prodn. of "Seofoto" prints the oxide layer must comply with requirements for sorption capacity, corrosion-resistance, hardness, transparency, and ductility. The Al may be anodized in oxalic, sulphuric, or chromic acid baths. In the "Seo" printing process selected areas are sealed by printing with a fatty, tough printing paste. The uncovered areas are dyed and the printing ink removed with solvent, so that a reproduction representing the reverse of a print is obtained.—T. A. H.

Metal Etching and Anodizing in Colour. R. E. Pettit (*Products Finishing*, 1951, **15**, (7), 14–23, 26–28).—The processes used by a company producing metal etchings, instrument dials, scales, identification plates, &c., are described in detail and illustrated. The H_2SO_4 anodizing method is used. —G. T. C.

The Surface Preparation of Aluminium for Paint Systems. —— (*Product Finishing* (*Lond.*), 1951, **4**, (1), 56–61).—Reproduced from a booklet of the same title issued by the Northern Aluminium Co., Ltd.—H. A. H.

Copper as a Finishing Material: Review of 1950. —— (*Indust. Finishing* (*Lond.*), 1951, **3**, (33), 664–667).—A review. The subjects considered include Cu plating, electroforming, the chem. colouring of Cu and brass, the corrosion of Cu water-pipes, and the use of Cu alloys in marine service. 21 ref.—H. A. H.

Blanching of Silver. K. Slark (*Metalloberfläche*, 1950, [B], **2**, (10), 155).—A brief note on blanching—in 10–15% H_2SO_4 at <85° C.—subsequent washing, drying, and final buffing procedures.—E. N.

Approved Proprietary Finishing Materials [for Aircraft]. —— (*Indust. Finishing* (*Lond.*), 1951, **3**, (33), 653–654, 656, 658, 660).—A list of some of the approved proprietary processes covered by specification D.T.D. 900C. The material, extent of approval, and manufacturer are tabulated, and the range includes cleaners for Al, paints and varnishes for metal surfaces, electrodeposition and chem. pretreatment processes for ferrous and non-ferrous metals.—H. A. H.

Metal Finishing Forges Ahead. Adolph Bregman (*Iron Age*, 1951, **167**, (8), 78–82).—B. describes the more important new processes and improvements to old techniques that have occurred during the past five years, and emphasizes the need of developing substitutes to take the place of materials in short supply. 17 ref.—J. H. W.

Refrigerator Finishing : Methods Used by Kelvinator Ltd. —— (*Product Finishing* (*Lond.*), 1951, **4**, (2), 58–67).—A description is given of the pretreatment and painting processes. A combined degreasing–phosphatizing soln. (Granodine 400) is used for processing steel and Zn-coated components before painting. Al parts are Pyluminized.—H. A. H.

The Production of Metal Cypher Plates for Watches. G. Meyer and W. Philippi (*Metalloberfläche*, 1950, [B], **2**, (9), 135–136).—A brief description, which includes details of the methods used to produce the various surface finishes and the designs.—E. N.

New Synthetic Resin Forms Surface Coating for Many Metals. —— (*Product Finishing* (*Lond.*), 1951, **4**, (2), 108, 110).—A short description of the use, properties, and advantages of " Araldite 985 " surface coating resin for light alloys, tinplate, steel, and cast Fe. Finishes prepared from this resin are stated to be particularly suitable for appn. to food cans, Al foil, collapsible metal tubes, and light alloy drums. —H. A. H.

Wire Finishes and Trade Requirements. C. R. Mehl (*Wire and Wire Products*, 1950, **25**, (8), 657–658, 685).—A short discussion of wire finishing and the use of borax compounds as compared with lime coating.—W. A. M. P.

Barrel Lacquering of Small Stampings. H. Weise (*Metalloberfläche*, 1950, [B], **2**, (9), 137–138).—A brief description of the plant, materials, and technique employed.—E. N.

Impurities [as a Cause of Defects] in Enamelling. Hans J. Karmaus (*Metalloberfläche*, 1950, [B], **2**, (9), 132–133).—Many faults—blisters, porosity, hair-cracks, &c.—found in enamel coatings on metal surfaces are often caused by the presence of impurities introduced into the enamel at some stage in its manufacture. K. discusses the likely sources of such contamination and the preventative measures necessary. —E. N.

The Treatment of Metal Surfaces Before Lacquering. —— (*Metalloberfläche*, 1950, [A], **4**, (7), 108–110).—Although this article deals with the treatment of steel sheets, it is of interest to the non-ferrous metallurgist as it emphasizes the importance not only of removing visible contamination such as oxide films, grease, and dirt, but also the invisible traces of residual acids, alkalies, salts, soaps, and other chemicals or their reaction products left behind by the cleaning, pickling, and degreasing soln.—E. N.

Cleaning and Peening with British Hard-Drawn Steel Pellets. —— (*Electroplating*, 1951, **4**, (2), 56).—The use and advantages of hard-drawn steel pellets for metal cleaning and peening are discussed. A particular advantage of pellets as compared with shot or grit abrasives is that the former can be manufactured in almost all ductile metals and alloys, e.g. stainless-steel pellets are employed for the specialized cleaning of bronze and brass castings, where Fe impacted in the surface might lead to objectionable stains during the finishing processes.—H. A. H.

Soak Cleaning Before Alkaline Electrocleaning. —— (*Plating*, 1951, **38**, (4), 377–378).—The mechanisms of soil removal are listed, and the function of individual components of cleaners are described. A table is presented which shows the effect of cleaner compn. on time of soil removal by soak cleaning.—G. T. C.

Etch Pre-Treatment Primers. E. E. Halls (*Machinery* (*Lond.*), 1951, **78**, (1992), 114–116).—These primers contain H_3PO_4, which etches the metal to improve adhesion, and are pigmented with $ZnCrO_4$ to provide corrosion-resistance. The results of humidity tests on cellulose-enamelled panels of Al, brass, Zn- and Cd-plated steel, treated with various primers, are tabulated.—J. C. C.

***Electrodeposits as Bases for Paints.** H. Silman and Dorothy Wernick (*J. Electrodepositors' Tech. Soc.*, 1950, **25**, 175–188).—Preliminary results are presented of a study to determine the behaviour of composite coatings of metals and paints. This has been made necessary by the increased use of pre-plated sheet and strip metals. The results are mainly confined to electrotinned and electrogalvanized steel. As these coatings are normally relatively thin, they need to be reinforced with a paint coating where any degree of severe service is required. The importance of correct prepn. of the plate before painting is emphasized. A comparison was made between the behaviour of paint coatings applied to electrotinned steel and to phosphatized steel. Half of each of the tinned specimens was given an oxidizing treatment, and the other half was untreated. Two thicknesses (0·00001 and 0·00005 in.) of Sn and two coats of a stoving alkyd paint were used. Salt-spray and outdoor exposure tests were made. The painted 0·00005-in. tinplate gave the best results. The painted 0·00001-in. tinplate was superior to phosphatized steel in salt spray, but not markedly so in atmospheric exposure. The oxidizing treatment of the tinplate has had little effect on the performance so far. The fact that the time factor between plating and painting is not important with tinplate as it is with phosphatizing is an important practical advantage of the former. Sets of Zn-plated steel panels with 0·0003 in. Zn from a cyanide soln. both untreated and dichromate-treated (Cronak process) were painted with one coat of three different paints. Salt-spray and humidity-chamber tests were carried out. The results showed that the Cronak coating is satisfactory for air drying and low-temp. stoving paints and that with a high-temp. stoving paint there was little difference between the treated and untreated Zn. The Cronak treatment, however, improves adhesion. Tests were also carried out which indicated that the protection provided by the Cronak treatment decreased as the temp. to which the treated specimens were exposed was increased, especially at temp. in excess of 100° C. This effect is probably less marked when the treated metal is coated with paint. Two-coat paint systems were applied to untreated and phosphatized steel, Zn-plated steel, phosphatized Zn-plated steel, and passivated Zn-plated steel, and salt-spray and outdoor-exposure tests were carried out. Phosphatizing, in each case, gave better results than were obtained on untreated specimens, both in salt spray and out of doors. Passivated Zn-plated steel gave better results in salt spray than phosphatized Zn-plated steel. Nitrocellulose paints have given poor adhesion on Zn-plate exposed out of doors, in some cases. Paint adhesion to Ni- or Cr-plated steel and brass can be obtained by roughening with emery-coated calico wheel and treatment in hot dil. soln. of H_3PO_4. 6 ref.—G. T. C.

Pre-Finishing and Contour Shaping. —— (*Product Finishing* (*Lond.*), 1951, **4**, (1), 75–82).—A description of American techniques for pre-finishing and contour-shaping metal. The former, which is performed using a special arrangement of 2 to 12 or more coated abrasive belt grinding and polishing machines arranged in series, is claimed to reduce or entirely eliminate expensive hand-polishing operations which are otherwise necessary after forming operations. In contour grinding, which is designed to grind and finish the contours of formed articles, a specially contoured back-up roll or wheel is employed which allows the abrasive belt to conform to the contours of the component being handled. Although the examples quoted all refer to ferrous metals, both methods are stated to be applicable to non-ferrous materials.—H. A. H.

Chemical Polishing [of Metals]. —— (*Metalloberfläche*, 1950, [B], **2**, (8), 118).—A brief description of the three acid process—mixtures of nitric, phosphoric, and acetic acids in various proportions—developed at the Battelle Memorial Institute.—E. N.

Liquid Polishing Compounds. Frank Spicer (*Product Finishing* (*Lond.*), 1951, **4**, (3), 68–69).—A note on American practice concerning the use of liq. polishing compounds for metal surfaces. They are claimed to be particularly successful on Zn-base die-castings and Al.—H. A. H.

Brush-Backed Polishing : A New Method of Metal Surface Preparation. —— (*Electroplating*, 1951, **4**, (4), 117–119).—A short description of the use and advantages of the brush-backed, coated abrasive-loaded polishing mop in metal finishing.—H. A. H.

Automatic Polishing Machines: Advanced American Designs Used on Small Articles Such as Costume Jewellery and Cigarette Lighters and Particularly on Cutlery. —— (*Electroplating*, 1951, **4**, (5), 149–151).—Brief descriptions of some of the most modern American polishing machines for small articles.—H. A. H.

Selection and Application of Cleaning-Room Equipment. Stanley Krzeszewski (*Trans. Amer. Found. Soc.*, 1950, **58**, 450–455).—Cf. *Amer. Foundryman*, 1950, **17**, (4), 96; *Met. Abs.*, 1949–50, **17**, 856.—V. K.

The Neutralization of Waste Acid Liquors. A. Gordet (*Galvano*, 1951, (**170**), 17–20).—A review of methods of neutralizing spent pickle soln.—H. A. H.

***Contribution to the Study of Inhibitors in Pickling.** G. Batta, L. Scheepers, and L. Bousmanne (*Rev. Mét.*, 1951, **48**, (2), 105–114; discussion, 114–115).—The action of a number of organic inhibitors was studied experimentally, chiefly in the pickling of steel, but also of Zn, Ni, and Al. Theoretical conditions developed from this work indicate that there is a certain relation between the molecular structure and the inhibiting action of the organic compounds used as addn. in pickling. In the case of the non-ferrous metals there is no sp. qualitative factor depending on the nature of the metal, but Zn is somewhat subnormal in its electrochem. behaviour. Certain of the organic inhibitors are sensitive to the *p*H; some, used in low concentrations, sometimes cause local corrosion. The pickling of Cr–Ni steels by *aqua regia* and by HNO_3–HF remains to be investigated.—J. H. W.

Establishing Finishing-Operation Rates with Elemental Time Standards. John S. Kelly (*Metal Finishing*, 1951, **49**, (3), 71–74).—Results obtained from the appn. of this method in a large finishing department are described.—G. T. C.

Works Organization in the Metal-Finishing Trade. A. W. Wallbank (*J. Electrodepositors' Tech. Soc.*, 1950, **25**, 211–220).—Presidental Address. Several aspects of this problem are very briefly considered.—G. T. C.

21 — JOINING

Bonding of Wood to Aluminium and Its Alloys. —— (*Machinery Lloyd*, 1950, **22**, (22), 97).—Report on a proprietary glue fibre that is supplied in sheet form and is placed between the Al and wood sheets to be stuck together. The Al alloy is degreased and etched in accordance with D.T.D. 915A. When dry, the metal should be bonded as soon as possible after etching-treatment, using a suitable pressure and temp. in the range 140°–150° C. The joints are resistant to dry and wet heat and to mould.—H. Pl.

Redux Bonding of Aircraft Structures. C. J. Moss (*J. Roy. Aeronaut. Soc.*, 1950, **54**, (478), 640–646; discussion, 646–650).—Features of Redux joints in light alloy sheets are described. Results of comparable panel tests of riveted and Redux-jointed panels under static loading showed the latter to be the stronger. Details of tests under cyclic loading of hollow structures subject to fluctuating internal pressure and of repeated-loading tests of a complete wing are given. Techniques of applying the glueing process in aircraft construction are detailed and considered in relation to design.—H. S.

Aircraft Riveting. H. Giddings (*J. Roy. Aeronaut. Soc.*, 1950, **54**, (480), 753–778).—In this comprehensive paper, G. discusses the functions of riveted joints, rivet materials, rivet manufacture, rivet types and setting techniques, the strength of riveted joints, and joint finish. Results of static and fatigue tests of riveted joints are given. The repeated loading problem associated with pressurized fuselages is discussed in relation to the ultimate strength of the structure. —H. S.

Induction Heating Successfully Solders Aluminium. C. H. Yetman (*Iron Age*, 1951, **167**, (11), 108–109).—Induction heating provides the uniform heat diffusion and accurate temp. control required for soldering Al, and eliminates oxidation problems. A high-strength bond at a rate of 320 pieces/hr. was obtained.—J. H. W.

Sealing Solder. G. S. Phipps (*Bell Lab. Record*, 1950, **28**, (7), 295–297).—P. describes a method of making soldered joints gas-tight by the appn. superficially of a film of sealing solder of low m.p.—D. M. D.

History of Soldering Technique. Erich Lüder (*Technik*, 1950, **5**, (3), 144–149).—L. describes the soldering of Au in Babylon, Egypt, and Troy, soft and hard soldering in ancient Rome and in medieval times, and the development of hot flames by means of bellows, blow lamps, blow pipes, blow torches, and welding torches. The development of soldering techniques in connection with the introduction of new materials such as tinplate, galvanized sheet, Zn sheet, &c., and of modern solders and fluxes is dealt with. L. concludes with the soldering of light metals.—H. K.

Production Brazing. Clinton E. Swift (*Weld. Eng.*, 1950, **35**, (4), 36–38).—The advantages of torch-brazing in the manufacture of metal furniture fabricated from sheet metal or thin-walled tubing is described and illustrated. Savings of 40–60% are reported.—W. D. B.

Assembly by Brazing. H. R. Brooker (*Metal Ind.*, 1951, **78**, (21), 426).—B. discusses British Standard No. 1723: 1951, dealing with brazing and tabulates the combinations of parent metal and filler metal for blow pipe (torch) brazing. —J. H. W.

Copper Furnace Brazing Saves Time, Labour, and Material in Formation of Assemblies. Mario R. Glick (*Western Metals*, 1950, **8**, (8), 22–23).—T. G.

Is Furnace Brazing Just as Good? Samuel Damon (*Machinist* (*Eur. Edn.*), 1951, **95**, (17), 616–618).—Experiments to determine how furnace brazing compared with other methods of brazing were carried out on tough-pitch Cu (99·9%), free-cutting brass (Cu 60–63, Pb 2·5–3·7%, balance Zn), and SAE 1111 steel, with the following alloys: (1) Cu 80, Ag 15, P 5%; (2) Cu 93, P 7% ("Phos-Copper"); (3) Cu wire (for the steel only); (4) Ag 50, Cu 15·5, Zn 16·5, Cd 18%; and (5) Ag 50, Cu 34, Zn 16%. The results showed that furnace brazing, especially with Phos-Copper, will produce a shear strength at least as high as that of other brazing methods.—J. H. W.

Brazed Joints for Mine Rails. —— (*Weld. Eng.*, 1950, **35**, (1), 32–33).—The advantages of brazed joints in mine rails are discussed, and the prepn. and procedure for making them is described and illustrated.—W. D. B.

Silver Brazing Eases Shotgun Welding. Kenneth Ireland (*Machinist* (*Eur. Edn.*), 1951, **95**, (12), 432–433).—By paying due attention to good fit and proper clearance, clean metal, proper fluxing, assembly and supporting, heating and flowing alloy and final cleaning, improved prodn. capacity and a pleasing appearance results from the use of a Ag–Cu–Cd–Zn brazing alloy for sporting guns.—J. H. W.

Silver Brazing with Induction Heat at Oneida, Ltd. Stewart M. Hill and Richard A. Gehr (*Indust. Heating*, 1950, **17**, (6), 972–974, 976).—An illustrated description of the prodn. of table knives. The hollow handles are Ag brazed to stainless-steel blades by means of an extremely small quantity of Ag solder. This replaced 12 g. of soft solder in the original assembly. The induction-heating equipment consists of 15-kW. generators fitted with output transformers. Heating time is reduced from 20 to 4 sec./cycle.—D. M. L.

[Koldweld] Cold Pressure Welding [of Aluminium]. —— (*Light Metal Age*, 1950, **8**, (1/2), 14–15).—A short note on the Koldweld process for welding Al at room temp. No operational details are given.—H. A. H.

Further Developments in the Koldweld [Cold-Pressure Welding] Process. —— (*Light Metal Age*, 1950, 8, (9/10), 10, 19).—Typical appn. for the Koldweld process for the cold pressure welding of non-ferrous metals (Al in particular) are described. No details of the process itself are given. —H. A. H.

The Welding of Aluminium and Its Alloys. E. Barron (*Machinery Lloyd*, 1950, **22**, (26), 75–78).—A general discussion on welding of Al and its alloys which includes notes on flux characteristics, degreasing, gas welding, and electric-arc welding.—H. Pl.

The Welding of Aluminium Alloys, Particularly for Structural Applications. W. K. B. Marshall (*Trans. Inst. Weld.*, 1950, **13**, (6), 178–185).—A survey of the present knowledge. M. deals with the properties of the weldable alloys and the basic metallurgy of welded joints in them. The welding processes are discussed in some detail, and comments are made on the influence of welding on design. The need for service data is emphasized. 12 ref. —W. D. B.

Electric Resistance Welding and Its Use in the Construction of Aircraft. E. C. Sykes (*J. Roy. Aeronaut. Soc.*, 1950, **54**, (472), 242–246).—S. describes and discusses spot and seam welding of sheet materials, with ref. to the general features of the types of welding machine used, methods of cleaning the material before welding, control of the process, and inspection of the welds.—H. S.

Strato-Ships Spot Welded. Howard E. Jackson (*Weld. Eng.*, 1950, **35**, (11), 17–19).—A description of the welding methods and plant used by the Boeing Airplane Co.—W. D. B.

Field Welding of Aluminium Tanks. Perry C. Arnold (*Metal Ind.*, 1951, **78**, (21), 427–429).—Read before the Twelfth Annual Ohio State Welding Conference. A. describes the erection and welding sequence developed for a new method of erecting large Al storage tanks.—J. H. W.

[Welding] Aluminium Pipe-Lines. W. B. Moore (*Weld. Eng.*, 1950, **35**, (7), 27–29).—Some details are given of the fabrication of Al-alloy pipe-lines, and the advantages, as compared with steel, are outlined.—W. D. B.

[Aluminium] Brackets Welded Inert Arc, Metal Arc. George M. Worden (*Weld. Eng.*, 1950, **35**, (2), 33).—Comparison in cost between metal-arc and He-shielded arc welding of Al and Al alloys showed a marked saving in favour of the inert-gas-shielded method.—W. D. B.

The Atomic-Hydrogen Welding of Light Alloys. Charles Guinard (*Rev. Aluminium*, 1950, (**171**), 429–435).—G. describes the principle of the method, the necessary equipment, the prepn. of the work for welding, the welding technique, and suitable appn. of the process. The mech. properties of joints in various Al alloys are listed.—J. E. T.

[Inert-Arc] Welded Aluminium Trailer. R. E. Ferris (*Weld. Eng.*, 1950, **35**, (1), 37).—The fabrication of Al trailer roofs by inert-arc welding is described.—W. D. B.

Argon-Arc Welding of Aluminium. E. Zurbrügg (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (12), 352–356).—A brief review dealing with the types of alloys most suitable for such welding, the effects of subsequent ageing procedures on the hardness of the seam, and some commercial appn. of the process.—E. N.

Argon-Arc Welding [of Aluminium]. —— (*Light Metal Age*, 1950, 8, (5/6), 14–15).—A short account of the A-arc welding of Al and its alloys, using the Linde FSM-1 A–metal-arc welding machine.—H. A. H.

***The Welding of Thick Aluminium Alloy Plates by the Argon-Arc Process.** J. E. Chard and N. MacDonald (*Weld. Research*, 1950, **4**, (4), 71r–78r).—A-arc welding equipment has been designed for welding currents up to 600 amp., and may be used for the welding of longitudinal and circumferential seams. Welds in A.W.10 plate, up to $\frac{5}{8}$ in. thick, using filler metal of similar compn., gave max. stress and P.S. equal to those of the parent plate, with acceptable ductility. —W. D. B.

***The A.C. Argon-Arc Process for Welding Aluminium: The Oscillographic Analysis of the Application of a Commercial High-Frequency Spark-Injector Unit.** L. H. Orton, J. C. Needham, and J. H. Cole (*Weld. Research*, 1950, **4**, (3), 47r–68r).—The nature of the output of a commercial H.F. spark-oscillator unit has been studied by the appn. of electromagnetic and cathode-ray oscillograph recording techniques. The output consists of a grouped series of sparks separated by dead periods; the spark groups should be correctly phased relative to the main welding current. The duration of the various phenomena was measured, the optimum relative phasing for successful welding was such that the middle of the spark group occurred just after the welding current reached zero. Correct phasing eliminates half-cycle rectification with a workpiece of Al–5% Mg alloy, change of the relative position causes variation in the time delays to arc re-ignition in half cycles of reverse polarity. The effects of using commercially pure Al and of surface condition are noted, and operational adjustments to the spark-injector unit are outlined. A simple switching method is described for obtaining optimum phasing of the inputs to the H.F. spark-injector unit and the welding transformer.—W. D. B.

Aluminium Window Production [by Flash Welding]. —— (*Welding*, 1950, **18**, (11), 464–471).—Some of the recent developments in the flash welding of Al alloy window-frames are described. A low kVA. demand for the machines employed results in a reduction of flashing losses, and a high rate of output can be obtained. In AW9 alloy (Mg 0·4/0·8, Si 0·3/0·6, and Fe 0·6%) welds at least equal in strength to the parent material are obtained. Illustrations show some of the dies, &c., employed.—W. D. B.

Improved Technique for Flash-Welding Aluminium [Alloy] Tubing Eliminates Interior Flash. Charles Bruno (*Steel*, 1950, **126**, (17), 79–80).—A general account of the strength of joints and of the use of flash welding. Details are given of percussive welding and its advantages and appn.—E. J.

†A Review of Selected Papers on the Flash and Butt Welding of Light Alloys. P. M. Teanby (*Weld. Research*, 1950, **4**, (1), 16r–20r).—A review of 8 papers on the flash and butt welding of light alloys, dealing with machines, timers, dies and clamping, prepn. for welding, &c. Hardness measurements indicate that the prodn. of high-strength welds in the hardened tempers requires careful control, the effects of over-ageing must be kept to a min., while the max. effect of work-hardening must be obtained. The final weld strength is dependent upon the opposing effects of these phenomena. The results are presented in tabular form.—W. D. B.

Slope Control Aids A.C. Welding of Aluminium. Charles Bruno and G. W. Birdsall (*Iron Age*, 1951, **167**, (14), 100–101).—A new electronic control makes possible good-quality resistance welds of Al with conventional single-phase 60-c./s. A.C. equipment. Al pick-up is thereby reduced and electrode life is increased 20–30 times.—J. H. W.

†A Review of Recently Published Information on the Spot Welding of Light Alloys. P. M. Teanby (*Weld. Research*, 1950, **4**, (5), 94r–102r).—A brief summary of the types of spot-welding machines now in use is followed by a review of information on such welding variables as current wave shape, striking velocity, electrode scuffing, variations in line voltage and pressure, electrode materials, bad fits, and surface prepn. Appendices deal with kinetic-energy storage, three-phase systems, and the measurement of surface resistance using the Kelvin double bridge. 31 ref.—W. D. B.

Design for Spot Welding Aluminium. Gilbert C. Close (*Light Metal Age*, 1950, 8, (1/2), 6–7, 32).—In order to obtain the max. benefits from spot welding as a fabrication method, certain sp. advantages and limitations must be taken into consideration at the design stage. These are discussed with particular ref. to the use of Al alloys, and tables are included to show the weldability of different alloy combinations (classified as excellent, fair, or not recommended), as well as recommended spot spacings for some of the alloys. The min. allowable shear loads which a spot weld may carry are also given for 16 Al alloys, together with the resp. min. edge distance.—H. A. H.

***The Application of the Resonance-Vibration Method to the Fatigue Testing of Spot-Welded Light Alloy Structures.** (Weck). See col. 55.

Gas Welding Aluminium Awnings. —— (*Weld. Eng.*, 1950, **35**, (4), 23).—The gas welding of $\frac{1}{16}$- and $\frac{1}{32}$-in. Al sheet in the manufacture of venetian-blind-type awnings is described. —W. D. B.

Copper and Copper Alloys as Fillers in Welding and Brazing. Benno Sixt (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (12), 336–340).—A review of the properties and appn., in oxy-acetylene welding and brazing, of filler rods of Cu, brass, and bronze, of various compn. —E. N.

Inert-Arc Welding of Everdur Beer Filter. A. C. Kracklauer (*Weld. Eng.*, 1950, **35**, (11), 26–28).—The welding of Everdur, using the inert-arc process and a filler rod of similar compn., is described and illustrated.—W. D. B.

Welding, Brazing, and Soldering Beryllium Copper.—I.–III. John T. Richards (*Weld. Eng.*, 1950, **35**, (6), 23–27; (7), 37–39, 56; (8), 35–37).—[I.—] Met. and mech. properties of Cu–2% Be alloys, recommended heat-treatment, pickling soln., &c., are described. Procedures for soft soldering are given; a 50 : 50 Sn–Pb alloy is most useful, but a number of others are also given, together with recommended fluxes. Suggested procedures for pre-tinning are described. [II.—] Ag brazing may be carried out at high or low temp.; in either case a post-ageing heat-treatment at 315° C. must be given. Fluxes and methods are discussed and some results tabulated. Recommended techniques for resistance welding and some data on the welding of dissimilar metals are outlined. [III.—] C-arc welding precautions and procedures for welding the alloy and for overlaying Cu or mild steel are dealt with. Metal-arc welding with Al bronze electrodes is frequently used. Inert-arc welding appears to be an ideal method, as it prevents oxidation of base metal and electrode. A.C. welders are preferred; some current requirements are given.—W. D. B.

Welding [High-Nickel Alloy] Jet Aircraft Engines. Frank G. Harkins (*Weld. Eng.*, 1950, **35**, (1), 18–21).—The welding of some high-Ni alloys by inert-arc, submerged-arc, and resistance-welding processes is briefly discussed. —W. D. B.

Some Extracted Notes on Hastelloy Fabrication and Welding. —— (*Weld. Research*, 1950, **4**, (4), 79r–80r).—The principal welding features of the Hastelloy (Ni-base high-strength) alloys are given.—W. D. B.

Tantalum Partly Replaces Niobium Content of Electrodes. R. D. Thomas, Jr. (*Iron Age*, 1951, **167**, (14), 109–111).—A new type 347 welding electrode, made by applying a Nb-bearing coating to a 20 : 10 Cr–Ni (type 308) core wire, conserves Nb and has about the same tensile and stress rupture properties, corrosion-resistance, and crack sensitivity. Owing to the shortage of Nb, a 40 : 20 : 40 Nb–Ta–Fe electrode has been developed. Since, however, Ta is also in short supply, the type 308 electrode can often be substituted for both the old and the new types. —J. H. W.

Salvage of Castings by Welding of Defects. G. E. Bellow (*Trans. Amer. Found. Soc.*, 1950, **58**, 669–674).—Equipment, materials, electrodes, and techniques used in the repair welding of ferrous and non-ferrous castings are briefly described.—V. K.

Automatic, Accurate, Flash-Butt Welding Process for the "Socket Joining" of Various Metals [and Alloys]. K. Ruppin (*Metall*, 1950, **4**, (23/24), 506–507).—A brief note on the appn. and advantages of automatic, flash-butt resistance welding for forming secure joints of the rod-and-socket type in similar or dissimilar metals or alloys. —E. N.

Assembly Time on Terne Plate Tanks Cut by Forty Per Cent. R. D. Wasserman (*Steel*, 1950, **127**, (5), 73, 75).—Describes a new welding technique which effects a saving when making Pb-coated steel tanks.—E. J.

Controlling Welding Distortion. C. G. Bainbridge (*Canad. Metals*, 1950, **13**, (2), 41–44).—B. discusses the main causes of welding distortion and suggests practical methods for its control. Heat input, size of weld, and influence of design are among the factors considered.—W. A. M. P.

Progress in Welding and Cutting: A Literature Review. J. Ruge (*Schweissen u. Schneiden*, 1951, **3**, (3), 91–96).—A brief review with 164 ref.—T. G.

Some Thoughts on Welding. C. Laville (*Usine nouvelle*, **1951**, (Numéro spécial), 33).—Suggestions are put forward to explain why welding has been adopted as a mass-prodn. tool by some French industries and not by others. It is recommended that more attention should be given to the possibilities of this method of joining metals.—H. A. H.

Cold Welding Gets Hot. —— (*Weld. Eng.*, 1950, **35**, (2), 17–19).—A summary of the cold-welding process in the form of an interview with an expert in this field. —W. D. B.

Autogeneous Pressure Welding. L. Wolff and O. Renner (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (12), 356–361).—A brief review dealing with the technique, the types of equipment, the mech. properties of the welds produced with various steels and Cu-base alloys, and the commercial appn. and economics of the process. 17 ref.—E. N.

The Physical Metallurgy of Gas–Metal Reactions in Fusion Welding. M. Hatherly (*Australasian Eng.*, **1950**, (Nov.), 81–88; discussion, 88–89).—The principles of gas–metal relationships are reviewed, and their appn. to the absorption of gases during fusion welding discussed. The effects of absorbed gases during solidification of the weld metal are considered, with particular ref. to the development of gas porosity in steel, Cu, and Al. Porosity is due to the absorption of gas, usually H, O, CO_2, or CO, in the molten weld metal and to the evolution and trapping of this gas during solidification. Some effects which the absorbed gases may have on the parent metal adjacent to the weld are mentioned also. 15 ref.—T. A. H.

Resistance Welding to Replace Soldering. Russell F. Smith (*Weld. Eng.*, 1950, **35**, (9), 20–22).—The relative advantages of the two processes are tabulated and some examples described and illustrated.—W. D. B.

Are You Up-To-Date on Resistance Welding? W. K. Boice (*Weld. Eng.*, 1950, **35**, (9), 38–41, 55).—An outline of some recent developments in resistance welding machines, e.g. frequency changes, metallic rectifiers, and slope controls. —W. D. B.

Design Considerations in Resistance Welding.—I.–II. Wallace A. Stanley (*Weld. Eng.*, 1950, **35**, (1), 34–36; (2), 22–26).—[I.—] Appearance, strength considerations, tolerances, limits, &c., of resistance welding are dealt with. [II.—] The weldability of a number of metals, including Al, Mg, brasses, bronzes, Zn-base die-castings, Pb, Cu, &c., is discussed, together with a tabular presentation of dissimilar metal combinations.—W. D. B.

Cabinets Reinforced by Carbon Arc. William C. Henzlik (*Weld. Eng.*, 1950, **35**, (5), 32–34).—The manufacture of steel kitchen furniture includes torch brazing with brass filler and C-arc welding with Cu–Si alloy filler metal. —W. D. B.

Light-Gauge Arc Welding. —— (*Welding*, 1950, **18**, (11), 473–480).—A description of the "Carbo-Flux Process", in which flux-coated filler rods are deposited by C-arc welding. The appn. of the process to the welding of Ni alloys, Cu, &c., are briefly described.—W. D. B.

Argon-Arc Welding. M. Brillié (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1950, **95**, (12), 348–352).—[In French]. B. describes recent work, carried out in France, dealing with the effects of impurities in the A, the dimensions and rate of feed of the W electrode, and the stability of the arc.—E. N.

Recent Developments in Fluxes for [Automatic Submerged-Arc] Welding. William M. Conn (*J. Metals*, 1951, **191**, (2), 98–99).—The most important properties of fluxes for use in arc welding are summarized, and details are given of the compn. and methods of prodn. of the three principal groups of fluxes which have been developed for submerged-arc welding and which are based on the use of raw (natural) clay, glass, and mullite. The flux consisting of raw clay and chemicals is used very little to-day. Advantages of the ceramic flux in comparison to the glass flux are: ceramic flux permits the use of higher voltages in welding; the linear footage of welding/lb. flux is increased because of its lower *d*; the compn. of the slag and its low viscosity in a molten condition permit the passage of large amounts of gas and vapour. 4 ref.—E. N.

Hard-Facing. F. W. Weigel (*Machine Design*, 1950, 22, (5), 126–128).—W. describes the appn. of hard-facing to the screw roll of a vegetable oil press. The roll is subjected to abrasive foreign matter in addn. to vegetable and animal acids. The materials used for hard-facing include Co-base alloys, composite alloy rods of Fe and Ni alloys or Fe-base alloy rods. Great care must be taken after hard-facing to avoid serious cracking of the coating from the base metal. —D. M. L.

Methods of Repairing Articles in Steel, Cast Iron, Copper and Its Alloys, and in Aluminium and Light Alloys, by Metal Spraying with an Oxy-Acetylene Pistol. M. Cauchetier (*Fonderie*, 1951, (63), 2385–2391).—C. describes various types of oxy-acetylene pistol and the methods of using them for repairing metal parts and explains the mechanism of the process.—J. H. W.

New [Colmonoy] Metal-Spraying Process. H. Reininger (*Metalloberfläche*, 1950, [A], 4, (7), 105–107).—A description of the Colmonoy metal-spraying process, and its appn. to the fabrication of machinery, and the repair of worn and broken parts. 9 ref.—E. N.

A Hundred Years of Progress and Development in Metal Joining Processes. F. A. Fox and A. J. Hipperson (*Sheet Metal Ind.*, 1951, 28, (289), 465–472).—F. and H. trace the developments in the art of joining metals since 1851. Developments are considered under headings such as resistance welding, gas welding, non-ferrous welding, brazing, soldering, &c. 38 ref.—R. J.

Welding Hazards: Industrial Health Bulletin. —— (*Canad. Metals*, 1950, 13, (1), 45).—Brief notes are given on the following health hazards associated with welding: (1) inhalation of toxic gases, fumes, and dusts; (2) exposure to radiant energy produced by the welding process; (3) elect. shock and burns; and (4) flying particles.—W. A. M. P.

22—INDUSTRIAL USES AND APPLICATIONS

A Review of the Properties and Uses of Super-Purity Aluminium. S. M. Lawrence (*Sheet Metal Ind.*, 1951, 28, (286), 141–145).—A review of the well-known properties of super-purity (99·99%) Al. Various appn. are enumerated. —R. J.

Demountable [Aluminium Alloy] Buildings. —— (*Light Metals*, 1950, 13, (146), 156–160).—An illustrated description is given of Al alloy demountable buildings built on the Briggs system. The whole assembly is boltless, with the exception of the holding-down bolts used when the buildings are erected on concrete foundations, and the framework is built up, not of individual members, but of a number of frames connected together with tie bars and sockets. —J. E. T.

The O.P.E.C. [Aluminium Alloy] House. Maurice Victor (*Rev. Aluminium*, 1950, (171), 405–413).—A description is given of a bungalow which is built with a light-alloy framework, Al roof sheets, interlocking Al alloy sidings, and internal plaster-board panelling. The Al alloys used are those contg. 3 and 5% Mg. Several details of construction are illustrated.—J. E. T.

Aluminium Air-Borne Prefabs: [U.S.] Air Matériel Command Tests a Prefabricated Structure of Aluminium Panels Designed to be Flown to Forward Bases. —— (*Light Metal Age*, 1950, 8, (1/2), 11).—H. A. H.

Aluminium Roofing: The Congress of the International Union of the Roofing, Plumbing, and Sanitation Industries. —— (*Rev. Aluminium*, 1950, (171), 417–420).—Summarizes certain resolutions made at the congress concerning the use of Al and Al alloy components. While the use of Al for water-pipes and plumbing fittings, with the exception of air ducting and gas flues, was rejected, a motion advocating its use for roofing was adopted. The necessary precautions are referred to in detailed reports.—J. E. T.

New Roofing Material [Aluminium]. —— (*Sheet Metal Ind.*, 1951, 28, (289), 462).—The new Al cork sandwich insulated panelling used for the Thames-side Restaurant at the [London] South Bank Exhibition is described.—R. J.

A Review of the History, Theory, and Uses of [Aluminium] Metal-Foil Insulation. H. Manley (*Sheet Metal Ind.*, 1951, 28, (286), 161–166).—Over recent years the use of metal foil for insulation has greatly increased, and M. traces the history of its use. Metal foils have three uses in that they insulate from: (*a*) heat and (*b*) cold, and they prevent condensation. Some of the problems of insulation are considered, and various methods of applying foil for insulation purposes are also discussed. Some well-known examples of successful Al insulation projects are briefly mentioned. M. concludes that the chief advantages of Al foil for insulation are lightness, cheapness, and ease of appn. 14 ref.—R. J.

Modern Aluminium Wall Plates Feature Alumilite Finishes. —— (*Modern Metals*, 1951, 6, (12), 34).—Appn. of Alumilited Al to switch plates is described.—R. J.

Roll-Up Aluminium Doors [Are] Easy to Operate; Save Space. —— (*Modern Metals*, 1951, 6, (12), 36).—The appn. of Al to the fabrication of roll-up doors for closing off kitchenettes proved ideal.—R. J.

Light Alloys at the Building Exhibition: Olympia, 1949. —— (*Light Metals*, 1950, 13, (144), 38–55).—A description is given of exhibits utilizing Al and Mg alloys which were shown at the 1949 Building Exhibition.—J. E. T.

Dramatic Application of Aluminium in British Architecture. —— (*Light Metal Age*, 1950, 8, (1/2), 12–13).—A note on the use of Al for the domed ceiling of the Royal Albert Hall, London, in railway coaches and bookstalls, &c.—H. A. H.

Fifty Years of Light Metal Progress in Architecture. —— (*Light Metal Age*, 1950, 8, (1/2), 8–10, 34).—A survey of the use of Al in American architecture over the past 50 years. —H. A. H.

Aluminium in Road Transport. —— (*Light Metals*, 1950, 13, (146), 160–165; (147), 186–191).—An account is given of the papers read at an Aluminium Development Association symposium on road transport appn. Goods and public-transport vehicles are discussed in some detail, and many examples of long service life and weight-saving are given. A paper on Al alloys for private cars deals with the selection of suitable alloys and concludes with an estimate that their use in place of steel in a 10-h.p. unitary-type car would reduce the weight by 300 lb. Papers on the properties of alloys suitable for use in this field and on the maintenance of light alloy vehicles are also summarized.—J. E. T.

Outlook for Aluminium in the Automobile Industry. C. P. Bedford (*Modern Metals*, 1951, 6, (12), 22–25).—A review of the potential market for Al in the U.S. automobile industry. Uses are sub-divided into: (*a*) items already in prodn., (*b*) items which appear feasible in the future. —R. J.

Problems of Change-Over from Steel to Aluminium in Fabricating Truck Bodies. —— (*Light Metal Age*, 1950, 8, (9/10), 8–9).—H. A. H.

Aluminium Alloys for Mining Equipment. J. C. Bailey (*Iron Coal Trades Rev.*, 1950, 161, (4314), 923–927; (4315), 977–980).—B. reviews the main characteristics of Al alloys from the standpoint of the mining engineer, and mentions various equipment which successfully incorporates the alloys in their construction.—E. J.

Light Alloys in Monotower Crane Construction. —— (*Engineer*, 1951, 191, (4969), 519).—Describes a crane fitted with a 160-ft. jib constructed mainly of Al alloy. —D. K. W.

How a Progressive Manufacturer Is Streamlining Aluminium Fabrication. Philip Brenton (*Light Metal Age*, 1950, **8**, (7/8), 8–9, 24–25).—An account of the manufacture of Al baby " strollers " by the A. E. Petersen Manufacturing Co. —H. A. H.

Precision Manufacture of Aluminium [Aircraft Equipment]. Gilbert C. Close (*Light Metal Age*, 1950, **8**, (5/6), 10–11, 20–21).—An account of the precision manufacture by the AiResearch Manufacturing Company, Los Angeles, of Al and Al alloy air- and heat-control equipment for fast, high-flying aircraft.—H. A. H.

New French Liners [Using Aluminium Alloys]. Pierre Vidal (*Rev. Aluminium*, 1950, (170), 374–376).—A description is given of the construction of a hangar on the liner " Liberté " for the transportation of passengers' motor cars. The material used was the Al–5% Mg alloy, and the hangar has a two-section, roller-type MacGregor hatch cover in its roof, which is made of the same material. In the liner " Kairouan," which has recently been refitted, the children's play room is constructed from sections and sheets of Al–5% Mg alloy, and 330 tourist-class bunks are made of the same material. —J. E. T.

Aluminium Survey Vessel " Ain-Al-Bahr." —— (*Engineering*, 1951, **171**, (4438), 188–189; and also *Engineer*, 1951, **191**, (4961), 260–262; *Metallurgia*, 1951, **43**, (256), 68).—Describes a 60-ft. survey vessel built on the " stressed skin " principle. The method of construction leads to a marked saving in weight, with consequent reduction in draught, size of propelling machinery, fuel tanks, &c. Extruded sections and hull plating are in AW5 alloy, and the deck is of AW10 alloy.—D. K. W.

The Developing Uses of Light Alloys for Boat Building. G. L. R. Watkins (*Light Metals*, 1950, **13**, (147), 206–214).—W. recapitulates the advantages of Al for marine appn. and gives numerous examples of its use in light craft. —J. E. T.

Aluminium-Sheathed Impregnated Pressure Cable. —— (*Machinery Lloyd*, 1950, **22**, (10), 99).—Al-sheathed cables are now available in a range of sizes and types. The sheath of an impregnated pressure cable is required to withstand an internal N pressure of 200 lb./in.^{2}, Al cables, unlike Pb, do not require an external steel reinforcement. Corrosion of Al in the buried condition is no greater hazard than when using Pb and steel reinforcement. The fatigue endurance of Al is much higher than Pb and it is unnecessary to take any special precautions for Al-sheathed cables installed in positions where vibration is severe.—H. Pl.

Aluminium Tubes for Electrical Conduit. Maurice Victor (*Rev. Aluminium*, 1950, (170), 363–367).—Cf. *Met. Abs.*, 1950–51, **18**, 399 and 558. Notes are given on the use of Al tubes for elect. conduit, with particular ref. to the installation at the Northern Aluminium Co.'s plant at Rogerstone.—J. E. T.

Connectors [in Aluminium Alloys] for Electrical Conduit. Jean Prieux (*Rev. Aluminium*, 1950, (170), 368–369).—The French standards for domestic elect. fittings accept the use of several types of light alloys for the manufacture of connectors for conduit. Such parts, made by gravity or pressure die-casting, are described, and their interesting characteristics, which include lightness, strength, ease of machining, and durability, are discussed.—J. E. T.

Uses of Light Metal Castings. A. Buckeley (*Neue Giesserei*, 1950, **37**, (10), 190–194).—A review of the developments of the light-metal casting prodn. in the British zone of Germany since the end of the war. B. gives some examples of castings for various appn. and shows ways to increase their use. He concludes with cost figures.—H. K.

Aluminium Castings Improve Design of Vane-Axial Fans. F. L. Church (*Modern Metals*, 1951, **6**, (12), 28–30).—The use of Al in vane-axial fans favoured by the U.S. Navy is described.—R. J.

Magnetic Separator Housed in Aluminium. —— (*Modern Metals*, 1951, **6**, (12), 32).—Al, being non-magnetic, is an ideal material for housing magnetic separators.—R. J.

[Aluminium Articles] In Hotel and Restaurant. —— (*Light Metals*, 1950, **(146)**, 115–119).—A description is given of equipment utilizing Al and its alloys shown at the 1950 Hotel, Restaurant, and Catering Exhibition, London. —J. E. T.

Aluminium Grain Bins Will Help Store Surplus Food Crops. —— (*Light Metal Age*, 1950, **8**, (1/2), 15). —H. A. H.

New Process for Manufacturing Aluminium Foil Tubes. —— (*Light Metal Age*, 1950, **8**, (5/6), 12–13).—A note on the introduction of a new machine for the prodn. of tubes, such as are at present used for toothpaste and shaving creams, from foil, laminated or coated on both sides with plastic. The advantages of this process are outlined.—H. A. H.

The Limits of Working Temperature for Copper Tubes Expanded into a Steel Tubeplate. J. G. Burnell (*J. Inst. Eng. Australia*, 1950, **22**, (10/11), 229–234).—Cu tubes expanded into a mild-steel tubeplate, commonly used in heaters and heat-exchangers, are often subjected to considerable pressures and temp. The stresses in an expanded element are briefly examined, and the effect of differential expansion between the tube and plate is discussed. The effects are briefly described of annealing and cold work on the properties of Cu. Tests are reported showing the holding power of Cu tubes expanded into steel tube-plates following various procedures. Temp. up to 320° F. (160° C.) may be used where every care is taken in expanding the tubes so as to secure 3% extrusion on the seatings. For temp. up to 500° F. (260° C.) 70 : 30 Cu–Ni tubes can safely be used. —T. A. H.

Fighter Wing Redesigned for Magnesium. D. H. Black (*Iron Age*, 1951, **167**, (13), 90–92).—Experimental wings of thick Mg sheet have been made and successfully applied to fighter aircraft. This construction eliminates many stiffeners, saves manufacturing time, and enables 30% more fuel to be carried.—J. H. W.

Magnesium Dockboard Employs Extruded, Cast, and Wrought Parts. Gene Beaudet (*Iron Age*, 1951, **167**, (5), 107–109).—B. describes the construction of an all-welded Mg dockboard that is as strong but only one-quarter as heavy as steel. A ¼-in. plate will carry up to 1000 lb. loads, a ⅜-in. plate up to 4000 lb., and a ½-in. plate up to 12,000 lb. with suitable reinforcement.—J. H. W.

Production of Ozone with Refrigerated [Platinum] Anodes. E. I. Lash, R. D. Hornbeck, G. L. Putnam, and E. D. Boelter (*J. Electrochem. Soc.*, 1951, **98**, (4), 134–137).—The use of Pt anodes for this purpose is described.—G. T. C.

***Service Behaviour of Electrodeposited Silver–Lead Bearings in Internal-Combustion Engines.** Martin Rossenbeck (*Z. Metallkunde*, 1950, **41**, (4), 109–114).—An illustrated description of service tests in single-cylinder I.C. engines, of electrodeposited bearings of Ag–Pb alloys contg. 0·4, 1·25, and 2·5% Pb, with layers of electrodeposited Ni and Cu interposed between the steel back and the bearing surface, resp. Irrespective of Pb content, such easily manufactured composite bearings were found to be : (1) much superior to those obtained with steel-backed electrodeposited Cu bearings, and (2) at least as good as those obtained with steel-backed bearings having cast-on liners of Pb bronze (Cu 70–85, Pb 30–15%), the successful manufacture of which is difficult. 10 ref.—E. N.

Tantalum Electrolytic Capacitors. M. Whitehead (*Bell Lab. Record*, 1950, **28**, (10), 448–452).—The design and characteristics of Ta capacitors are described.—D. M. D.

Titanium : Huge Expansion for Defence Paves Way for Civilian Use. F. L. Church (*Modern Metals*, 1951, **6**, (12), 19–21).—A summary of the potential uses of Ti. The properties which make Ti attractive as a commercial material are briefly reviewed, and C. then considers future appn. Some possible uses include appn. in the chem. and petroleum industries, for marine appn., and in aircraft, &c. Although the usefulness of the material is established, the major factor limiting its commercial exploitation is the rate of prodn. C. considers that Ti will mature at a much faster rate than either Al or Mg.—R. J.

Uses of High-Purity Zinc-Alloy Bearings. F. Richter (*Werkstatt u. Betrieb*, 1950, **83**, (4), 142–144).—The use of high-purity Zn-alloy bearings for various appn. is described. R. quotes practical results obtained with such bearings covering a wide field in order to draw conclusions for other appn.—H. K.

Lighter and Brighter Metals Find Increasing Application for Architectural Metalwork. Charles R. Fleishman (*Western Metals*, 1950, **8**, (7), 17–18).—A short survey on the appn. of Al, bronze, and stainless steel for building purposes.—T. G.

Critical-Metals Usage Slashed in Revised Television Circuits. A. H. Allen (*Steel*, 1951, **128**, (14), 78–81).—Describes the use of substitute and alternative metals and alloys for those in short supply in components such as those used in television sets.—E. J.

Centrifugally-Cast Hard-Metal Alloys Bring Longer Life and Improved Design. Charles E. Rogers (*Western Metals*, 1951, **9**, (1), 22–23).—Deals with the appn. of centrifugally cast hard metals, ferrous and non-ferrous. The base material of the non-ferrous hard metals is Co in which hard particles of Cr, W, &c., carbides are embedded.—T. G.

Cemented Carbides in Industry. H. W. Greenwood (*Machinery Lloyd*, 1950, **22**, (19), 88–93).—The addn. of TiC to WC eliminated the early cratering trouble in cemented-carbide cutting tools. The incorporation of the TiC brought about a rise in temp. at which there was a tendency for the tool material to weld to the machined part on the turning. TaC has been used with Ni as a cutting tool for steel. Pt has been used successfully to supply the liq. phase in those cases where Ni and Co have been prone to chem. attack. The use of cemented carbides as liners for dies in making powder-metal products has speeded up prodn. and reduced overall costs. The small rivets of brass, Au, and stainless steel are made by producing the button head in a cemented carbide tool rotating at high speed. Guides, eyelets, and needles made of cemented carbides have shown something of the order of 100 times the life under high-speed textile threads when compared with hardened steel, and very much more when compared with porcelain. Cemented carbides are also being successfully used as horseshoe studs, studs for shoes used by policemen and others having to patrol hard pavements, dirt-track riders' boot tips, and gramophone needles. —H. Pl.

Working Surfaces for Radiochemical Laboratories: Glass, Stainless Steel, and Lead. Paul C. Tompkins and Oscar M. Bizzell (*Indust. and Eng. Chem.*, 1950, **42**, (8), 1469–1475).—T. and B. describe a systematic attempt to develop simple tests for the comparison of the susceptibility of protective coatings and structural materials to radioactive contamination and their subsequent ease of decontamination. The data are reported in terms of two newly defined quantities called the decontamination index, $DI =$ log(activity on surface before contamination)/(activity on surface after contamination), and the spill index, $SI = -$log(activity on surface after decontaminating spill)/(total activity of sample spilled). The DI is proposed as a measure of the degree of removal of air-dried radioisotopes as contamination from surfaces, and the SI as a measure of the situation created in case of an accidental spill which is cleaned up within an hr. The presence of isotopic carrier in the decontamination reagent is demonstrated to be of little value, and the justification for the use of complexing agents is shown to rest on erroneous conceptions of the decontamination mechanism. Conclusions regarding the suitability of glass, stainless steel, and Pb for radiochem. laboratory surfaces are drawn, and a schedule for safe operation at various levels of activity is presented. 15 ref.—D. P. M.

Chemical-Engineering Materials of Construction: A Review of the Literature. —— (*Indust. and Eng. Chem.*, 1950, **42**, (10), 1951). **Aluminium Alloys.** E. D. Verink, Jr., and H. W. Fritts (*ibid.*, 1955–1956). **Wrought Copper and Copper-Base Alloys.** C. L. Bulow (*ibid.*, 1970–1977). **Nickel and High-Nickel Alloys.** H. O. Teeple (*ibid.*, 1990–2001). **Tin and Its Alloys.** Robert J. Nekervis (*ibid.*, 2020–2022). **Less Common Metals.** D. B. Broughton (*ibid.*, 2023–2025). **Properties of Chemical-Engineering Materials of Construction.** —— (*ibid.*, 2026–2076).—General reviews, including tables of compn., mech. properties, appn. and uses, together with an extensive *bibliography*.—D. P. M.

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23 — MISCELLANEOUS

†**Chemical and Metallurgical Equilibria in the Works of Henry Le Chatelier and in Present-Day Science.** Maurice Rey (*Rev. Mét.*, 1951, **48**, (3), 161–172).—R. illustrates Le Chatelier's work and its relation to modern theories by discussing his investigations into chem. equilibria at high temp. and into the equilibria of soln. 14 ref.—J. H. W.

Mining and Metallurgical Meeting at Freiberg, 17–19th July, 1950. —— (*Metall*, 1950, **4**, (17/18), 380–381).—A brief report of some of the papers presented at the meeting. — *Wagemann* spoke on Re, its prodn. and utilization. Tech. pure Re, contg. ≯0·01% Fe can be recovered from flue dusts in Pb smelting. It can be used in electrolytic apparatus, catalysers, pyrometry, jewellery, electroplating, and opt. equipment. — *Morawietz* discussed the prodn. of pure Ti and Zr—by the Kroll and van Arkel processes, and a modification of the latter used at Bitterfeld. The properties and economic utilization of the metals were mentioned. *C. F. Althof* dealt with inter- and transcryst. corrosion—especially with regard to stress corrosion and its damaging effects on steels as well as light metals. *O. Emicke* described materials, rolling mills and rolling procedures for the prodn. of Al foil.—E. N.

The Status of the Aluminium Industry. —— (*Light Metal Age*, 1950, **8**, (9/10), 10, 26).—A report by the Aluminum Association of America.—H. A. H.

Aluminum Company of America Point Comfort Works. —— (*Mines Mag.*, 1950, **40**, (8), 13–16).—N. B. V.

Alcoa Today: Vast Growth Under Way for War or Peace. —— (*Modern Metals*, 1950, **6**, (11), 22–25, 28–29, 31, 33–34).—Describes the expansion of the Aluminum Company of America from 1911 to the present day. Includes a description of the Company's 31-story skyscraper in which extensive use is made of Al.—R. J.

Canada's Arvida [Aluminium] Works. —— (*Esso Oilways*, 1950, **3**, (4), 1–6).—Plant and processes at the Arvida (Quebec) works of the Aluminum Company of Canada are described and illustrated. Ref. are made to the history, properties, and appn. of Al.—J. R.

The Aluminium Industry [in India]. S. K. Ghaswala (*Sci. and Culture*, 1951, **16**, (8), 343–347).—A general survey of the Al industry with special ref. to Al prodn. in India. Alloy developments are traced and their appn. in various engineering and indust. fields is outlined from the standpoint of methods of manufacture. G. concludes with statistical data on world prodn. of ferrous and non-ferrous metals. —S. K. G.

Italian Aluminium Makers. —— (*Esso Oilways*, 1950, **3**, (4), 12–13).—Prodn. and processing of Al in Italy are illustrated and briefly discussed. Mention is made of Italian conditions (i.e. the availability of hydro-elect. power and the comparative lack of coal deposits) encouraging the substitution of home-produced Al for imported metals. An instance of this is the wide use of Al in place of Cu in elect. appn., an important factor in a country the industry of which is largely dependent on elect. power.—J. R.

Norwegian Power and Aluminium. —— (*Esso Oilways*, 1950, **3**, (4), 14–16).—The Norwegian Al industry is discussed with particular ref. to the utilization of hydro-elect. power. The illustrations include a view of one of the pot-houses at A/S Årdal Verk.—J. R.

Aluminium Technology in Switzerland. —— (*Esso Oilways*, 1950, **3**, (4), 17–18).—A brief outline is given of the development of the Swiss Al industry. A diagram illustrating the raw-material and power requirements for the prodn. of a ton of Al is reproduced.—J. R.

Aluminium and Its Alloys in 1950 : Some Aspects of Research and Technical Progress Reported. E. Elliott (*Metallurgia*, 1951, **43**, (256), 69–72).—A review with 42 ref. —F. M. L.

Aluminium Supply and Demand : Development of New Applications Likely to be Retarded. —— (*Metallurgia*, 1951, **43**, (256), 65–66).—A review.—F. M. L.

A Hundred Years of Aluminium : A Review of Its Evolution and Development. E. G. West (*Sheet Metal Ind.*, 1951, **28**, (289), 441–450).—W. follows the gradual development of Al and discusses prodn.; alloy development; gradual development in appn., including these in elect., food and chem., building, marine, transport, and mining industries. —R. J.

Cobalt. —— (*J. Metals*, 1951, **191**, (1), 17–24).—A report on this strategic metal, discussing : (1) sources—the Belgian Congo, N. Rhodesia, U.S.A., Canada, and French Morocco; (2) current and future appn., with special ref. to its use in magnet steels, Stellite-type alloys, alloys resistant to high temp. and erosion, and die and cutting tools, and (3) future supply possibilities.—E. N.

Ancient Syrian Coppers and Bronzes. Robert J. Braidwood, Joseph E. Burke, and Norman H. Nachtrieb (*J. Chem. Education*, 1951, **28**, (2), 87–96).—Cast and wrought Cu and bronze specimens from two archæological sites embracing cultural sequences from ~5000 to 2000 B.C. in Syria are illustrated, and chem. analyses and metallographic data are presented and discussed. The first appearance of Cu tools is in the period ~3500–3100 B.C., and by ~3100–2800 B.C. a relatively elaborate metal industry showing evidences of sound casting techniques and intentional addn. of such elements as As and Sn was established; this coincidence of a high standard of met. art with the first appearance of metal objects is an outstanding finding from examination of the specimens. Some corrosion phenomena observed are briefly discussed.—J. R.

Greek Bronze Statue. Pierre Huberson (*Fonderie*, 1951, (62), 2366–2367).—H. describes a fifth-century statue of Christ, 1·80 m. (70 in.) high, cast in bronze in seven pieces. —J. H. W.

Gold–Copper Alloys in Ancient America. William C. Root (*J. Chem. Education*, 1951, **28**, (2), 76–78).—Casting and gilding of Cu–Au alloys by sixteenth-century American Indians are discussed.—J. R.

World Production and Consumption of Lead: Economic Aspects (1938–1949). —— (*Metall*, 1950, **4**, (17/18), 386–387).—E. N.

Severe Investigation Shows Magnesium Logging Wedges Not a Fire Hazard. E. M. Cramer and A. R. Hard (*Light Metal Age*, 1950, **8**, (7/8), 10–11, 21).—A report on tests by the Division of Industrial Research of the State College of Washington to determine if a fire hazard exists by the use of Mg alloy wedges in logging operations. It is concluded that this use does not constitute a fire hazard.—H. A. H.

The International Nickel Company of Canada. —— (*Metall*, 1950, **4**, (17/18), 385).—E. N.

A Rapid Immersion Silvering Method of High Efficiency. P. B. G. Upton and E. F. G. Herington (*Research*, 1950, **3**, (6), 289–290).—A method of deposition of Ag films is described; this is by reduction of a soln. of Ag ammonionitrate by hydrazine sulphate, in which the deposition of Ag elsewhere than on the desired surface is prevented by the presence of pyridine. Inhibitors other than pyridine can be used, such as cyclohexylamine or NH_3, provided that an addn. of cetyl pyridinium bromide is also made to the soln.—F. A. F.

Chemical Deposition of Silver on Non-Conductive Bodies. John T. Owen (*Plating*, 1951, **38**, (4), 353–357).—A summary of the chem. deposition of Ag on non-conducting bodies by the spray method is presented and in addn. information is given on the prepn. of objects and soln. for use in other methods also. Details are included for the silvering of glass and of plastics and the lacquering and colouring of the deposits and appn. of electrodeposits over them is briefly mentioned. Information is presented on the measurement of brightness and on the detn. of film thickness. —G. T. C.

The Asiatic Tin Industry. —— (*Metall*, 1950, **4**, (23/24), 531–532).—A brief outline of its growth during the years 1930–49.—E. N.

Wood's Metal Replaces Assembly Fixtures. Ben C. Brosheer (*Machinist* (*Eur. Edn.*), 1950, **94**, (52), 1882–1884).—Turbine blades for hydraulic torque converters were set in place with Wood's metal instead of by complicated fixtures, resulting in no adjustments, plenty of speed, and reduced costs.—J. H. W.

Virgin Metals in North America. I.—The U.S.A. Richard Weyl (*Metall*, 1950, **4**, (23/24), 527–530).—A statistical review of the annual prodn. and consumption, from 1939 to 1949, of: Au, Ag, Hg, Pt, Cu, Sn, Pb, Zn, Al, Mg, Cr, W, V, Mo, Sb, and Ti.—E. N.

***" Alsithermal " Metallurgy.** Arturo Paoloni (*J. Four Elect.*, 1950, **59**, (1), 18–20; (2), 44–46).—The uses and properties of Al, C, or Si as a reducing agent in sp. thermal met. reactions is briefly discussed. " Alsithermal " metallurgy employs an alloy contg. Al 50, Si 35, Fe 10, Ti 2·5–3, and C 1–1·5% as the reducing agent. Experiments on the " Alsithermal " reduction of pyrites and of a Mn and a Ti mineral are described.—H. A. H.

†Radioactive Isotopes in Industry. I.—Metallurgy. —— (*Atomics*, 1951, **2**, (2), 39–42).—Radioactive Fe (Fe^{55} or Fe^{59}) has been used to determine the diffusion rates that enter into such phenomena as annealing, hardening, recrystn., grain-growth, and high-temp. behaviour in steel. The diffusion of C in α-Fe has been studied with radioactive C^{14} at temp. below 725° C. Radioactive Zn and Cu have been used to study self-diffusion in stressed metals. Radioactive W^{185} has been used to locate the W in a 70 : 25 : 5 Ni–Cr–W alloy; it was found in the dendrites. In a Co-base alloy it was shown that the W is generally associated with the intermetallic compounds and solid soln. between the dendrites. The few tenths % of Al present in Si rectifiers are located by irradiation with neutrons; radioactive P has been used to measure and locate P in alloys. Radioactive Fe and other radioactive isotopes have been used to study corrosion problems, metal-cleaning processes, and in the study of friction between piston-rings and radioactive Cr-plated surfaces; a very much smaller amount of worn metal can be detected than with conventional techniques. Factory-scale investigations carried out by radioactive-tracer technique include a study of gas velocities and times of contact between gases and solids in blast furnaces, and a study of the distribution of pyritic and organic S present in coal for blast-furnace coke. The ratio of pyritic to organic S in coal does not influence the S content of the finished coke. It is suggested that the flow of metal during extrusion, casting, or forging be studied by the new tracer technique.—J. S. G. T.

Radioisotopes Aid Studies of High-Temperature Materials. W. E. Jones (*Steel*, 1951, **128**, (13), 78).—The use of a radioactive W isotope for the study of the structure of a Co-base alloy is described. Health precautions are discussed. —E. J.

24 — BIBLIOGRAPHY

(Publications marked * have been received.)

American Society of Mechanical Engineers. *Bibliography on Thermostatic Bimetals, Low-Expansion Alloys, and Their Applications.* 11 × 8½ in. Pp. 52. [1950.] New York: The Society. ($1.00.)

***American Society for Metals.** *Atom Movements.* A Seminar on Atom Movements held during the Thirty-Second National Metal Congress and Exposition, Chicago, October 21–27, 1950, Sponsored by the American Society for Metals. 9 × 6 in. Pp. [vii] + 250, illustrated. 1951. Cleveland (3), O.: The Society, 7301 Euclid Avenue. ($5.00.)

[Contents: L. S. Darken, "Formal Basis of Diffusion Theory"; Cyril Wells, "Chemical Techniques and Analysis of Diffusion Data"; R. E. Hoffmann, "Tracer and Other Techniques of Diffusion Measurements"; H. B. Huntington, "Mechanisms of Diffusion"; J. Bardeen and C. Herring, "Diffusion in Alloys and the Kirkendall Effect"; C. E. Birchenall, "Volume Diffusion—An Empirical Survey"; D. Turnbull, "Grain-Boundary and Surface Diffusion"; Carl Wagner, "Diffusion and High-Temperature Oxidation of Metals"; F. N. Rhines, "Gas–Metal Diffusion and Internal Oxidation"; Pol Duwez, "Diffusion in Sintering"; J. E. Burke, "The Migration of Grain Boundaries"; Robert F. Mehl, "Summary".]

***Anglo-American Council on Productivity.** *Non-Ferrous Metals (Wrought).* Report of a visit to the U.S.A. in 1950 of a Productivity Team representing the Wrought Non-Ferrous Metals Industry. 9¾ × 7¼ in. Pp. xiv + 96, with 28 illustrations. 1951. London: The Council, 21 Tothill Street, S.W.1. (4*s.* 6*d.*, post free.)

Bianchi, L. W. P., and E. M. Zweekhorst. *Metalen en Metaalbewerking.* 26·5 × 19 cm. Deel I.—*Metalen, legeringen, metaalonderzoek.* Pp. 224, with 198 illustrations. Deel II.—*Bewerken van metalen in plastische toestand, walsen, smeden, persen en trekken.* Pp. 180, with 312 illustrations. 1951. Rotterdam: Nijgh and Van Ditmar N.V. (Deel I, fl. 11.75; Deel II, fl. 15.50.)

Boston, Orlan William. *Metal Processing.* Second, revised edition. Pp. 763. 1951. New York: John Wiley and Sons, Inc. ($7.50.)

***Coghlan, H. H.** *Notes on the Prehistoric Metallurgy of Copper and Bronze in the Old World.* (Occasional Papers on Technology, No. 4). 9¾ × 7¼ in. Pp. 131, with 21 illustrations in the text and 16 plates. 1951. Oxford: Pitt Rivers Museum. (15*s.*)

***Du Mond, T. C.** Edited by. *Engineering Materials Manual.* 11½ × 8½ in. Pp. [ii] + 386, illustrated. 1951. New York (18): Reinhold Publishing Corp. ($4.00); London: Chapman & Hall, Ltd. (36*s.*)

[A reprint of 28 of the "Materials and Methods Manuals" published in *Materials and Methods* in recent years.]

***Giesserei.** *Sonderausgabe "Konstrukteur und Giesser."* 29·5 × 21 cm. Pp. 88, illustrated. 1951. Düsseldorf: Giesserei-Verlag, August-Thyssen-Strasse 1. (DM 7.20.)

[Contents: J. H. Küster, "Form- und giessgerechte Konstruktion und deren Einfluss auf die Preisgestaltung der Gusstücke"; E. Hugo and C. W. Pfannenschmidt, "Werkstoffmerkmale und Gestaltungsrichtlinien für Grauguss"; A. Heuvers, "Hochwertiger Stahlguss"; W. Trommer, "Gegossene Konstruktionselemente"; G. Schmidt, "Giessgerechtes Konstruieren beim Stahlformguss"; H. Resow, "Einige Konstruktionsregeln fur Stahlguss"; F. Roll, "Beitrag zu Konstruktionsrichtlinien des Werkstoffes Temperguss"; A. Buckeley, "Verwendungsgrundsätze fur Aluminiumguss"; Ph. Schneider, "Über die konstruktive Gestaltung von Leichtmetall-Sandguss- und Kokillengussteilen"; C. A. Piel, "Die Planung beim Einsatz von Kokillengussteilen in Schwermetallen auf Kupferbasis"; J. Hagen, "Konstruktions- und Modellgestaltung"; E. Piwowarsky, "Schweissen oder Giessen?"; ——, "Schrifttum über Konstruktion von Gusstücken".]

***Gillett, H. W.** *The Behaviour of Engineering Metals.* 9 × 6 in. Pp. xvi + 395, with 58 illustrations. 1951. New York: John Wiley & Sons, Inc. ($6.50); London Chapman & Hall, Ltd. (52*s.*)

Haas, Ph. *Altstoffe und Abfälle.* Pp. 96. [1950.] Berlin: Marktwart-Verlag, G.m.b.H. Sold by Georg Achterberg, Berlin-Lichterfelde. (Brosch., DM 3.60.)

Haase, L. W. *Werkstoffzerstörung und Schutzschichtbildung im Wasserfach.* 2. Auflage. 17·5 × 15 cm. Pp. 218, with 102 illustrations. 1951. Weinheim/Bergstrasse: Verlag Chemie. (Geb., DM 15.80.)

***Hot-Dip Galvanizers Association.** *International Conference on Hot-Dip Galvanizing Held at Copenhagen 17–21 July 1950.* 9½ × 6¼ in. Pp. 161, illustrated. 1951. Oxford: Zinc Development Association, Lincoln House, Turl Street. (30*s.* net.)

[Contents: W. L. Hall and D. N. Fagg, "Twenty Years of Research for the Galvanizing Industry"; M. L. Hughes, "The Influence of Aluminium on the Alloying Action in Hot-Dip Galvanizing"; R. W. Bailey, "Some Thermal Consideration in the Heating of Galvanizing Baths"; H. Bablik, "Wet Galvanizing versus Dry Galvanizing"; T. Liban, "Some Notes on the Relative Merits of Flux Galvanizing and Dry Galvanizing"; R. W. Bailey, "A Survey of the Formation of Zinc Ash, Dross, and Flux Skimmings from Galvanizing Baths During 1949"; W. G. Imhoff, "Can Dross Be Eliminated in Hot-Dip Galvanizing?"; E. F. Pellowe, "The Corrosion-Resistance of Galvanized Coatings with Special Reference to Galvanized Window Frames"; R. Lewis Stubbs, "The Outlook for General Galvanizing".]

***Institute of Metals.** *The Hot Working of Non-Ferrous Metals and Alloys.* A Symposium on Metallurgical Aspects of the Subject held in London on the occasion of the Annual General Meeting of the Institute, 30 March 1950. (Institute of Metals Monograph and Report Series No. 9.) Demy 8vo. Pp. 208, illustrated. 1951. London: The Institute, 4 Grosvenor Gardens, S.W.1. (15*s.* ($2.50), post free; 7*s.* 6*d.* ($1.50) to members.)

[Contents: F. Kasz and P. C. Varley, "The Hot Rolling of Aluminium and Its Alloys"; Christopher Smith, "The Extrusion of Aluminium Alloys"; F. E. Stokeld, "The Hot Forging and Hot Stamping of Aluminium and Its Alloys"; R. G. Wilkinson and F. A. Fox, "The Hot Working of Magnesium and Its Alloys"; Maurice Cook and Edwin Davis, "The Hot Working of Copper and Copper Alloys"; D. W. Dugard Showell, "The Hot Working of Tin Bronzes"; L. H. Back, "The Hot Working of Lead and Lead-Rich Alloys"; C. W. Roberts and B. Walters, "The Rolling of Zinc and Zinc-Rich Alloys."]

Kececioglu, D. Compiled by. *Bibliography on Plasticity: Theory and Application.* 11 × 8½ in. Pp. 191. 1950. New York: American Society of Mechanical Engineers. ($2.25.)

***Kortüm, G., and J. O'M. Bockris.** *Textbook of Electrochemistry.* Volume I. 24 × 16 cm. Pp. xvi + 351, with 48 illustrations. 1951. Amsterdam: Elsevier Publishing Company; London: Cleaver-Hume Press, Ltd., 42A South Audley Street, W.1. (50*s.*)

Krämer, Oskar P. *Die Hartverchromung.* 21 × 15 cm. Pp. 120, with 78 illustrations. 1951. Berlin W15 (Westberlin): Metall-Verlag G.m.b.H., Düsseldorfer Str. 38. (Geb., DM 8.40.)

***Langdon-Davies, John.** *N.P.L.: Jubilee Book of the National Physical Laboratory.* 8¼ × 6 in. Pp. 104, with 168 illustrations. 1951. London: H.M. Stationery Office. (4*s.* net.)

***Mohler, J. B., and H. J. Sedusky.** *Electroplating for the Metallurgist, Engineer, and Chemist.* 8½ × 5½ in. Pp. 257, with 26 illustrations. 1951. New York (10): Chemical Publishing Co., Inc., 212 Fifth Avenue. ($5.00.)

***Mond Nickel Co., Ltd.** *Isothermal Transformation Diagrams for Nickel Steels.* 9¾ × 6½ in. Pp. 57, illustrated. [1951.] London: The Company, Sunderland House, Curzon St., W.1.

Patent Publications. *Electropolishing and Electrobrightening.* 8½ × 11 in. Pp. 19. 1951. Washington (15): Patent Publications, Box MF–4094, D.C. 19. ($5.00.)

***Physical Society.** *Reports on Progress in Physics.* Volume XIV (1951). Cr. 4to: Pp. [iv] + 412, with numerous illustrations. 1951. London: The Society, 1 Lowther Gardens, Prince Consort Road, S.W.7. (50*s.* to non-Fellows; 27*s.* 6*d.* to Fellows; postage 1*s.* extra.)

[Contents: W. C. Price, " Recent Advances in Ultra-Violet Absorption Spectroscopy "; W. E. Lamb, Jr., " Anomalous Fine Structure of Hydrogen and Singly Ionized Helium "; H. Kuhn, " New Techniques in Optical Interferometry "; E. Wolf, " The Diffraction Theory of Aberrations "; A. B. Meinel, " The Spectrum of the Airglow and the Aurora "; B. J. Mason and F. H. Ludlam, " The Microphysics of Clouds "; M. Deutsch, " Angular Correlations in Nuclear Reactions "; G. D. Rochester and W. G. V. Rosser, " The Nuclear Interactions of Cosmic Rays "; E. W. Foster, " Nuclear Effects in Atomic Spectra "; N. C. Gerson, " A Critical Survey of Ionospheric Temperatures "; W. V. Mayneord, " Some Applications of Nuclear Physics in Medicine ".]

Ricci, John E. *The Phase Rule and Heterogeneous Equilibrium.* Pp. 505, illustrated. 1951. New York: D. Van Nostrand Co., Inc. ($12.00.)

***Rogers, Bruce A.** *The Nature of Metals.* (Self-Education Series). 7½ × 5 in. Pp. vi + 248, with 121 illustrations. 1951. Cleveland (3), O.: American Society for Metals, 7301 Euclid Avenue, jointly with the Iowa State College Press. ($3.00.)

Sachs, George. *Principles and Methods of Sheet-Metal Fabricating.* 9 × 6 in. Pp. 526, illustrated. 1951. New York: Reinhold Publishing Corp. ($10.00.)

Schönert, Karl, und Rudolf Eschelbach. *Praktische Metallprüfung.* Band 1: *Die mechanisch-technologischen Prüfverfahren und ihre Anwendung.* Pp. 274, with 206 illustrations. 1950. Brunswick: Georg Westermann Verlag. (Geb., DM 28.—.)

Seelmeyer, Günther. *Rost- und Steinschutz in Niederdruckanlagen.* Pp. 288, with 189 illustrations. 1951. Weinheim/Bergstrasse: Verlag Chemie. (Geb., DM 19.80.)

Vogel, Otto. *Handbuch der Metallbeizerei.* I. Band: *Nichteisenmetalle.* Neubearbeitet von Hermine Vogel, Karl Kaiser, Hugo Krause und Friedrich Sierp. 2. Auflage. Pp. xv + 410, with 263 illustrations. 1951. Weinheim/Bergstrasse: Verlag Chemie. (Geb., DM 49.20.)

***Wiggin (Henry) and Co., Ltd.** *High-Nickel Alloys for Heat-Resisting Equipment.* 10 × 7¼ in. Pp. 20, illustrated. [1951.] Birmingham (16): The Company, Wiggin Street. (Gratis.)

Winkler, Helmut Gustav Franz. *Struktur und Eigenschaften der Kristalle. Eine Einführung in die physikalische und chemische Kristallkunde.* Pp. viii + 258, with 62 illustrations and 79 tables. 1950. Berlin–Göttingen–Heidelberg: Springer-Verlag. (Brosch., DM 16.80.)

25 — BOOK REVIEWS

Progress in Metal Physics. Volume II. Edited by Bruce Chalmers. 9½ × 6 in. Pp. viii + 213, with 107 illustrations. 1951. London: Butterworths Scientific Publications (45*s.*); New York: Interscience Publishers, Inc. ($8.00.)

The contents of this second volume of the series confirm the impression gained from the first one that the Editor's policy is to emphasize the metallurgical side of metal physics, a policy with which readers of this *Journal* must surely agree. The impression comes not so much from the choice of subjects for the reviews—although the less practical ones are certainly avoided—as from the way in which they are expounded. The reader does not have to know much more about metal physics than is to be expected of an average metallurgist, and will often find that his scientific pill has been sugared with a few practical tips.

Little specialized knowledge is assumed in the very readable essay by Dr. H. Lipson on " Order–Disorder Changes in Alloys ", with which this volume begins. Avoiding the severely mathematical theories of co-operative phenomena, the author has chosen to give instead a descriptive, though critical, account of the main experimental facts and their interpretations in terms of intuitive physical models. The reader is led skilfully and gently from elementary notions about lattices and X-ray diffraction to very recent and advanced experiments on short-range order and domain structures. More mathematical at first, although becoming distinctly practical towards the end, the article on " Anisotropy in Metals " contributed by Dr. W. Boas and Mr. J. K. Mackenzie is extremely good. Realizing that one cannot discuss anisotropy properly without using the mathematics of directional properties, they begin with a lucid introduction to vector relations, and this places them in a strong position for the subsequent discussion of various interesting metallurgical effects.

The metallurgical aspect is emphasized heavily in Mr. H. G Warrington's article on " Developments in Magnesium Alloys." This contains little metal physics, although sections are included on such topics as the nucleation theory of grain refinement, and magnesium–lithium alloys. One feels that the author's real interest is in metal economics, and that he would have preferred to continue along the line of thought suggested by his interesting theory of how and why the number of successful commercial alloys changes with time (p. 124). A different view of time and change is taken in Mr. I. I. Betcherman's essay on " Rate Processes in Physical Metallurgy." In this case activation energy is the root of all changes; and these are timed by the Arrhenius equation. Although its importance is beyond question, it is less certain that this equation is the best starting point for understanding such processes as precipitation, creep, and recrystallization. So many subtle and unexpected things happen when atoms move about in solids that one could hardly expect to pick on the right explanations just by considering energies and entropies of activation.

A glimpse into some of these subtleties is provided in three very topical articles on " Polygonization," with which this volume concludes. In the first of these, a clearly written and well-documented survey by Dr. R. W. Cahn, sound reasons are given for recognizing that at least five different processes, recovery, polygonization, recrystallization, grain-growth, and secondary recrystallization, may take place when a worked metal is heated; the phenomenon of strain-ageing shows that even this list is not exhaustive. In this article and those which follow, by MM. A. Guinier and J. Tennevin, and by MM. C. Crussard, F. Aubertin, B. Jaoul, and G. Wyon, which do justice to the distinguished French contributions in this field, there is an exciting feeling that here one is actually witnessing the making of metal physics.

The excellent printing and binding of the first volume set a high standard, which is fully maintained in the present one. And so it should be, since 45*s.* is a lot of money for 213 pages, even though these carry articles of very high scientific value. It is to be hoped that potential customers will not demonstrate the truth of Mr. Warrington's thesis that, in practice, economic factors always predominate over scientific interests.

A. H. Cottrell.

Gmelins Handbuch der anorganischen Chemie. 8. Auflage. Herausgegeben vom Gmelin-Institut in Clausthal-Zellerfeld. System-Nummer 68: **Platin.** Teil A–Lieferung 6: **Die Legierungen der Platinmetalle: Osmium, Iridium, Platin.** 25·7 × 17·5 cm. Pp. xxii + 719–854, with 74 illustrations. 1951. Weinheim/Bergstrasse: Verlag Chemie G.m.b.H. (Brosch., DM 36.—.)

Those who are familiar with Gmelin's series of handbooks will know what to expect, and they will not be disappointed with the volume under review. They should stop reading this notice at this point.

It might, however, be useful to say a little more to newcomers who may not be fully acquainted with these excellent publications. It is worth noting that although a wide and detailed knowledge of German is probably required to make the fullest possible use of the text, anyone who has done "Basic German for Science Students" and taken it seriously, can use the Handbuch with profit.

Lieferung 6 completes the series devoted to the platinum metals. It provides summarized information on most of the published work on the alloys of osmium, iridium, and platinum up to December 1949, and thus fills an appreciable gap. The Howe–Baker bibliography is not yet available beyond 1940, and Hansen's "Der Aufbau der Zweistofflegierungen" is getting badly out of date. Moreover, the volume under notice is of much wider scope; it is far more than a bibliography, and it covers in addition to equilibrium diagrams, mechanical properties and heat-treatment, magnetic and electrical properties, and some of the more noteworthy chemical properties including reference to catalytic activity. The information is not limited to binary alloys, and it is comprehensive. One can learn from the first page, for example, that there is no published information about alloys of osmium with polonium and its isotopes.

Those who say, with regret or otherwise, that there is little interesting work to be done on constitutional diagrams of binary alloys, might like to note that there is no diagram worthy of the name for any of the alloys of osmium or iridium.

The collaborating authors have cast their nets very wide, as can be seen from the many references to Russian journals, manufacturers' literature, and other sources which do not come to the notice of metallurgists in the ordinary way. References to "Metals Handbook" seem, strangely enough, to be to the 1939 edition only. Perhaps they've got something there.

All metallurgists who are interested in the platinum metals should have this volume handy.

A. G. Dowson.

Bibliography of Electron Microscopy. Edited by V. E. Cosslett. 8½ × 5½ in. Pp. 350. 1950. London: Edward Arnold and Co. (40*s.* net.)

Electron microscopy has become a well-organized craft which has been particularly fortunate in the publication of several valuable bibliographies of its literature.

Much the most elaborate of these is now becoming available; this new bibliography covers the literature to the end of 1948, and, under the auspices of the Institute of Physics, quarterly supplements to it are being prepared and published in the *Journal of the Royal Microscopical Society*.

The entries in the main bibliography and in each supplement are arranged in alphabetical order of the authors' names. Checking against previous bibliographies shows that approximately 80% of the papers then listed are included in the new effort. Just 40% of the 2200 references in the present volume carry abstracts, but no obvious system seems to have been used in choosing a particular type of paper for abstraction.

The bibliography is available printed on one side only of a series of sheets which may be cut up by each user and filed by any indexing system. The quarterly supplements are also to be printed on one side of the page. In addition, the main bibliography has been issued as a bound volume printed on both sides of the paper. In this volume also, the references are in alphabetical order of the authors' names and are not otherwise indexed. The exact purpose which this volume, which is the form submitted for review, is intended to serve is obscure, but to anyone who requires a complete and well-printed and bound collection of abstracts, here it is.

G. A. Geach.

N.P.L.: Jubilee Book of the National Physical Laboratory. By John Langdon-Davies. 8½ × 6 in. Pp. 104, with 168 illustrations. 1951. London: H.M. Stationery Office. (4*s.* net.)

The National Physical Laboratory completed the first 59 years of its existence on 1 January 1950, but, by order of the Lord President of the Council, the Jubilee is being celebrated during the Festival of Britain. And here, prepared for the Department of Scientific and Industrial Research by the Central Office of Information, is part of that celebration. I can only say that the book disappoints me.

The National Physical Laboratory has a wonderful tale to tell, but it can't possibly be told in a series of snippets, culled apparently from the long list of the Laboratory's annual reports. It is, however, quite typical of this age of smatter-knowledge, quizzes, town forums, and "have-a-go", where, too often, incomplete knowledge, or even complete ignorance, masquerades as almost omniscience, and is correspondingly rewarded. It says both too little and too much—too little to be of any real value to scientists, and too much that cannot be in any way intelligible to the non-scientific reader. Look at the outside page of the back cover, on which is pictured "a model built to show changes that take place in a series of alloys on heating or cooling"—just that and no more; no reference to the constituent metals nor to the significance of the three co-ordinate axes. What would a lay reader make of it? I showed it to three of my friends and all three thought it might be a preliminary model of the Giant Dipper at the Fun Fair at Battersea!

Yet the book does contain some interesting information. Thus we learn from p. 34 that the gross annual expenditure rose from £204,507 in 1920 to £751,984 in 1950, exclusive of expenditure on printing, telephones, electricity, and maintenance of buildings; during the same period, income from industry, &c., for testing and research rose from £45,848 to £236,850. I am intrigued to know how and why the N.P.L. is the custodian, on behalf of the Ministry of *Transport*, of the electrical standards representing the ampere, the volt, and the ohm (p. 45). It is interesting to read that a strain gauge, measuring only 3 × 2 mm. overall, has been constructed for determining the loads on teeth during the mastication of food (p. 76). They are evidently all set at the Laboratory for the next arrival of our meat supplies. Or is it the Government's intention to specify the breaking strength to which our National Health dentures must, in future, conform? And why, oh why, have experiments been made on the towing of ship models *upside down* (p. 52)?

Of most interest to metallurgists is the short section devoted to materials, including metals and alloys (pp. 59–70). Here they will find all too brief reference to systematic research on alloys, the effects of impurities on the properties of copper, air-warfare alloys including Y alloy, impurities in iron, fatigue, strength of gas cylinders, stresses in metals, tools for metal research, and heat-resistance of metals, from which it is certainly very interesting to learn that the Engineering and Metallurgical Divisions of the Laboratory are, at present, concentrating on preparing for the demands of the gas-turbine age, and that the element titanium is now being as systematically studied as aluminium was examined a generation ago. Altogether, this paper-bound book has given me a very pleasant couple of hours' reading, but has left me hungering for very, very much more. Still, all interested in science will find it very good value for 4*s.*

J. S. G. Thomas.

Treatise on Powder Metallurgy. Volume II.—**Applied and Physical Powder Metallurgy.** By Claus G. Goetzel. 9 × 6 in. Pp. xviii + 910, with 628 illustrations. 1950. New York: Interscience Publishers, Inc. ($18.00); London: Interscience Publishers, Ltd. (144*s*.)

This second volume, which well maintains the high standard of the first volume of Dr. Goetzel's monumental compilation (see *Met. Abs.*, 1949–50, **17**, 410), may conveniently be considered as divided into three sections. The first, and by far the longest, describes in considerable detail the methods employed in the production by powder metallurgy of an astonishingly wide range of industrial materials, and lists their characteristics. The second is an encyclopædic survey of the properties and possible uses of " the metals and alloys which, if produced by the powder method, have potential uses in industry or science ". The third presents a masterly review of all the theories of bonding and sintering which have so far been proposed.

The ten chapters comprising the first section deal, respectively, with refractory metals and alloys, hard metals, electrical materials, magnetic materials, ferrous and non-ferrous structural parts, porous products, friction materials, dental alloys, and miscellaneous metal-powder applications. In all, they present a remarkably complete account of the achievements of powder metallurgy; and in particular they provide a most comprehensive summary of all that has been published up to the end of 1949. Inevitably, they differ to some extent in their weight and authority. The accounts of the production of sintered iron parts, copper-infiltrated iron pressings (called " cemented steels "), and sintered copper and copper alloys, bear the stamp of first-hand experience. The chapters on tungsten and " hard metals " are adequate, but in the references to some of the more unusual applications it is possible to detect signs of faltering. Thus, the requirements of the user have not been fully appreciated in the discussions on insulated magnetic cores and on dental amalgams, and there are some rather facile generalizations. The reviewer has noted also a few omissions: for instance, there is no reference to Swanger's work on ductile rhodium, or to the platinum–tungsten–thoria wires made in this country from sintered compacts during the last war for sparking-plug electrodes. This, however, is only to say that, as in all human endeavour, the survey is not perfect; it is, nevertheless, first-class.

The second section, on materials " for potential industrial use " is less successful. In his attempts to be all-embracing, the author has swept together all the odd tit-bits of information on unusual alloys he has been able to find—often the unconfirmed results of a few preliminary tests—with the consequence that this section is " bitty ", repetitive, uneven in quality, and often unreliable and even misleading. It should be regarded as an annotated guide to the literature rather than as a source of inspiration.

The last part of the volume, however, can be unreservedly recommended. Here Dr. Goetzel passes in review, very fully and without bias, all the theories that have been advanced to explain the changes in volume and structure which go on in the operations of pressing and sintering. He has, as he says, resisted the temptation to add to the number with a theory of his own; but in the final summary of established facts and proposed explanations the truth must surely lie not far beneath the surface.

With this volume Dr. Goetzel's main task has been completed, for the third volume is to be essentially a bibliography. Viewed as a whole, it compels admiration for the thoroughness with which every aspect of the subject has been treated and with which nearly every published word has received some mention. Its faults stem from this urge to be all-inclusive. There is some lack of unity and too much avoidable repetition in the two volumes. Dr. Goetzel has admirably supplied the facts and the documentation—it is to be hoped once and for all. Perhaps, after a period of rest and assimilation, he will now provide the understanding, and will give us a single-minded and authoritative account of the subject; for it is quite certain that the ideal text on powder metallurgy has yet to be written.

J. C. CHASTON.

Symposium on the Role of Non-Destructive Testing in the Economics of Production. Presented at the 53rd Annual Meeting, American Society for Testing Materials, Atlantic City, N.J., June 27, 1950. (Special Technical Publication Number 112.) 9 × 6 in. Pp. [v] + 157, illustrated. 1951. Philadelphia (3), Pa.: American Society for Testing Materials, 1916 Race Street. ($2.50.)

This symposium, sponsored by the A.S.T.M. Committee E-7 on Non-Destructive Testing, consists of six papers, with individual discussions, and a short general discussion. The papers are: (1) " A Basic Guide for Management's Choice of Non-Destructive Tests," by R. C. McMaster and S. A. Wenk (79 pp.); (2) " Discontinuities in Cast and Wrought Products that Can be Revealed by Non-Destructive Testing," by K. R. Van Horn (18 pp.); (3) " The Role of Non-Destructive Testing in the Economics of Castings," by J. W. Juppenlatz (6 pp.); (4) " The Economics and Practical Application of Cobalt 60 in the Radiographic Inspection of Steel Weldments," by W. L. Schwinn (19 pp.); (5) " The Economics of Wrought Steel Inspection," by C. D. Moriarty (8 pp.); (6) " Management's Responsibility for Insistence on Non-Destructive Testing in the Development of New Engineering Products and Processes," by L. W. Ball (4 pp.).

In an introduction to these papers, the Chairman of the Symposium Committee, H. H. Lester, states that their main object is to draw the attention of those concerned with management to the value of non-destructive testing in the maintenance of quality and economy in production. The first paper comprises approximately half the symposium, and is essentially a survey of the literature of non-destructive testing classified in twelve sections. Each section is introduced by a short description of a principle of operation which is later illustrated by references to particular methods of testing. In addition to a bibliography appropriate to each section, information relating to United States patents is tabulated. The value of this collection of references to the literature as a guide would have been greatly enhanced by a critical consideration of the material to which reference is made.

Van Horn's paper refers particularly to discontinuities detectable by radiography, fluorescent penetrants, magnetic particles, and ultrasonic reflection. The value of these methods is generally established, and the present paper is a useful record of their fields of application and their limitations. It contains an interesting series of photographs showing classified types of defects.

The relative costs of visual inspection and non-destructive tests are compared in the third paper, which, although short, comments usefully on the extent to which non-destructive testing is justifiable economically. Schwinn's is a technical paper comparing cobalt 60 as a source of gamma-rays with radium and X-ray generators of 250–2000 kV. and seems rather out of place in this symposium. The fifth paper is directly relevant to the title of the symposium, and deals with the importance of applying non-destructive tests at the correct stage in a manufacturing cycle to obtain economic advantages. Finally, Ball comments briefly on the economic advantages to be obtained by the use of non-destructive testing in the development of new products.

Considering the symposium as a whole, the reviewer is of the opinion that the emphasis on economics and management in the general title and in some of the titles of the papers is not entirely justified.

R. F. HANSTOCK.

Casting of Brass and Bronze: Some Practical Aspects of Brass and Bronze Casting in America, 1900–1950. By Daniel R. Hull. Med. 8vo. Pp. [vi] + 186, illustrated. 1950. Cleveland (3), O.: American Society for Metals. ($3.50.)

This book, as its author points out, does not constitute a review of the literature of the subject, but is an account based on his own experience. Although the sub-title indicates that the volume is concerned with some practical aspects of brass and bronze casting in America from 1900 to 1950, it is none the less of interest and value to British readers. This book is no mere compilation of the technical details involved in casting brasses and bronzes, but is a very readable treatment of the subject enriched by the author's own informed views and opinions.

In the 177 pages of usefully illustrated text, divided into nine chapters, the author has succeeded in this relatively small compass not only in covering the developments which have taken place in techniques over the years, but also in dealing critically and constructively with the many issues involved. After an interesting first chapter on "Things as They Were", two following chapters are devoted to melting and pouring. Tin bronze is given a chapter to itself, and oxygen, hydrogen, and deoxidizers are considered in another. As might be expected in dealing with such a subject, a proportionately lengthy space is given over in two chapters to moulds and mould dressings, and the author concludes with some random observations and an expression of opinion on future trends. To any one interested in or concerned with the casting of brasses and bronzes this book can be unhesitatingly recommended as a most readable and instructive account of this branch of metallurgy.

Maurice Cook.

Kurs Odlewnictwa [Foundry Course]. By Kazimierz Gierdziejewski. 2nd edition. (Biblioteka Podręczników Akademickich). 24 × 17 cm. Pp. xxxvi + 640, with 325 illustrations and 87 tables. 1950. Warsaw: Spółdzielnia Wydawniczo-Oświatowa "Czytelnik". (1200 złoty.)

This is a text-book on metal-melting procedures and melting furnaces, essentially for the university student and the industrial technician or technologist.

The present edition is a considerably enlarged and revised version of the original book published in 1930, and should be regarded as the first volume of a much larger work covering all aspects of foundry practice. It opens with an introductory chapter in which the author outlines the purpose and the principal theme of his work and briefly surveys the history of metal founding. It is further divided into two principal parts: the first (108 pp.) deals with raw materials, fuels, and refractories used in iron, steel, and non-ferrous foundries, with emphasis on the iron foundry; the second (over 500 pp.) describes the melting of alloys and the furnaces employed, again with emphasis on iron-melting processes and furnaces.

The section on metallic raw materials gives the characteristics and properties of various kinds of pig iron, copper, aluminium, tin, lead, zinc, nickel, magnesium, and ferro-alloys. Scrap for iron and steel and non-ferrous foundries is also described and classified.

A separate chapter is devoted to the proper preparation and calculation of the cupola charge. The fuel section deals chiefly with foundry coke; and other fuels such as pulverized coal and liquid and gas fuels are discussed only briefly. The chapter devoted to refractories is very short and gives only the principal characteristics and classification of the best-known materials.

The principal subject of the second part of the book is cupola practice and iron-melting processes, which are described and explained at considerable length, with not only the theoretical background and physical chemistry of the processes involved, but also with a critical review of cupola design, charging methods, air-blast equipment, &c.

The succeeding chapters are devoted to crucibles, reverberatory, rotary, rocking, electric, arc, high- and low-frequency furnaces, and convertors, and are written from the point of view of the needs of the small and medium-size foundry. The advantages and disadvantages of the various types of crucible furnaces are described, including the Morgan melting furnaces. Among the reverberatory and rotary furnaces the Brackelsberg and Sesci rotary furnaces are dealt with at some length, and their application to the iron-founding industry is outlined and discussed.

A special chapter is given to Professor Grum-Grzymajlo's "Hydraulic Theory of Gas Flow" and to the application of this theory to the design of melting furnaces.

The last section of the book gives a brief survey of electric furnaces (arc and high- and low-frequency) commonly used for melting in foundries. At the end of this chapter there are examples of heat balances, and a comparison of the melting costs for different types of foundry furnaces.

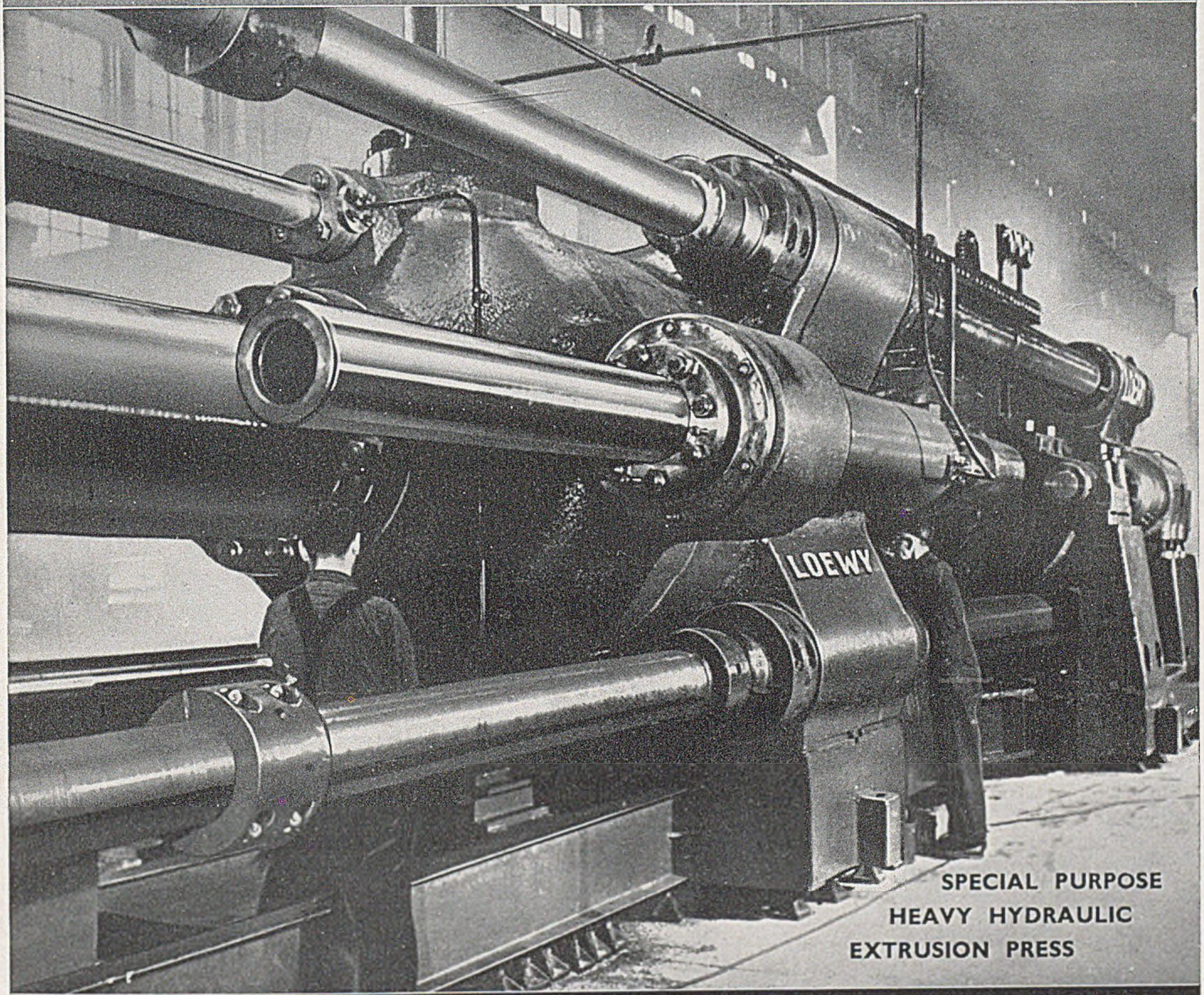
Each chapter of the book has a comprehensive bibliography, but this unfortunately covers only the period up to 1940.

T. B.

Konstrukteur und Giesser. Sonderausgabe der Zeitschrift *Giesserei*. 29·5 × 21 cm. Pp. 88, illustrated. 1951. Düsseldorf: Giesserei-Verlag, August-Thyssen-Strasse 1. (DM 7.20.) [For contents see 24—*Bibliography*, col. 87.]

The statement that "the success of a casting is decided on the drawing-board"—if one deducts the usual percentage allowance on broad generalizations—is certainly a good guiding principle in the manufacture of shaped castings. The great difficulty lies, however, in presenting the problem of relation of the designer to that of the foundryman in a scientific manner. Although the present series of papers in *Giesserei* just falls short in this respect, nevertheless, the treatment is very stimulating and interesting. No distinctly new material is included, but the subject is brought reasonably up-to-date and presented in a concise and logical manner. Designers of castings, foundrymen, and users of cast products will find the reading of the papers both useful and refreshing.

V. Kondic.



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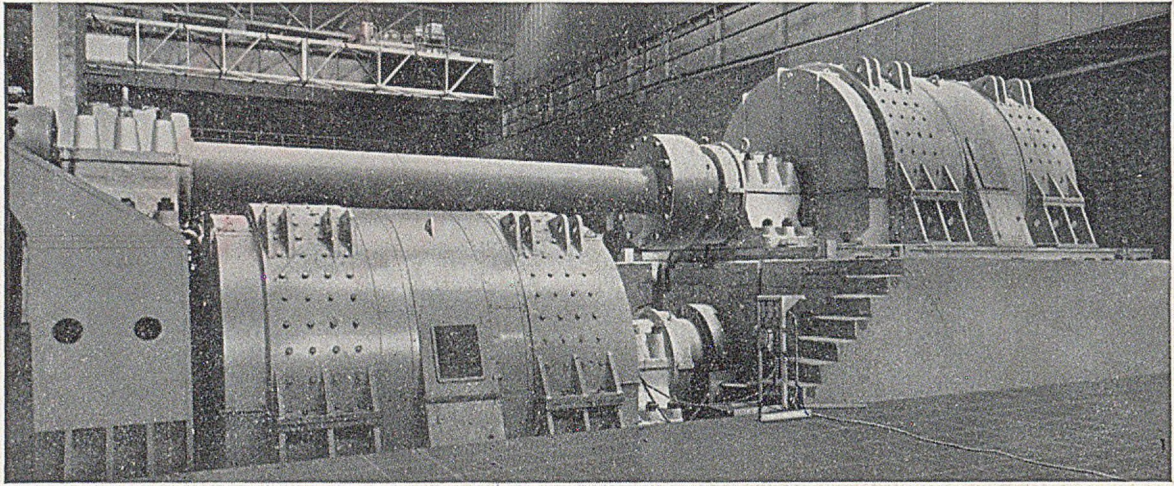

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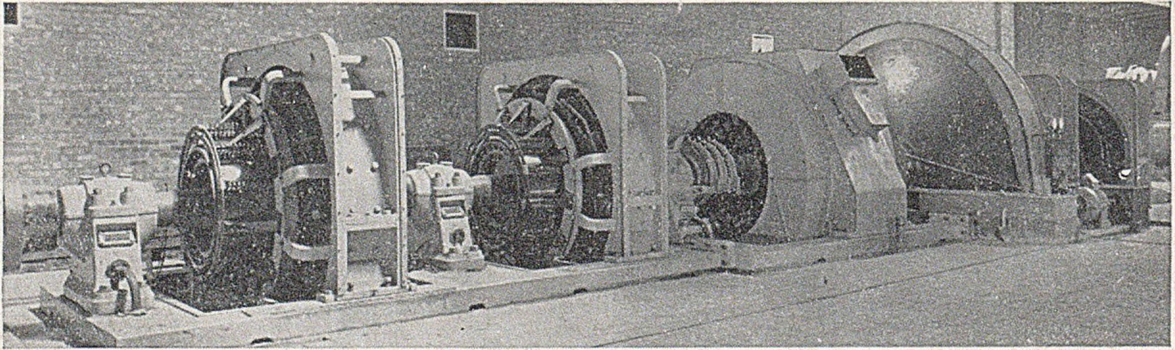

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