THE JOURNAL OF THE INSTITUTE OF Metals

WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents:

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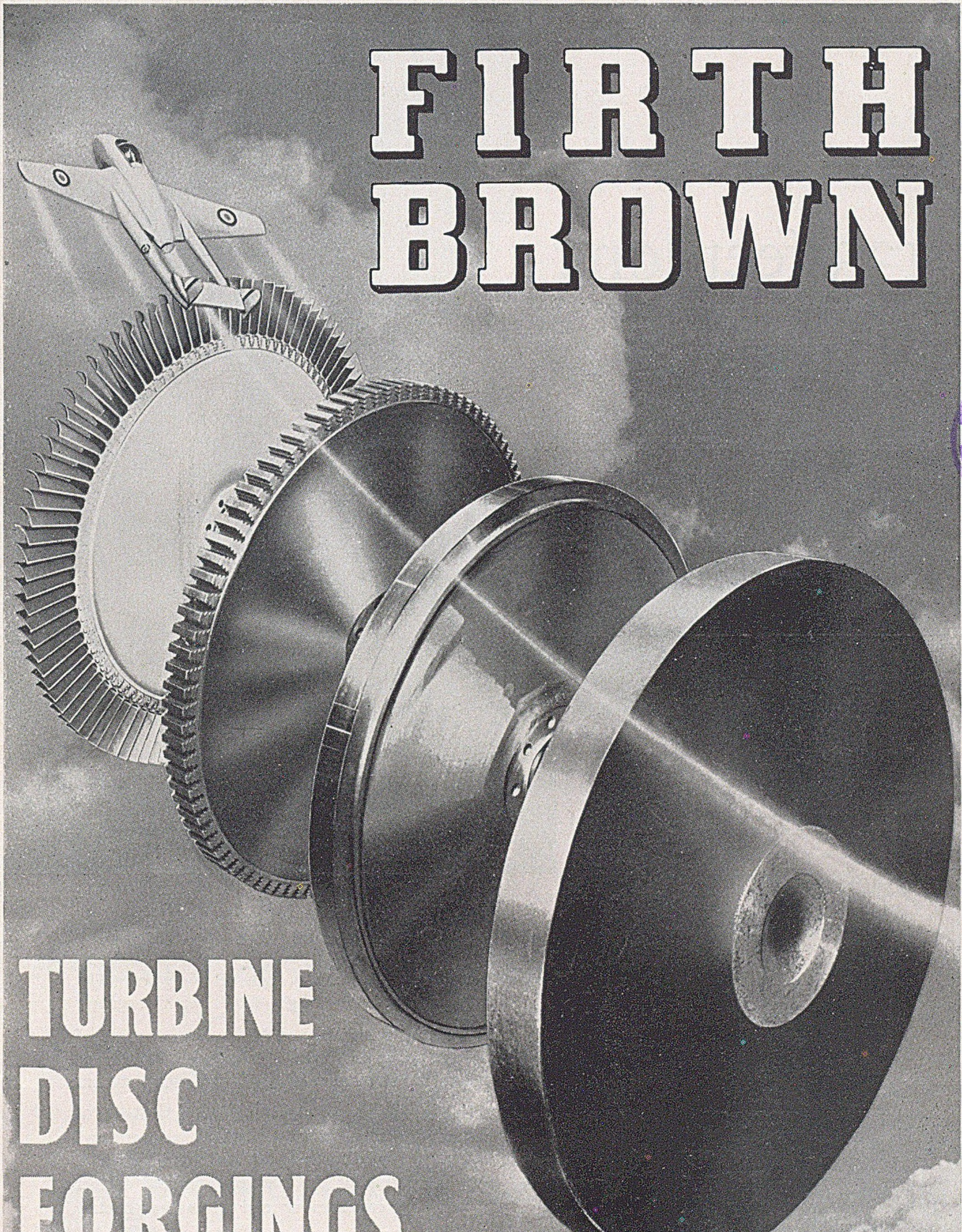
MAY 1953



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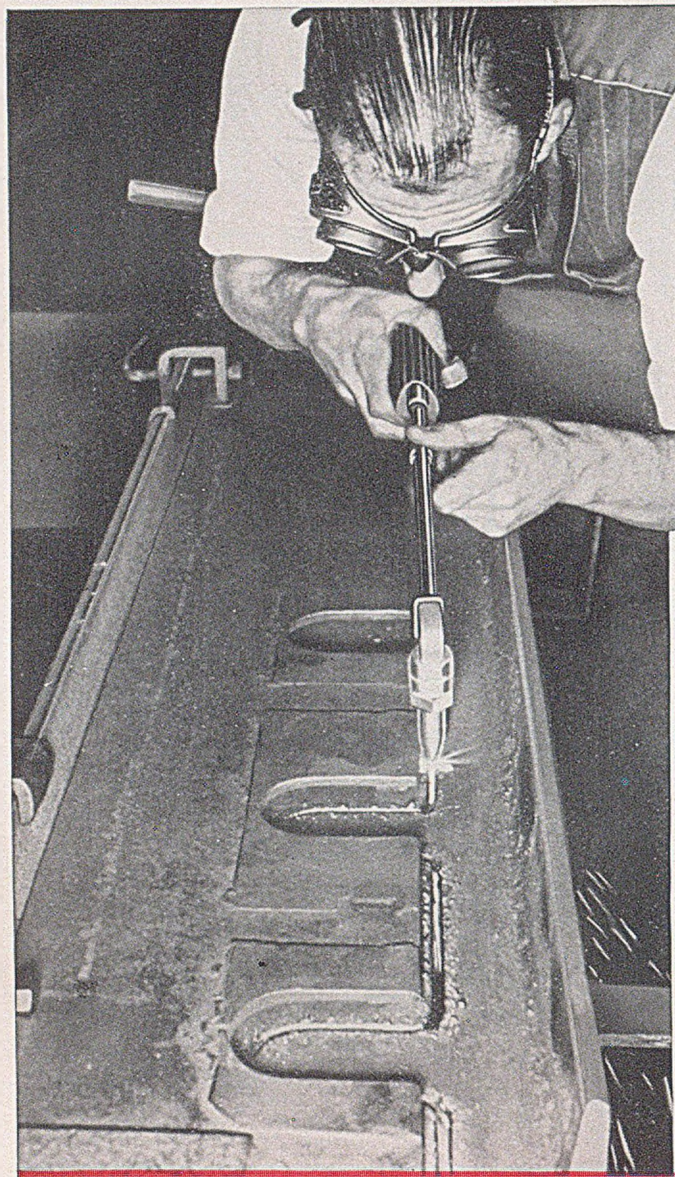

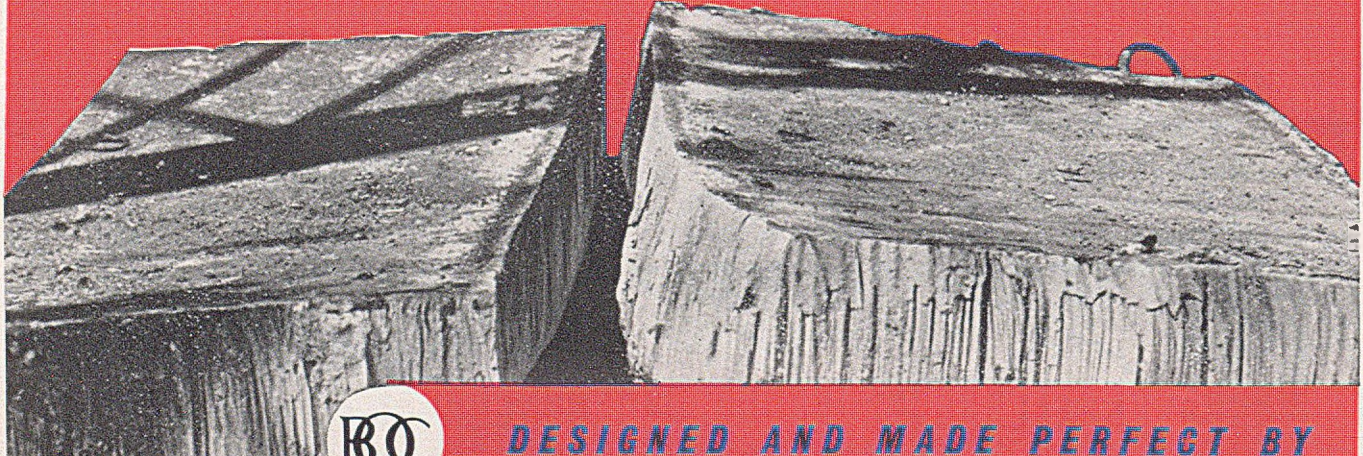

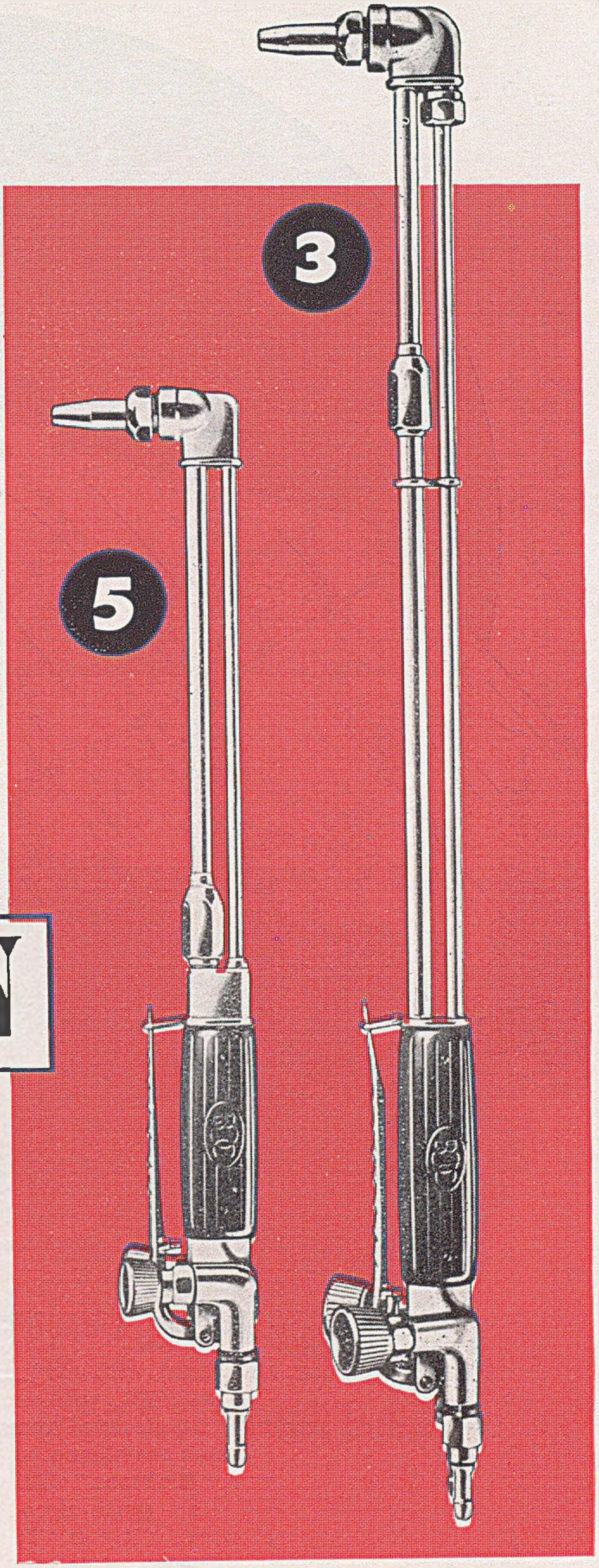
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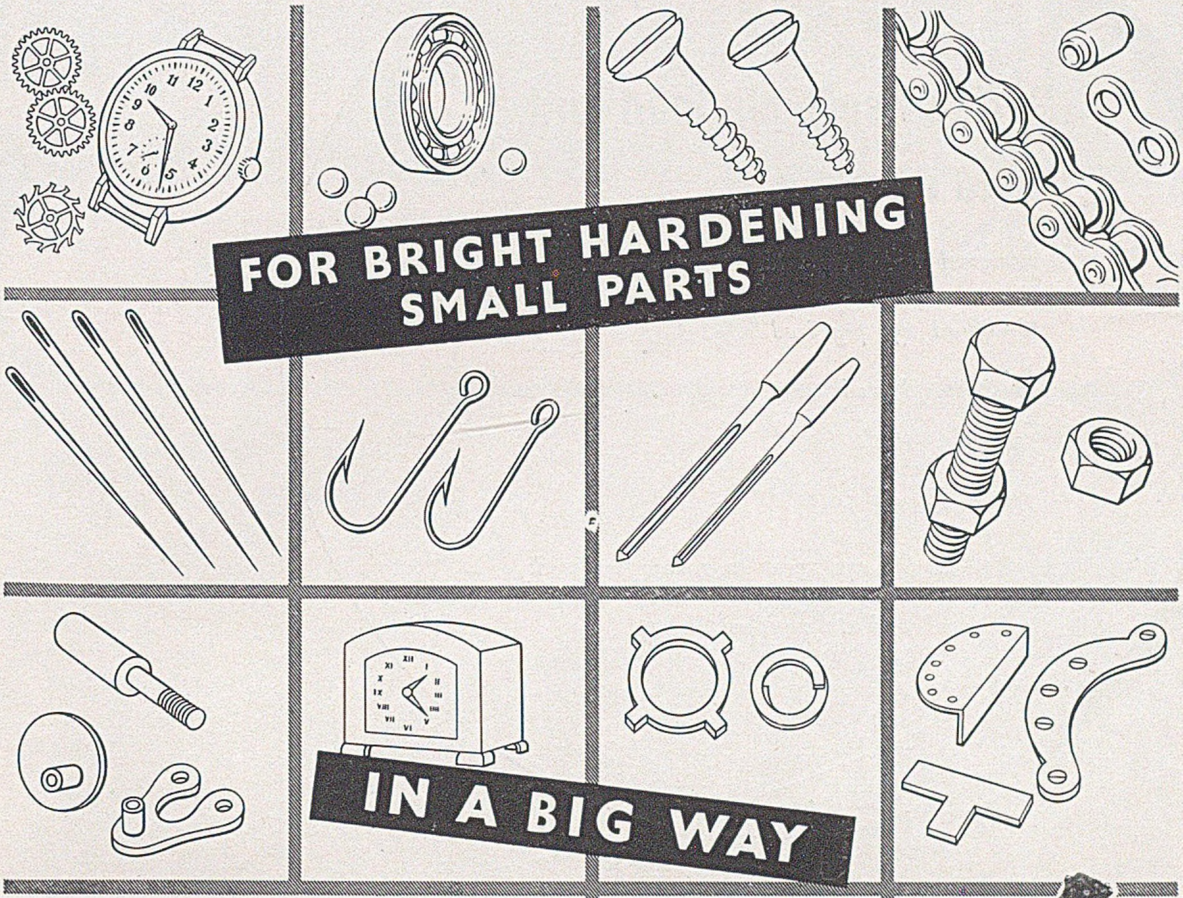

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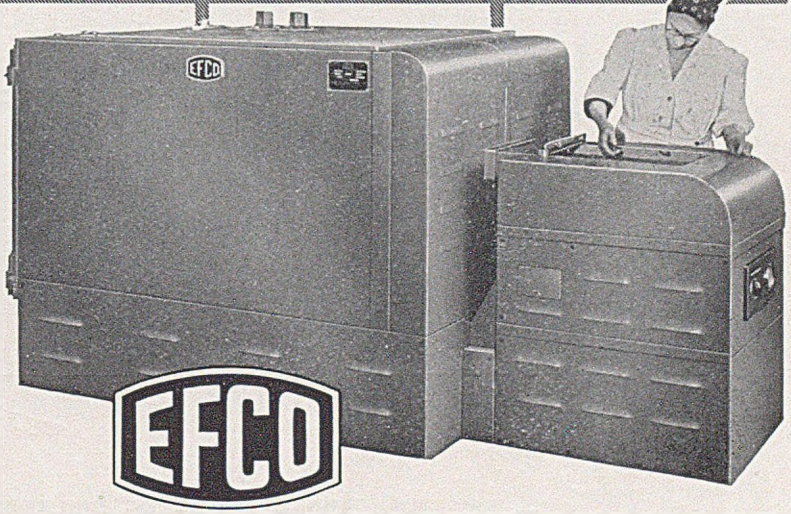

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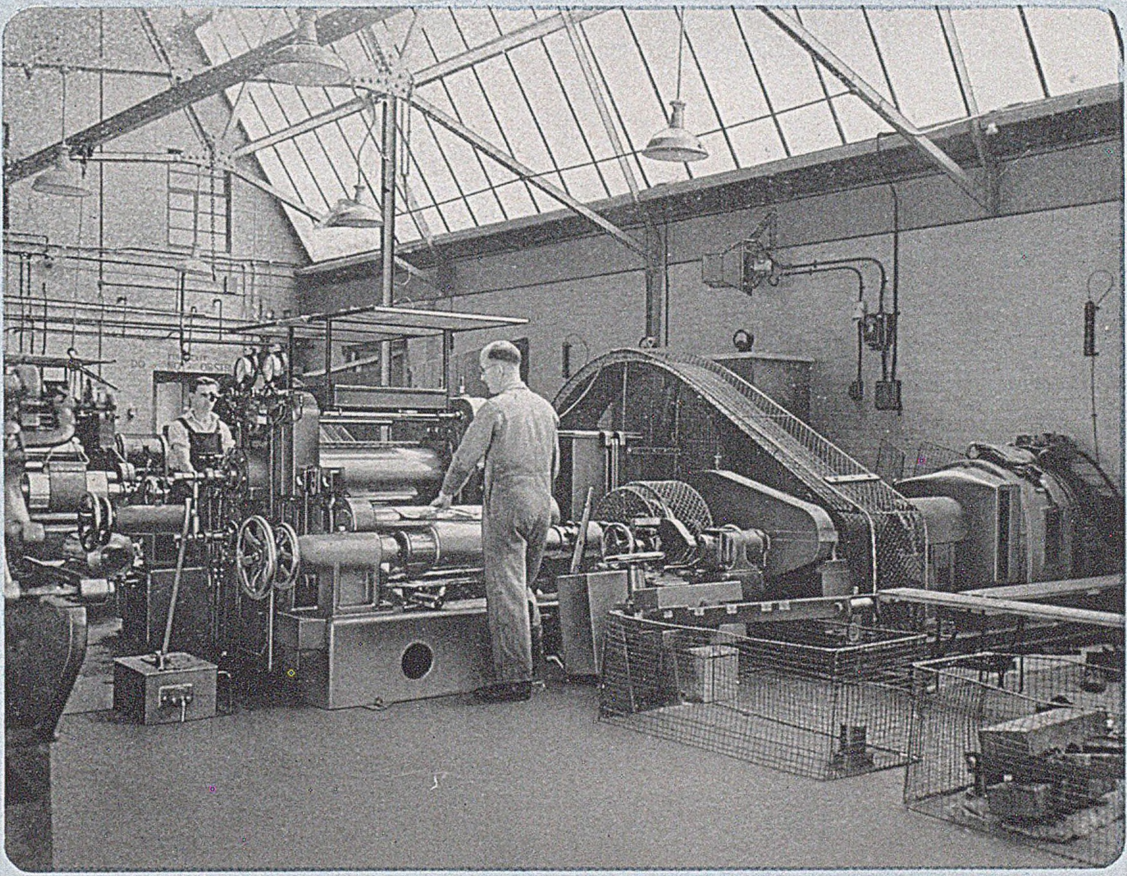

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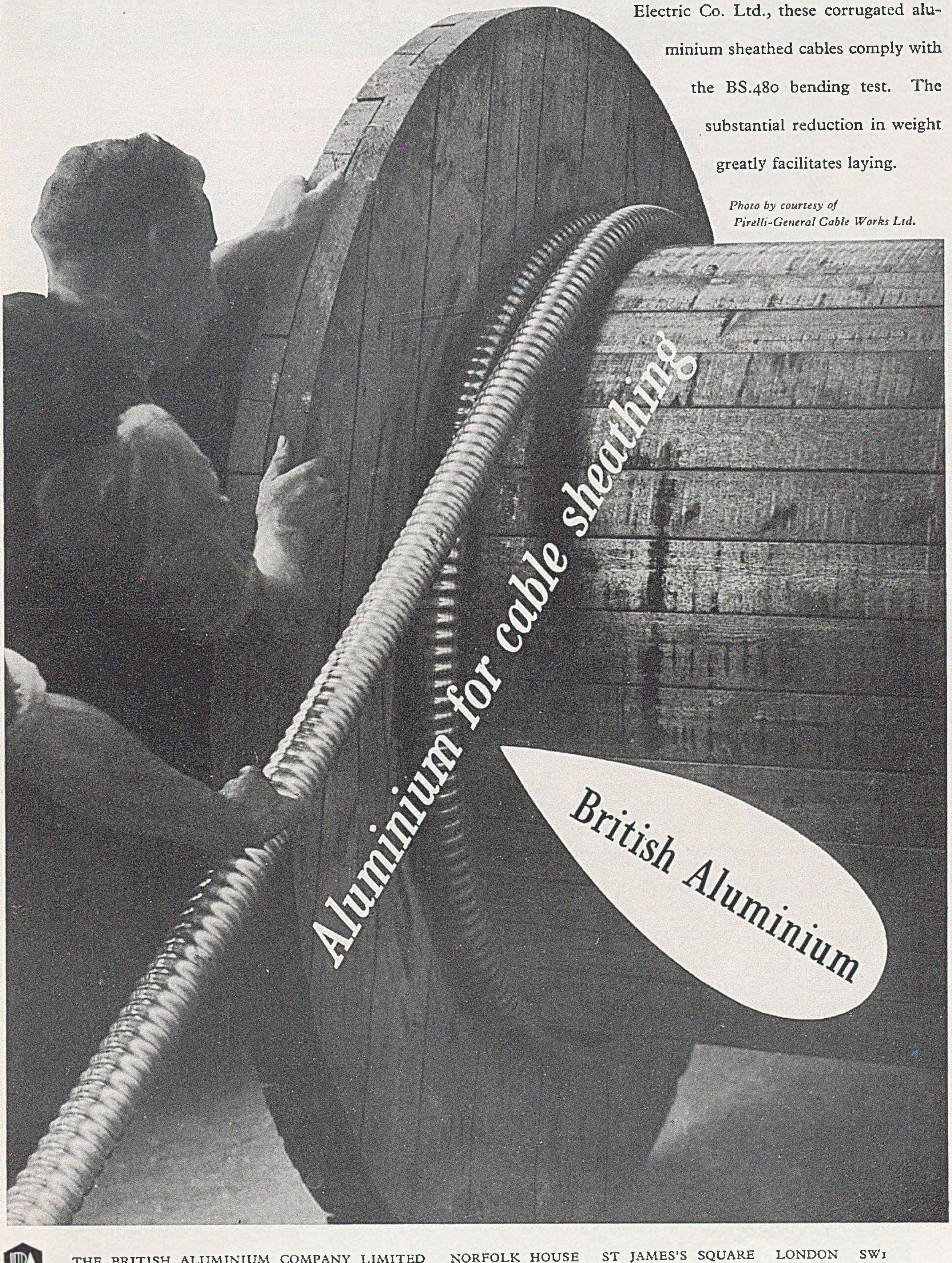
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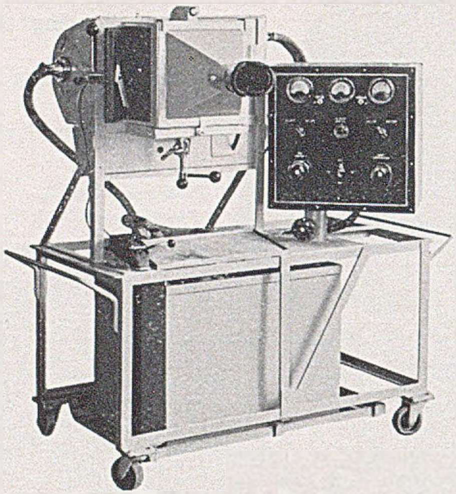

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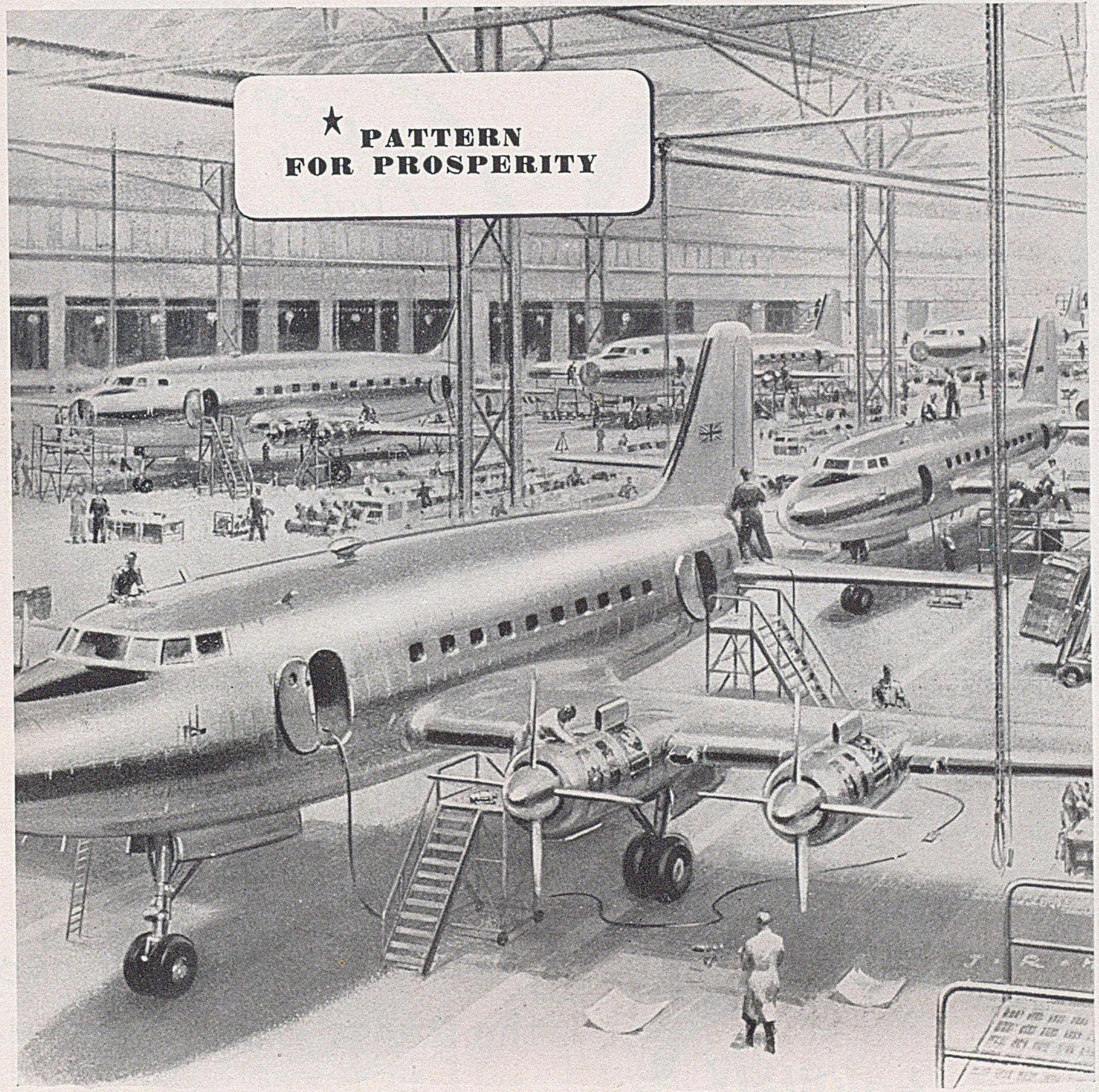
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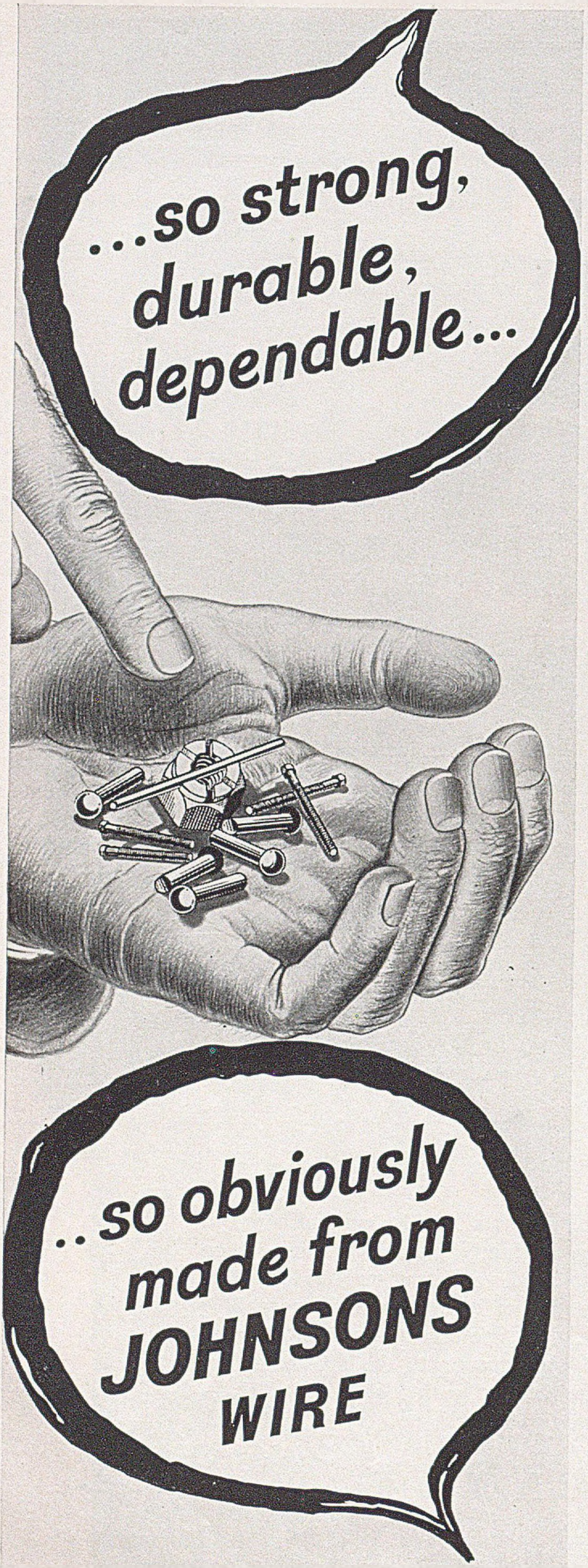
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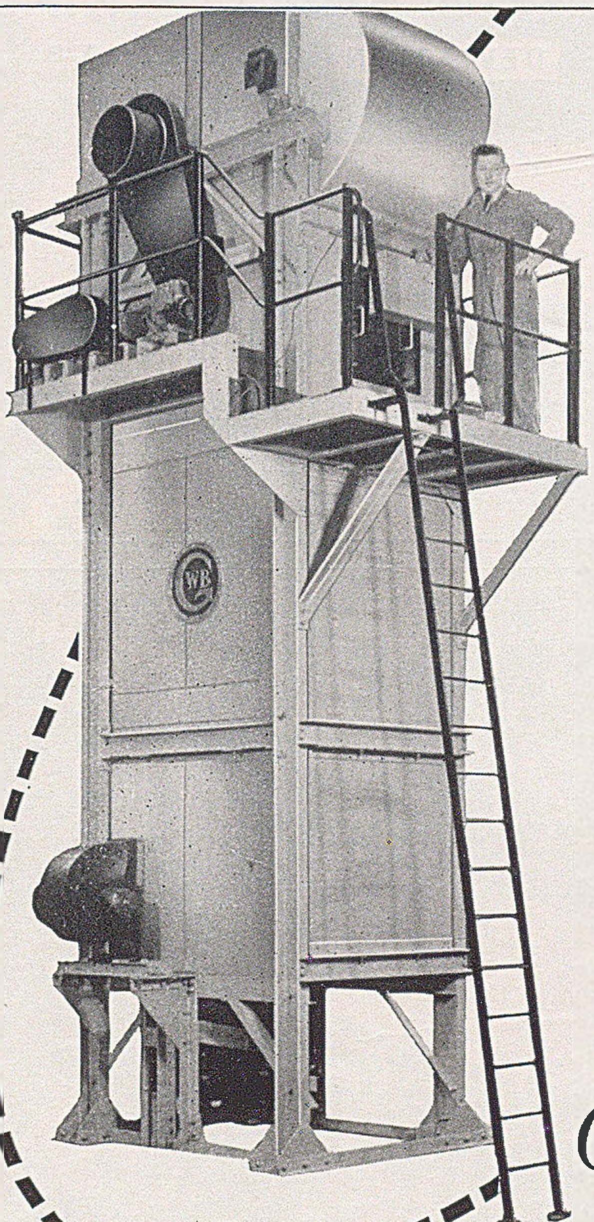

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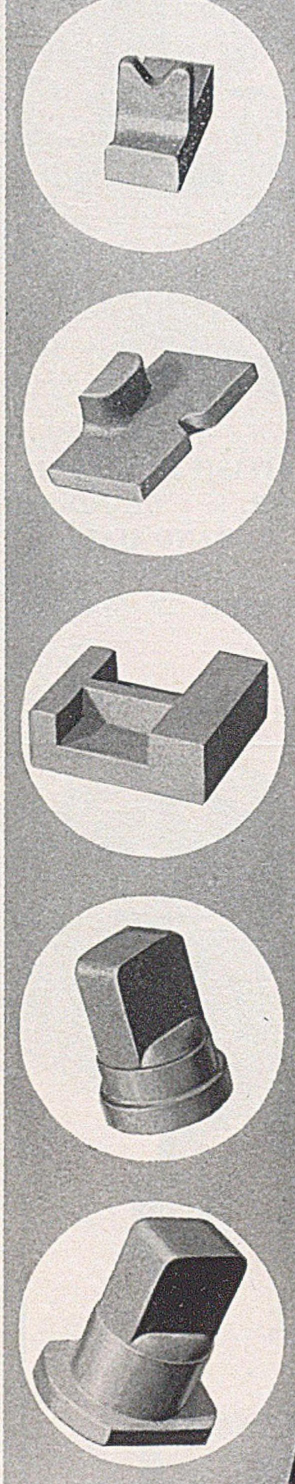

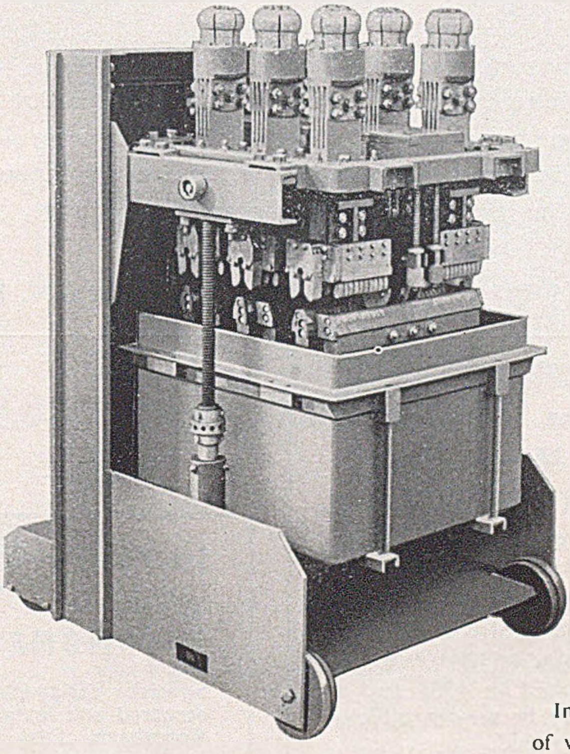

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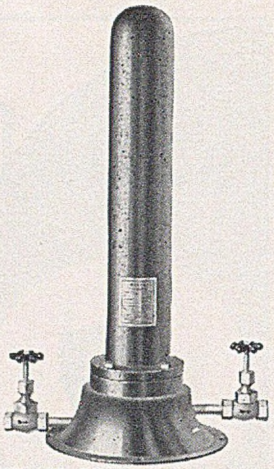

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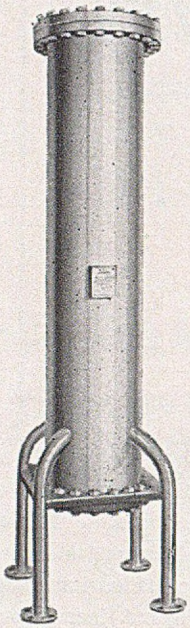

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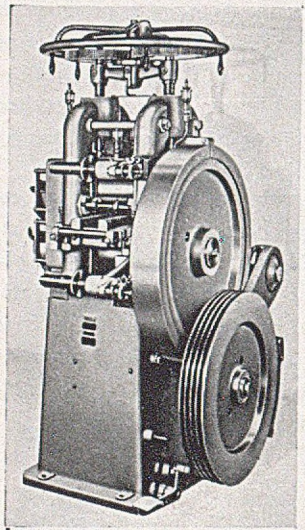

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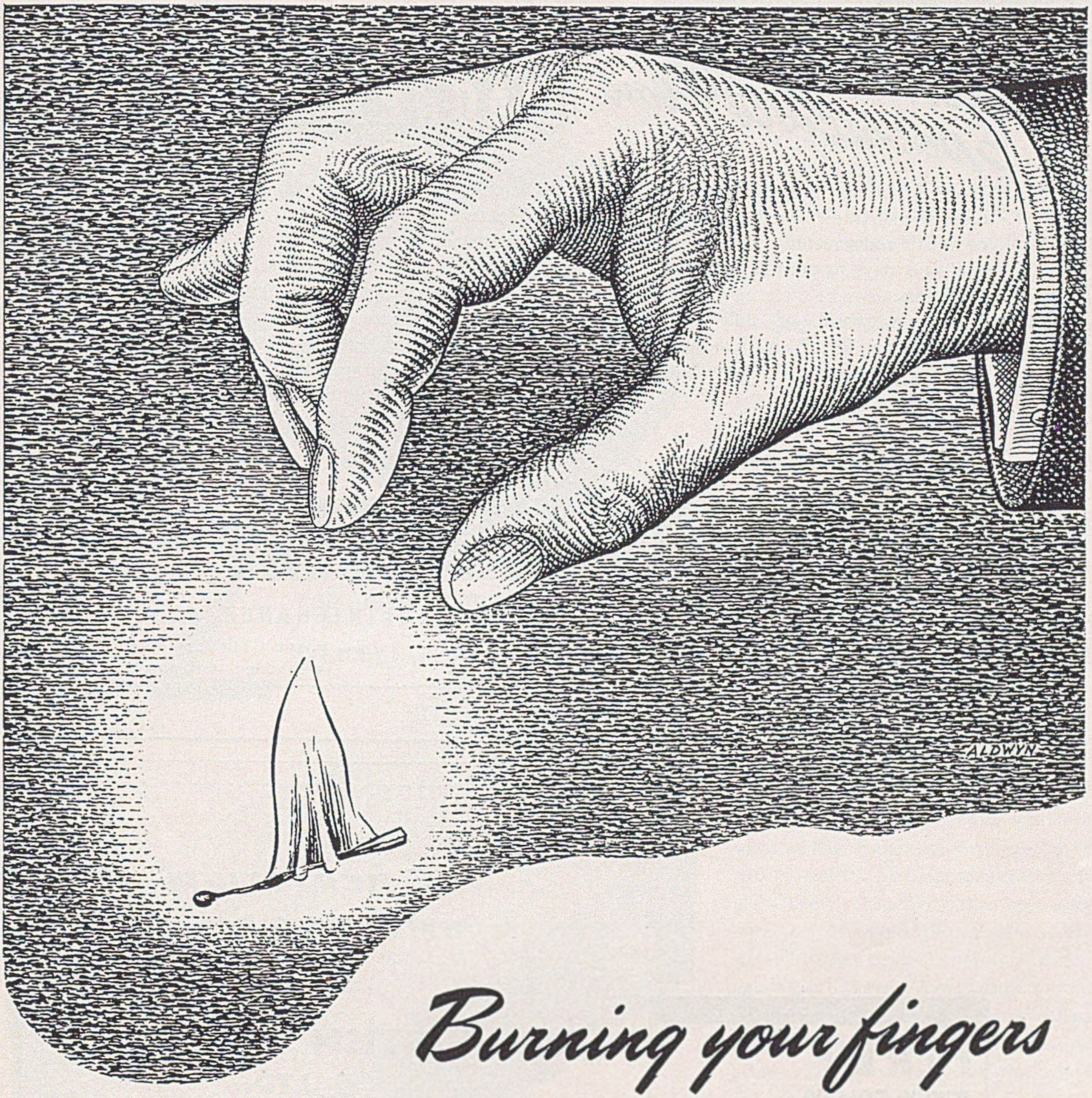
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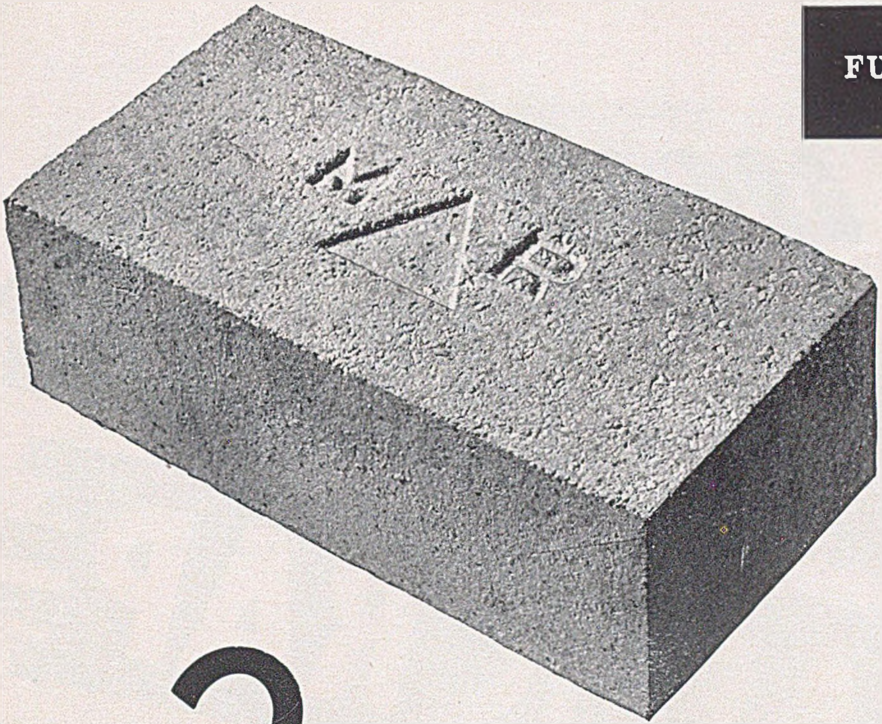
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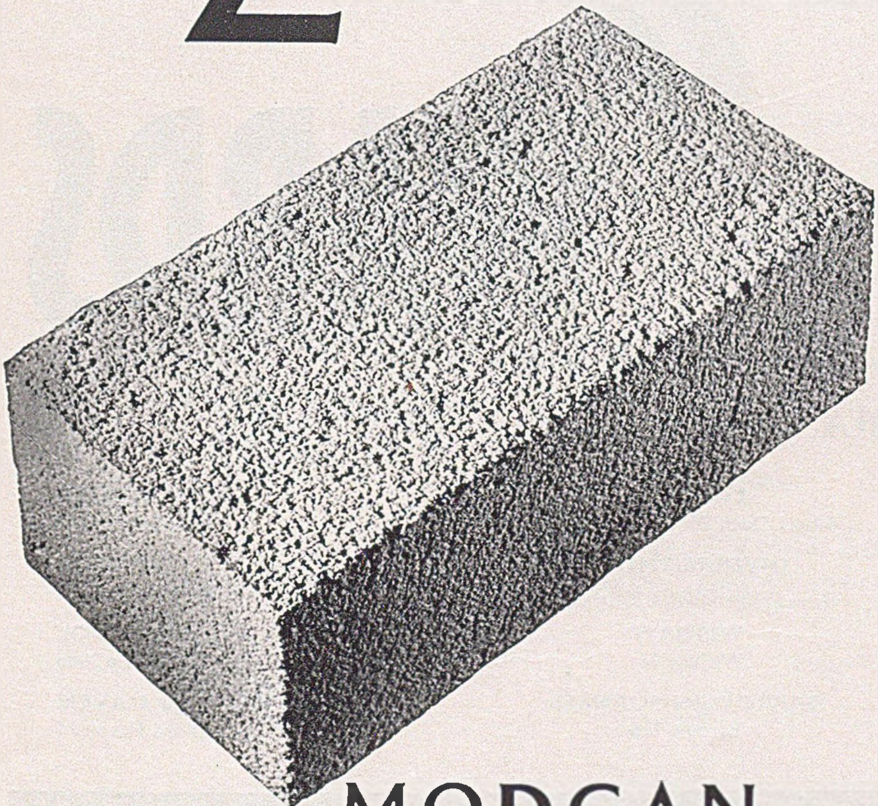

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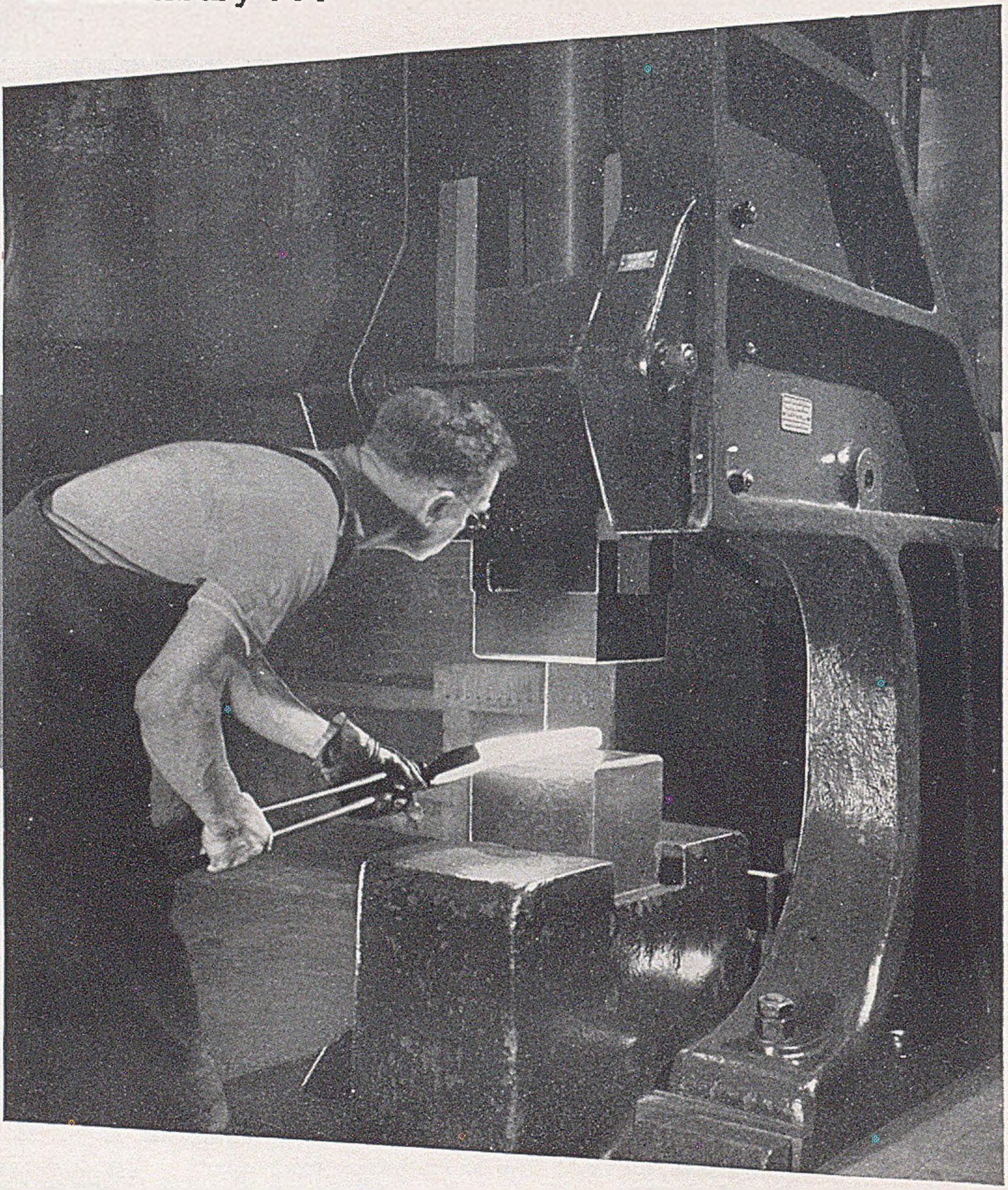

JOURNAL OF THE INSTITUTE OF METALS

WITH THE BULLETIN AND METALLURGICAL ABSTRACTS

SUBSCRIPTIONS
Members: Free

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

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4 Grosvenor Gardens, London, S.W.1
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1953, PART 9 PUBLISHED MONTHLY MAY

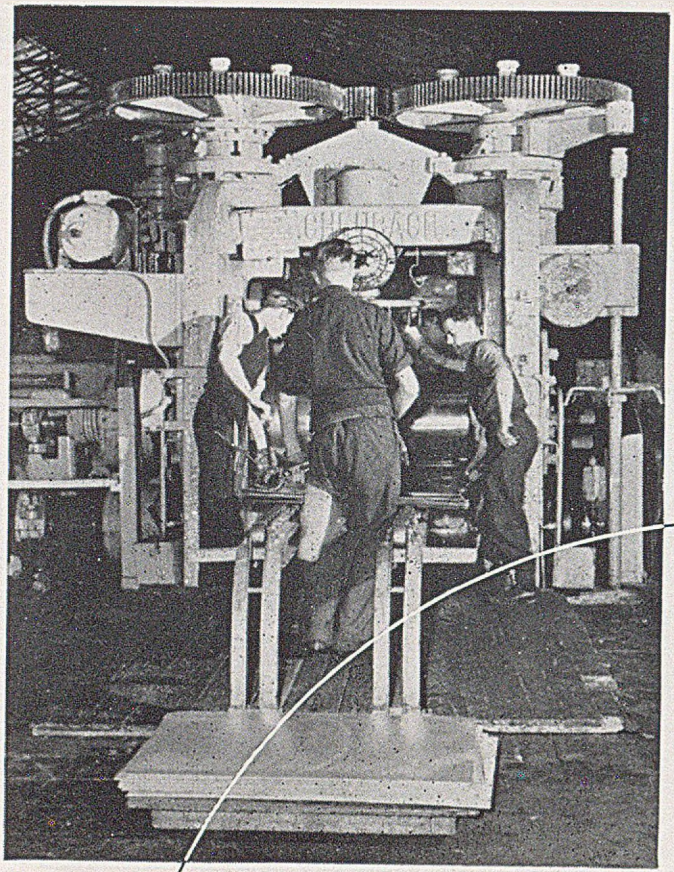
CONTENTS

BULLETIN

METALLURGICAL ABSTRACTS

JOURNAL

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



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BULLETIN

OF THE INSTITUTE OF METALS

VOLUME I MAY 1953 PART 21

INSTITUTE NEWS

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

As previously announced, the 1953 Autumn Meeting will be held in South Lancashire, by invitation of the Councils of the Manchester Metallurgical Society and the Liverpool Metallurgical Society. Because of hotel accommodation problems in Chester, Liverpool, and Manchester, the meeting will be held at Southport, where the accommodation for such conferences is adequate and from which a wide range of visits is possible.

Reception Committee

The following have agreed to serve on the Reception Committee:

Lieut.-Commander G. K. Rylands, *Chairman* (Rylands Brothers, Ltd., Warrington).
Dr. W. E. Alkins (Thomas Bolton and Sons, Ltd., Stoke-on-Trent).
Mr. H. Allison (Metropolitan-Vickers Electrical Co., Ltd., Manchester).
Dr. S. F. Barclay (Mather and Platt, Ltd., Manchester).
Mr. Eric Bowyer (British Insulated Callender's Cables, Ltd., Prescot).
Mr. C. F. Brereton (Richard Johnson and Nephew, Ltd., Manchester).
Mr. R. S. Brown (Rylands Brothers, Ltd., Warrington).
Mr. G. T. Callis (Manganese Bronze and Brass Co., Ltd., Birkenhead).
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Mr. C. T. Davey (Thomas Bolton and Sons, Ltd., Widnes).
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Mr. I. W. F. Fairholme (Richard Johnson and Nephew, Ltd., Manchester).
Mr. W. L. Govier (Imperial Chemical Industries, Ltd., Metals Division, Kirby).
Capt. A. C. Jessup (Magnesium Elektron, Ltd., Clifton Junction).
Dr. W. L. Kent (The British Aluminium Co., Ltd., Warrington).
Mr. I. S. Kerr (The Lancashire Steel Corporation, Ltd., Irlam).
Mr. L. Rotherham (Ministry of Supply, Risley).
Mr. F. R. Smith (The English Electric Co., Ltd., Stafford).
Professor F. C. Thompson (University of Manchester).
Mr. W. T. T. Wilcox (Shell Refining and Marketing Co., Ltd.).

Mr. S. V. Radcliffe (Rylands Brothers, Ltd.) has agreed to act as Honorary Secretary to the Reception Committee.

Provisional Programme

The following provisional programme has been approved:

Mon., 21 Sept.
Aft. Members and guests register.
Evg. Autumn Lecture, by Dr. Maurice Cook, in the Town Hall.

Tues., 22 Sept.
Morn. Official Welcome to Members by the Mayor of Southport.
Scientific Session "A" (papers on subjects relating to industrial metallurgy), side by side with
Scientific Session "B" (physical metallurgy).
Aft. Visits to works and laboratories and Ladies' visits.
Evg. Civic Reception, with dancing, at the Town Hall, by invitation of the Mayor of Southport.

Wed., 23 Sept.
All-day. Visits to works and laboratories located at a considerable distance from Southport (e.g. in the Manchester area) and Ladies' visit to Chester.
Evg. Banquet at the Prince of Wales Hotel.

Thurs., 24 Sept.
Morn. Scientific Session "A" (papers of general interest), side by side with
Scientific Session "B"—Informal Discussion on "Damping Capacity", to be arranged by the Metal Physics Committee.
Aft. Visits to works and laboratories and Ladies' visits.
Evg. Free.

Fri., 25 Sept.
All-day. Tour to the Lake District.

Accommodation

Accommodation for members and their guests has already been reserved at the Prince of Wales Hotel and the Brunswick Hotel, at Southport, particulars of which are given below. It has not been found possible to book inexpensive accommodation in Southport for students, but arrangements have been made by which students and other members may stay at Derby Hall (a University hostel) at Liverpool, travelling to Southport by train (20 minutes) daily. The station at Southport is close to the place of meeting and the centre of the town. Particulars are given below. The hotel marking is the Automobile Association grading.

**** **Prince of Wales Hotel**, Lord Street, Southport. Inclusive daily terms: 39*s.* 6*d.* or 42*s.* per person, according to position of room. Private bathroom, 7*s.* 6*d.* extra. Some private suites are available. Bed and breakfast terms, daily:

30*s.* per person. An allowance of 5*s.* will be made to members attending the Banquet. The hotel is in first-class order and can be confidently recommended.

Brunswick Hotel, Lord Street, Southport (A.A. approved). Rooms have been reserved in the hotel, with additional rooms at nearby houses, with the amenities and all meals in the hotel. Inclusive daily terms: 28*s.* 6*d.* per person sleeping in the hotel; 26*s.* 6*d.* for those sleeping out of the hotel.

Derby Hall, North Mossley Hill Road, Liverpool. Daily charges for bed and breakfast: Bona fide students, 10*s.* per day; senior members, 14*s.* per day. Dinner (if ordered), 3*s.* 6*d.*

Bookings of hotel accommodation can now be requested and, in the case of the reserved accommodation mentioned above, *must be made with the Institute and not with the hotel or hostel.* The Institute must pay a deposit on rooms booked at Derby Hall. Accommodation at Derby Hall is also being made available to other conferences, and members wishing to stay there should, therefore, book at the earliest possible opportunity, so that rooms may be reserved for them.

In an Annexe to Derby Hall, 12 double rooms are available. Members requiring them should apply with the least possible delay.

General Information

Particulars of the programmes of papers to be discussed and works and laboratories to be visited will be published as early as possible. It is expected that a full programme, with Reply Forms, will be posted in July to members resident in Europe. Members resident outside Europe who hope to attend the meeting should apply to the Secretary to receive a copy of the full programme (which will also be published in the *Bulletin*) and the Reply Form.

Election of Members

The following 27 Ordinary Members, 1 Junior Member, and 11 Student Members were elected on 23 March 1953:

As Ordinary Members

BOYS, Sydney Joseph Henry, L.I.M., Chief Metallurgist, Thompson Brothers (Bilston), Ltd., Bilston, Staffs.

BRAITHWAITE, Wilfrid Rolf, B.Met., Metallurgist, Magnesium Elektron, Ltd., Clifton Junction, near Manchester.

BRUCE, Fraser W., Managing Director, Northern Aluminium Co., Ltd., Bush House, London, W.C.2.

BULL, Harry, Assoc.Met., F.I.M., Managing Director, Brown Bayley Steels, Ltd., Leeds Road, Sheffield 9.

BULL, James Reginald Ashworth, Managing Director, Sanderson Brothers and Newbould, Ltd., Attercliff Works, Newhall Road, Sheffield.

COUSINS, (Miss) Kathleen Marjorie, A.I.M., Metallurgist, Rubery, Owen and Co., Ltd., Darlaston, Staffs.

DALTON, Charles Arthur, B.Sc., A.I.M., Chief Metallurgist, Armstrong Siddeley Motors, Ltd., Parkside, Coventry.

DESOER, Adrien, Ing. Civil des Mines, Directeur Général Adjoint, Société Anonyme d'Ougrée-Marihaye, Ougrée (Liège), Belgium.

DÜRRWÄCHTER, Eugen, Dr.Ing., Direktor, Platinschmelze Dr. E. Dürrwächter, Postfach 118, Pforzheim, Germany.

GORE, James, Technical Superintendent, British Insulated Callender's Cables, Ltd., Prescot, Lancs.

HASELGROVE, J., Chief Metallurgist and Assistant Works Manager, United Wire Works (Birmingham), Ltd., Adderley Street, Birmingham.

HAWKINS, Charles Cecil, Director, Mark Perks, Ltd., 43 Vyse Street, Birmingham 18.

HENSHAW, George Stanley, A.C.T., Development Engineer, T. I. Aluminium, Ltd., Birmingham 11.

HODIERNE, Francis Arthur, B.Sc., A.I.M., Research Metallurgist, Department of Development and Research, T.I. (Group Services), Ltd., Plume Street, Birmingham 6.

KEIL, Albert, Dr.phil.nat., Leiter des Forschungs-Laboratoriums, Platinschmelze Dr. E. Dürrwächter, Postfach 118, Pforzheim, Germany.

LUELING, Henry, Dipl. Ing., Research and Development Engineer, George Fischer, Ltd., Schaffhausen, Switzerland.

MULLINS, James Edgar, B.Eng., Graduate Apprentice Metallurgist, D. Napier and Son, Ltd., East Lancashire Road, Liverpool.

ORLANDO, Attilio, Dott. Ing., In charge of Rolling Department, Società Metallurgica Italiana, Fornaci di Barga (Lucca), Italy.

PAGE, Frederick Vernon, Technical Representative, London Electric Wire Company and Smiths, Ltd., 31 Spenser Street, London, S.W.1.

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SMITH, Edward William, Manager, Mark Tyzack and Sons, Ltd., Carver Street, Sheffield 1.

TALLIS, Charles Edward, Group Development Engineer, Denbro, Ltd., 222 Soho Hill, Birmingham 19.

TRACEY, Lewis Johnson, Managing Director, Stock Sons and Taylors, Ltd., 19 Gas Street, Birmingham 1.

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YOUNGKRANTZ, Howard R., B.S., Chief Engineer, Apex Smelting Company, 2537 West Taylor Street, Chicago 12, Ill., U.S.A.

As Junior Member

KNOWLES, Allan John, Senior Analyst, Elkington and Co., Ltd., Goscote Works, Goscote, near Walsall, Staffs.

As Student Members

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FLETCHER, Harry Olwyn, Metallurgical Student, Stewarts and Lloyds, Ltd., Bilston, Staffs.

GUPTA, B. K., M.Sc., Trainee, Enfield Rolling Mills, Ltd., Enfield, Middx.

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PIERCY, George Robert, M.A.Sc., Research Student, Department of Metallurgy, University of Birmingham.

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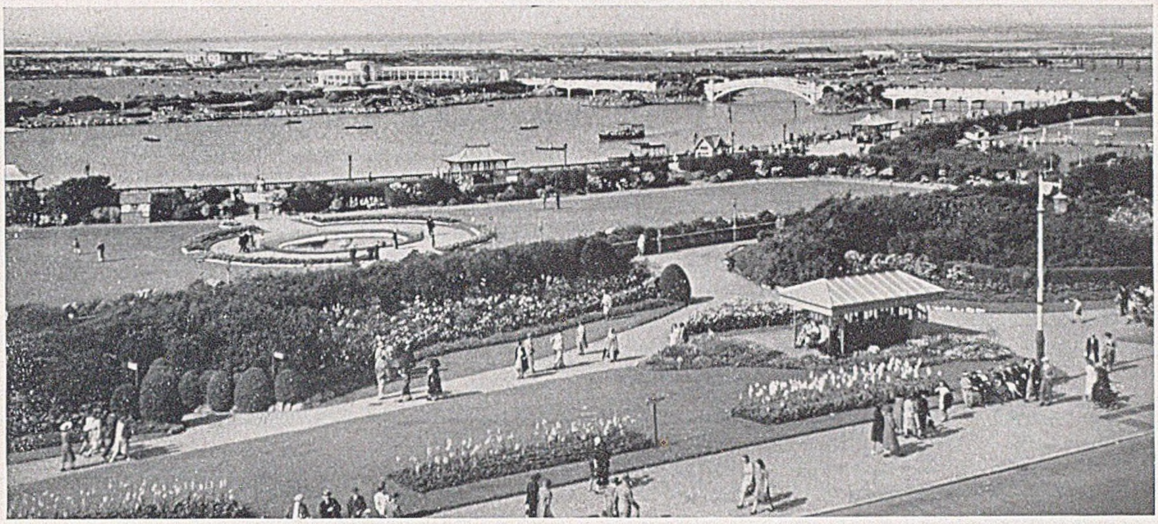
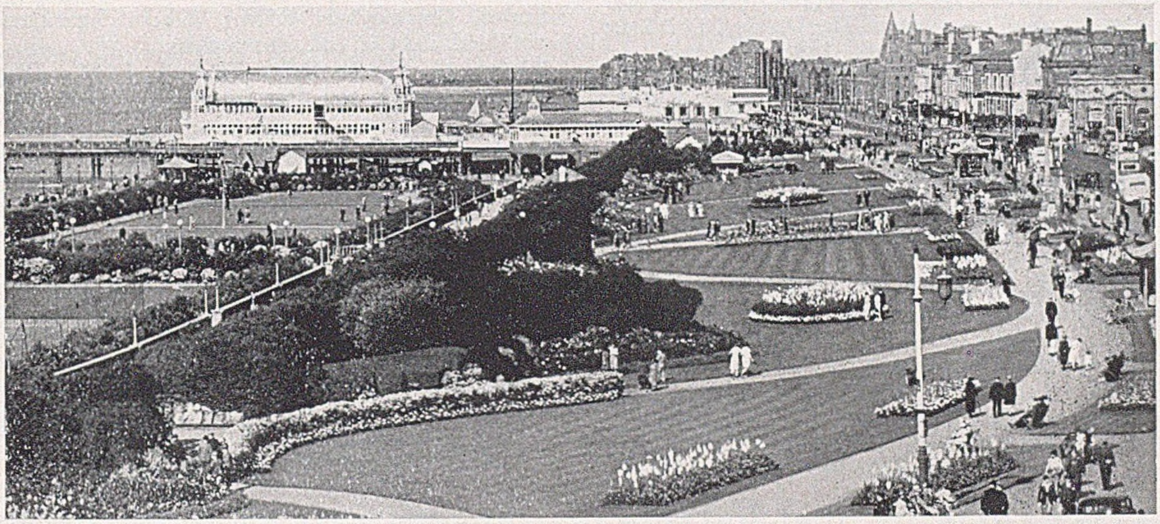
WILLIAMS, (Miss) Joan, B.Sc., Chemist, Richard Thomas and Baldwins, Ltd., Wern Aluminium Works, Briton Ferry, Neath, Glam.

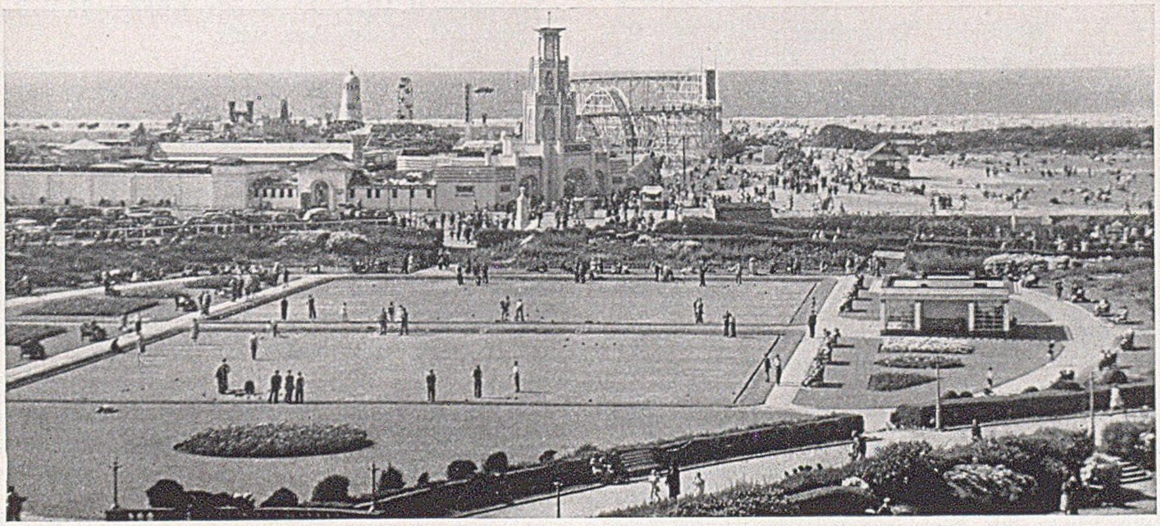
WILLIAMS, Keith Juxon, Student of Metallurgy, University of Manchester.



The Promenade, Southport.



South Marine Park and King's Gardens, Southport.



Bowling Greens and Pleasureland, Southport.

Photos: Valentine, Dundee.

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Thomas, W. J., M.I.Mech.E., M.I.E.E.; Assistant Managing Director, The British Aluminium Co., Ltd.; Director, North British Aluminium Co., Ltd.; Director, Lochaber Power Co.; Director, Imperial Magnesium Corporation, Ltd.; Director, William Mills, Ltd.; Director, Alumina Co., Ltd.; Director, Gold Coast Bauxite Co., Ltd.; Director, Aluminium Corporation, Ltd.; Director, British and Colonial Bauxite Co., Ltd.; Director, Alliance Aluminium Holdings, Ltd.; Director, AluminiumWire and Cable Co., Ltd.

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Oxford

Parker, R. T., Ph.D., B.Sc., A.R.S.M., Director of Research, Aluminium Laboratories, Ltd., Banbury, Oxon.

Scottish

Fowler, E. A., B.Sc., A.R.T.C., Chief Metallurgist and Chemist, Scotts' Shipbuilding and Engineering Co., Ltd., Greenock.

Sheffield

Hallett, M. M., M.Sc., F.I.M., Research and Development Manager, Sheepbridge Engineering, Ltd., Chesterfield.

South Wales

Spring, K. M., A.I.M., Assistant Technical Officer, Imperial Chemical Industries, Ltd., Metals Division, Landore, Swansea.

Representatives of Other Bodies :

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

The Admiralty

Grylls, Captain (E.) H. J. B., R.N.

The War Office

Joslin, Major-General S. W., C.B.E., M.A., M.I.Mech.E.

The Institution of Metallurgists

Cook, Maurice, D.Sc., Ph.D., F.I.M.
West, E. G., Ph.D., B.Sc., F.I.M.

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PERSONAL NOTES

PROFESSOR P. G. BASTIEN, has been appointed Director of Scientific Research at the Ecole Centrale des Arts et Manufactures, Paris, in succession to Professor A. Portevin.

MR. G. W. BOOTH has been elected President of the British Bronze and Brass Ingot Manufacturers' Association for the third year in succession.

MR. L. M. CLAREBROUGH has returned to the Division of Tribophysics, Commonwealth Scientific and Industrial Research Organization, University of Melbourne, after spending 15 months in England.

DR. E. C. ELLWOOD, until recently Senior Lecturer in Metallurgy at King's College, Newcastle-upon-Tyne, has joined the Tin Research Institute, Greenford, Middlesex, as Chief Metallurgist.

MR. P. FLATLEY has left Associated Lead Manufacturers, Ltd., Chester, and taken up a post as Assistant Smelting Manager with McKechnie Brothers, Ltd., Widnes.

MR. R. S. JACKSON has left David Brown Foundries Co., Penistone, and has joined the staff of the British Non-Ferrous Metals Research Association as an Investigator in the Castings Section.

MR. J. E. JENKINS has resigned his post with Enna, Ltd., Perivale, in order to take up an appointment as metallurgist with Magnesium Elektron, Ltd., Clifton Junction.

DR. O. T. MARZKE, Superintendent of the Metallurgy Division at the U.S. Naval Research Laboratory, Washington, D.C., has been appointed Consultant to the Director of Research at N.R.L. in the fields of chemistry, electricity, mechanics, and metallurgy.

DR. A. J. W. MOORE has left the Research Laboratory on the Physics and Chemistry of Surfaces at Cambridge and has returned to Australia, where he will rejoin the Division of Tribophysics of the Commonwealth Scientific and Industrial Research Organization at Melbourne University.

DR. F. R. N. NABARRO, Lecturer in Metallurgy at Birmingham University, has been appointed to the Chair of Physics at Witwatersrand University, Johannesburg.

PROFESSOR HUGH O'NEILL, President of the Institution of Metallurgists, has been admitted as a permanent member of the General Committee of the British Association for the Advancement of Science.

MR. A. R. POWELL, Research Manager of Johnson, Matthey and Co., Ltd., has been elected a Fellow of the Royal Society.

MR. S. T. QUAASS has left the National Gas Turbine Establishment and returned to the Aeronautical Research Laboratories, Melbourne.

MR. W. RACHINGER has recently been awarded the Ph.D. degree of Melbourne University.

MR. W. J. ROBERTS has joined Arrow Press, Ltd., Watford, as Editor of *Industrial Finishing*.

SIR ARTHUR SMOUT retired from the Board of Imperial Chemical Industries, Ltd., on 28 February. He was appointed to the Board in 1944 and in the same year became Metals Group Director. In July 1951 he also took over responsibility for the Nobel Division. Sir Arthur served for 48 years with the Company and its predecessors, having joined Elliotts Metal Co., Ltd., in 1905. From 1934 until 1942 he was Chairman of the Delegate Board of I.C.I. (Metals), Ltd. From 1942 to 1945 he was Director-General of Ammunition Production at the Ministry of Supply, being knighted for his services in 1946. He was President of the Institute from 1948 to 1950.

DR. P. VAJRAGUPTA has been transferred from the Government Department of Mines to the Department of Metallurgical Works, Ministry of Industry, Bangkok.

MR. H. L. WAIN has been awarded the Ph.D. degree of Birmingham University.

Deaths

The Editor regrets to announce the deaths of:

MR. KENNETH JOSEPH FORREST, Technical Officer, Imperial Chemical Industries, Ltd., Metals Division, Birmingham, at the age of 32.

Mr. Geoffrey Bastion Harris, M.A., a Bursar of the British Iron and Steel Research Association, working at the Inorganic Chemistry Laboratory, Oxford, as the result of an accident on 7 March.

Mr. Douglas Jepson, Head of the Department of Metallurgy at the Birmingham College of Technology since 1947, on 24 February, aged 51.

OBITUARY

Mr. W. A. Fowler

William Arthur Fowler, who died suddenly on 16 February 1953, at the age of 60, had spent the whole of his professional career in the light alloy industry in various capacities with The British Aluminium Co., Ltd.

After serving throughout the 1914–18 war, Mr. Fowler received his metallurgical education at Sheffield University, of which he was an Associate in Metallurgy. He was first employed by The British Aluminium Company in their Metallurgical Laboratories at Milton in 1922. He became a Superintendent at the Warrington factory in 1925 and Assistant Manager in 1930. In 1934 he returned to Milton to take over the management of that Works. From 1943 to 1945 he managed not only the Milton, but also the Warrington factory. In 1946 he was transferred to Scotland to manage the Company's rolling mill at Falkirk. He held this position until January 1949, when he moved to the Head Office of the Company to become Production Manager for Manufactured Materials. In this capacity he was concerned with the production side of the Company's manufacturing interests, not only at their rolling mills and extrusion factory at Milton, Warrington, and Falkirk, and smelting works at Latchford, but elsewhere in certain subsidiary and associated companies. He held this position until his death.

Mr. Fowler was extremely well liked and respected by all who worked or had associations with him, both in his own Company and outside it. He possessed a most active and original mind, and was keenly interested in every aspect of production metallurgy and management.

Mr. Fowler was joint author, with the writer, of a paper dealing with problems affecting production economy in the rolling of aluminium, which was read before the Institute at its Autumn Meeting in Paris in 1949, and for which the authors were subsequently awarded the W. H. A. Robertson medal.

Mr. Fowler was a Fellow of the Institution of Metallurgists and had been a Member of the Institute of Metals since 1920.

W. J. Thomas.

LETTER TO THE EDITOR

Dislocation Theories

In your abstract of Mme. A. R. Weill's paper "Mechanism of Plastic Deformation. . . . According to Recent Papers of W. A. Wood",[1] it is remarked that my interpretation of my results makes no appeal to the theory of dislocations and polygonization. May I qualify that?

The main results are: (i) plastic deformation in general reduces the metallic grain to a sub-structure; (ii) the size distribution of the elements depends systematically on the conditions of deformation, e.g. temperature and rate of strain; and (iii) the size distribution largely determines the strain-hardening.

I interpret the first two in terms of dislocations. It has been evident from the beginning that any boundary is a surface of dislocations, vacant sites, and interstitial ions and therefore that the sub-boundaries form by a segregating of such imperfections. Differences have arisen only in picturing the segregation. I have taken the view that it was the natural and virtually immediate consequence of trying to bend or otherwise deform a lattice already studded with imperfections. The alternative view postulates polygonization. This, as first proposed, was a two-stage process: first a considerable bending of the lattice; then a flow of imperfections to the boundaries-to-be, but only after a heat-treatment. My difficulty was with this need for temperature-activation. It seemed contrary not only to our own evidence but also to the beautiful X-ray work with micro-beams at Cambridge and elsewhere, which demonstrates that plastic deformation produces sub-structures at room temperature and even at liquid-air temperature.[2] If it turns out that temperature-activation is unnecessary, there will be little to choose between the two views.

The third result presented more difficulty. The obvious interpretation was that the network of dislocations and other imperfections forming the sub-boundaries also provided a force-field sufficient to impede the flow of other dislocations and so cause the strain-hardening. A hardening effect by sub-boundaries was reasonable on general grounds. It was a logical extension of the well-known observation that a metal is harder, the smaller its grain-size. It also received some theoretical support from Bragg's theory of yield. Moreover, in the later work[3] on "stress-recovery", we could directly follow and correlate changes in hardness with changes in sub-grain size. But it was difficult to reconcile the observations with the generally accepted dislocation theory of hardening due to G. I. Taylor. This postulated that the necessary field was provided by an array of positive and negative dislocations alternating in space like the sodium and chlorine atoms in a rock-salt structure. It seemed that a purely hypothetical array for which there has never been any direct experimental evidence was taking the place of a natural array for which evidence has continually accumulated.

However, it now appears from the theory of strain-hardening recently put forward by Mott[4] that the Taylor solution is not unique and may be effectively replaced by a sub-boundary array after all. It is too early to say whether the network pictured by Mott is also that observed experimentally, but it would seem that the theoretical and experimental approaches are converging for the first time. I should like to make it clear, therefore, that the interpretations in question are not at variance with the general theory of dislocations, but only with certain forms which that theory has taken and which I suggest are open to legitimate criticism.

W. A. Wood.

The Baillieu Laboratory,
University of Melbourne.

REFERENCES

1. A. R. Weill, *Mem. Artillerie Franç.*, 1952, **26**, (1), 41; see *Met. Abs.*, 1952–53, **20**, 232.
2. P. A. Beck, private communication.
3. W. A. Wood and J. W. Suiter, *J. Inst. Metals*, 1951–52, **80**, 501.
4. N. F. Mott, *Phil. Mag.*, 1952, [vii], **43**, 1151.

ROLLS AND THEIR MAINTENANCE IN THE NON-FERROUS METALS INDUSTRY

B4

An informal discussion on " Rolls and Their Maintenance in the Non-Ferrous Metals Industry ", was held at Birmingham University on the 8 January 1953, the Chair being taken by Mr. W. J. Thomas, Chairman of the Metallurgical Engineering Committee.

In accordance with the Chairman's request in his opening remarks, the discussion comprised, firstly, an account by roll users of their experience and their views on desirable developments in rolls ; secondly, comments by representatives of roll makers on what had been said by roll users ; and, finally, a general exchange of views on various aspects of roll maintenance.

The following is a summary of the discussion under these headings.

Roll Usage : Experiences &c., of Roll Users

Rolling of Light Alloys

(a) *Hot Rolling*

With rolling speeds of about 450 ft./min. and metal temperatures ranging from 250° to 530° C., two-high hot mills having roll barrels up to 35 in. in dia. by 96 in. in length, and four-high mills of sizes up to 24 and 56 in. by 88 in., are employed.

Various types of rolls, namely chilled iron, cast steel, composite forged steel, and solid forged steel, have been used in two-high hot mills. Chilled-iron rolls, which have been used quite successfully in small hot mills, cannot be employed in the larger modern mills because the heavier rolling load calls for a journal/barrel diameter ratio of 4 : 5, and conventional mill designs will not accommodate rolls of these proportions. Composite forged-steel rolls have been used with some success in the smaller mills, but have proved unsuitable for the larger mills. Repeated failures of composite roll shells, either by total fracture or by their lateral displacement on the arbors or even fracture of the arbors themselves, have resulted in the rejection of this type of roll for hot-rolling duty. In the case of a mill having composite rolls 34 in. in dia. by 72 in. in length, it has been found that the average mill throughput before scrapping was 12,000 tons per roll, each roll having been reshelled once during its life.

The present trend for hot-rolling duty is to use solid-forged rolls of 65°–70° hardness, as measured on the Shore Type C Scleroscope, the hardness being critical, since experience has shown that rolls of 78°–80° hardness are subject to severe spalling, necessitating, on occasion, barrel reductions of $\frac{1}{2}$–$\frac{3}{4}$ in. for its removal. An analysis of the tonnage rolled with 78°–80° hardness rolls before scrapping reveals a figure of 22,000 tons per roll. Rolls of 50°–55° hardness, although not subject to spalling, suffer from persistent superficial cracking. Solid-forged rolls of 65°–70° now in use have proved to be fully resistant to indentation, and although the tonnage rolled already exceeds 41,000, no cracking, crazing, or spalling has been experienced. In general, the reason for changing these rolls has been the development of pitting of the barrel surface that usually appears at service tonnages of between 5000 and 8000. The cause of this pitting is under investigation.

One of the principal problems requiring the attention of roll makers and rolling-mill manufacturers is the question of roll journal wear. It has been observed that whereas formerly the ultimate potential roll life was determined by the rate of barrel wastage through grinding, the extent and frequency of such grinding have been so reduced that journal wear now threatens to dictate ultimate potential roll life. Attention to the degree of hardness and possibly also local hardening to resist wear is therefore necessary. In addition, attention should be paid by mill makers to the possibility of equipping two-high hot mills with totally enclosed bearings, which have the virtue that a separate cooling cum lubricating solution can be used for the journal bearings.

In four-high hot mills solid-forged steel work-rolls are used in conjunction with cast- or forged-steel back-up rolls. The principal trouble experienced is the total failure of both cast and solid-forged steel back-up rolls. These failures have generally taken the form of total fracture along a plane roughly at right angles to the axis of the roll, and a development in manufacturing technique is required to eliminate the causes of failure of these costly rolls, which is usually attributed to internal stresses.

(b) *Cold Rolling*

For strip rolling, solid-forged steel work-rolls are almost universally used in both two-high and four-high mills, and considerable experience has been gained with both American and British rolls for this duty. Until a few years ago difficulty was experienced in this country in obtaining, in the sizes needed, rolls which were sufficiently hard to prevent rapid deterioration of the barrel surface, but which, at the same time, were resistant to indentation and spalling. American rolls possessed these desirable features, particularly the former. The American rolls were described as " deep hardened ", having a usable hardened case $\frac{3}{8}$ in. thick, and experience has shown that for a diameter reduction of $\frac{1}{4}$ in. on these rolls, no deterioration in the surface hardness occurs. Up to one or two years ago British rolls of the same size developed spalling and severe indentations, and after a barrel diameter reduction of $\frac{1}{4}$ in. the surface hardness dropped by 10°–18°. As the indentations might occur at any time while the rolls were in service in the mill, they involved the roll user in the problem of roll changing on a day-to-day or even hour-to-hour basis. Similarly, rolls of inadequate surface hardness involved the user in too-frequent roll changing owing to the rapid deterioration of the barrel surface.

Records show that American rolls were, up to one or two years ago, giving about two and a half times the length of service between regrinds as were given by those of British manufacture, but rolls recently made in this country are showing considerable promise in extended periods of service between regrinds and resistance to indentation without an undue increase in the incidence of spalling.

This recent experience indicates that the development of manufacturing technique in producing work-rolls for the cold strip rolling of light alloys should continue to be in the direction of increasing the depth of hardened case, with attention being paid also to the intrinsic hardness of the barrels.

Back-up rolls for four-high cold strip mills may be of forged steel, either solid or of composite construction, or cast steel. Composite forged-steel rolls have given very variable performances, whereas both solid-forged and cast-steel rolls have given consistently good results. In general, solid-forged and cast-steel rolls are now favoured for this duty. It has been noted that cast-steel rolls wear roughly twice as quickly as forged-steel rolls when engaged on identical rolling duties. Actual figures taken over a period of 9 years show that the tonnage rolled per inch of reduction of the barrel diameter in the case of cast-steel back-up rolls was 350,000, and that for forged-steel rolls 650,000. The hardness of the cast rolls was 50°–55° and that of the forged rolls 70°–75°.

Cast-steel rolls, which are at present limited in hardness to about 55°, require to be developed to give the higher hardness necessary to resist wear. Forged rolls, when supplied in the hardness range 70°–75°, have the virtue that they mark or suffer indentation rather than the work-rolls when pieces of the rolled stock pass accidentally into the " bite " of the work and back-up rolls. Higher hardness back-up rolls are used, however, and one roll maker subsequently pointed out that in the U.S.A. the tendency is to increase the hardness of back-up rolls in order to increase their resistance to deformation during rolling.

In the case of the flat-sheet rolling of light alloys, the problems in terms of surface finish are similar to those in strip rolling. The types of roll employed are solid-forged steel and both the clear and indefinite chilled irons. Past difficulties in controlling the defects of mottled surfaces and colour variation are now being overcome in the indefinite chill roll of 90°–95° Shore hardness. In general, the forged-steel roll gives a 40% better mill life than the chilled-iron roll, but the choice of the type of roll to be used is a domestic issue, depending on the type of mill in use and the product being rolled.

Rolling of Copper, Brass, Zinc, Cupro-Nickel, &c.

Hot rolling is carried out at temperatures ranging from 250° to 950° C. In general, indefinite chill rolls are employed, but some forged-steel rolls are used, particularly where a variety of alloys is being processed. The initial cost of forged-steel rolls is much higher than that of chilled-iron rolls, but, under comparable operating conditions, the roll life is proportionately longer. It was suggested that a composite roll comprising a steel arbor with a cast-iron shell might be the most suitable.

In cold breaking-down mills of the four-high type, forged-steel work-rolls of 80°–90° hardness are used. Although rolls of this hardness tend to mark easily, any increase in hardness gives rise to spalling owing to the heavily loaded condition of these mills. The tonnage rolled with present-day forged-steel rolls shows considerable improvement over that obtained about 16 years ago. The average output per roll obtained in a 15-in. and 36 × 32-in. four-high mill in 1936 was 14,000 tons, whereas in this same mill recent output figures average 51,400 tons per roll.

Cold-finishing mills employing chilled-iron rolls generally demand a hardness of 95° minimum, but even at this hardness marking is prevalent, and maintenance of the required surface finish, particularly on sheets, is difficult. A number of forged-steel rolls stated to be 28 in. in dia. by 16 in. long, obtained before the War from the Continent and during the War from America, gave very good service, but there is reluctance on the part of some British manufacturers to supply forged-steel rolls of such a size for this duty.

Rolling of Nickel Alloys

Some of the most severe working conditions encountered by rolls in the non-ferrous metals industry arise in both the hot and cold rolling of nickel alloyed with such metals as cobalt, iron, chromium, molybdenum, and copper.

Hot-rolling mills operate at temperatures up to 1300° C., and difficulty is experienced in obtaining rolls that will withstand the heavy rolling pressures at these high temperatures. In the case of a three-high mill with 16-in.-dia. rolls, handling billets 7 in. wide × 2 in. thick, chilled-iron rolls failed owing to fracture, severe crazing, or rapid wear of the barrels. The more expensive solid-forged steel rolls of 63°–65° hardness give reasonable service and are now used for hot-rolling duty, but trials are being carried out with so-called " steel-base " rolls, the results of which are not yet available.

Cold breaking-down of nickel alloy strip from about 0·30 in. gauge to about 0·048 in. is carried out on either two-high or four-high mills. In a two-high mill, employing rolls 14 in. in dia., cast-steel rolls have entirely superseded chilled iron for this arduous duty. A four-high mill engaged on similar work is equipped with 7 in.-dia. forged-steel work-rolls of 100° hardness and 12-in.-dia. steel back-up rolls. The hardness of the back-up rolls has been limited to 75°, as it has been found that rolls of greater hardness tend to skid on the work-rolls.

In the cold-finish rolling of nickel alloys, perfection of surface finish is not the prime consideration, but accuracy of gauge is important. A high degree of work-hardening is desirable when cold rolling to thicknesses less than 0·001 in., and this subjects the rolls to intensive loading. Rolls of extreme hardness, capable of withstanding high pressures, are essential for cold-finishing duty in order to prevent grooving of the barrels by the passage of the strip.

Mills of various types are used for cold-finishing work including four-high mills with 5-in.-dia. work-rolls for reductions in strip thickness down to about 0·005 in., four-high mills with 2-in.-dia. work-rolls for reductions down to 0·0015 in., and twelve-roll mills with $\frac{3}{8}$-in.-dia. work-rolls for rolling to the finest gauges. The hardness of forged-steel rolls on this duty is limited to 105°, since higher values are accompanied by spalling and cracking.

The application of solid carbide work-rolls of $\frac{3}{8}$ in. dia. has proved successful on the twelve-roll mill, with the advantage of greater reductions. Both solid carbide and composite carbide-shell rolls having a steel arbor are under consideration for 2-in.-dia. work-rolls.

Roll Manufacture and Usage: Comments by Roll Makers

Spalling of rolls is common to most branches of the non-ferrous metals industry, particularly where the duty calls for a high degree of roll surface hardness. It was suggested that the origin of a spall lay in small flaws or inclusions giving rise to fatigue cracks which develop under the very considerable stresses set up during rolling, but, on the other hand, it was stated that such flaws might not be inherent in the roll but might be caused by severe local overloading. (On this point, subsequently, a roll-user suggested that breakdown of lubrication might also give rise to spalling.)

Spall resistance in chilled-iron rolls is greater in the indefinite chill type than in the clear chill rolls because the microscopic graphite seeds present in the outer hardened case of the roll give a cushioning effect by taking up some of the local overloads that occur during rolling. The hardness limit for indefinite chill rolls is 95°.

The structure of a chilled-iron roll comprises a white-iron outer case, which is carbidic in essence and, hence, brittle, but has a very high compressive strength. Immediately behind the carbidic zone is the transition zone of mottle iron followed by the core, which is of grey iron. Hard alloy chilled-iron rolls are simple alloys containing nickel, chromium and molybdenum or nickel and molybdenum or manganese and nickel.

Where chilled iron or steel rolls exhibited " craze-cracking ", it was contended that the remedy lay with the roll users, who should either modify their rolling technique, or alternatively remove the cracking in its early stage by frequent grinding.

Fully hardened alloy forged-steel rolls are manufactured in this country with barrels up to 32 in. in dia. and 96 in. in length, the hardened portion of a 95°–100° Shore hardness roll having a tensile strength of the order of 200 tons/in.^{2} and a compressive strength probably in the range 300–350 tons/in.^{2}.

Premature failure of forged-steel rolls may be associated with the quick stress and temperature reversals that take place in reversing mills, since there is some evidence that premature failures are attributable to a process of fatigue of the roll metal taking place progressively from the outside towards the inside of the roll section.

The problems of initial roll surface hardness and its penetration to a predetermined depth in the body of the roll are receiving active attention from roll makers in this country. Rolls with a surface hardness up to 105° Shore, and also rolls of lower surface hardness but having deeper penetration, are obtainable from manufacturers in this country.

A plea for standardization of drive-end design was registered by roll makers, with a departure from the lobular type of roll wobbler, which calls for extensive machining on forged-steel rolls.

Roll Servicing, Maintenance &c.: General Discussion

Hardness Measurement

The instrument employed for the hardness measurement of finish ground rolls is the Shore Scleroscope, the principal merits of which are portability and measurement without destruction of the surface. It provides neither an absolute measure of hardness nor a measure of resistance to indentation, but rather a comparison of material resilience. Furthermore, facilities appear to be lacking for the proper recalibration of these instruments against recognized standards, wide discrepancies having been noted between different instruments.

Roll makers employ the diamond hardness test during the manufacture of a roll and, by making standard sample blocks of various hardness values, are able to check their Scleroscopes against the diamond hardness figures; roll users, however, naturally avoid the use of the diamond test or any other test instrument which is destructive of the surface being measured.

It was concluded that the Scleroscope is a satisfactory comparator subject to the institution of periodic calibration against recognized standards.

In the absence of information about the physical characteristics and specification of the rolls supplied to him, the roll user is generally obliged, when comparing roll performances, to identify the rolls by their Scleroscope hardness. Roll makers should, therefore, give more information about the mechanical properties and metallurgical composition of their rolls so that users can determine from experience the most suitable roll specification for a particular duty.

Roll Journals and Oil Seals, &c.

Adequate oil sealing in high-speed rolling mills engaged on light alloys constitutes a major problem, with particular reference to the preferential roll journal wear over those areas in contact with the oil seals. Failure of an oil seal often occasions a roll change because of the loss of bearing oil into the rolling oil system, with its attendant effect on roll-bite conditions, staining of the product, and the resulting rejection of the rolling lubricant for contamination. A possible solution lies in the provision by roll makers of locally hardened journal areas to resist seal wear. Wear of the journals might be adjusted by some form of rebuilding, such as metal spraying, since their reconditioning by this means has met with some success where journal loadings are not high.

Roll Grinding

Grinding of sheet and strip mill work-rolls is necessary when the surface condition has deteriorated to such an extent that the surface finish of the product is no longer acceptable or that the life of the roll is endangered. Deterioration may take the form of indentation, cracking, spalling, grooving, or other roll-surface markings which affect the rolled product. Back-up rolls, on the other hand, are often left in service, particularly in light-alloy mills, for a definite planned period, frequently as long as six months, without any apparent detriment to the roll. In order to ensure that any fatigued or work-hardened metal is removed from cast-steel back-up rolls after such a period of service, it was urged by a roll maker that the roll diameter should be reduced by about 0·040 in.

The attainment of the roll surface finishes and cambers currently required calls for precision grinding on accurate machines, using grinding wheels of good and consistent quality. The point was made that in the case of British machines there was scope for improvement in respect of more robust construction and precision of design, so that a uniform high roll finish could be consistently achieved with the machine without making a heavy demand on the skill of the operator.

Heavy grinding work should be confined to machines allocated for that purpose, leaving accurate finish grinding to machines allotted to precision duty. Close attention should be paid to the concentration and temperature of the grinding-wheel coolant, and above all it is imperative that grinding wheels be accurately balanced.

Appreciable variation is experienced by some users in the quality of grinding wheels to the same specification, particularly with fine wheels such as 500 grit. The surface finish obtained with a particular wheel is often difficult to reproduce with a second wheel to the same specification.

It was suggested by one roll user that it would be advantageous if grinding machines were so designed as to equip them with means for selecting a choice of camber shapes, since, at present, the only shape obtainable is that resulting from the particular camber mechanism incorporated by the makers in their machine. (This remark may occasion some surprise, and it is thought that the speaker intended to convey that British machines are not equipped with a device for adjusting camber shape, although, of course, changes can be achieved by fitting modified camber generating mechanisms.)

As an illustration of high-precision grinding of rolls for use with nickel alloys, grinding tolerances of less than 0·00002 in. are demanded for rolls engaged on strip gauges down to 0·0002 in. These requirements were said to be met by a machine of Swiss manufacture, but no indication was given of the measuring means appropriate to such fine tolerances.

Protection of Roll Surface

After grinding the rolls, it is essential that the ground surfaces be quickly protected from atmospheric corrosion. Various materials are used for this purpose, including corrosion-resistant greases, non-setting paints, and hard varnishes. Vapour inhibitors are also used with some success, but the choice of protective material is a domestic issue, experience dictating the most suitable medium. After application of the protective material, the rolls should be stored in a warm, dry atmosphere clear of the shop floor and supported on timber bearers.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

Sheffield Local Section

At a joint meeting of the Sheffield Local Section with the Sheffield Society of Engineers and Metallurgists and the Sheffield Metallurgical Association, held at the University on 16 February, Mr. W. A. BAKER, Research Manager to the British Non-Ferrous Metals Research Association, gave a lecture on:

The Constitution and Properties of Some Titanium-Base Alloys

After a brief reference to the composition and other characteristics of iodide titanium and of the sponge (Kroll) material, the lecturer referred to the effects of various alloying elements on the stability of the α and β phases, illustrating his remarks by reference to three typical phase diagrams. The mechanical properties of the metal and of a number of its alloys were then discussed and were related to certain features of the constitutional diagrams. The lecture concluded with a few general comments on the engineering properties of titanium and its alloys and on the behaviour of the materials in fabrication.

OTHER NEWS

Heat-Treatment Practice

Preliminary notice is given by the British Iron and Steel Research Association of a conference on Heat-Treatment Practice to be held on June 15, 16, and 17 next, at Ashorne Hill, Leamington Spa. It is intended that there shall be papers and discussion on the heat-treatment of engineering steels and large forgings and sections, surface treatments, isothermal heat-treatment, and temperature measurement and control.

The conference, which is being organized by the Metallurgy (General) Division of B.I.S.R.A., is expected to provide a meeting ground for engineers and steelmakers affected by these problems. Any persons wishing to attend should apply to Mr. A. M. Sage, B.I.S.R.A., 11 Park Lane, London, W.1.

DIARY

7 May. Leeds Metallurgical Society. Annual General Meeting, followed by Junior Members' Papers. (Chemistry Department, The University, Leeds 2, at 7.15 p.m.)

13 May. Society of Chemical Industry, Corrosion Group. Annual General Meeting, and Chairman's Address by Dr. W. H. J. Vernon. (Chemical Society, Burlington House, Piccadilly, London, W.1, at 6.30 p.m.)

14 May. Institute of Metal Finishing, North-West Branch. "The Metal Finishing of Non-Metallic Materials", by H. Ashton. (Engineers' Club, Albert Square, Manchester, at 7.30 p.m.)

18 May. Institute of Metal Finishing, London Branch. "Instruments for the Determination of Coating Thickness: A Review of Commercial Instruments", by E. S. Spencer-Timms. (Northampton Polytechnic, St. John Street, London, E.C.1, at 6.0 p.m.)

APPOINTMENTS VACANT

GALVANIZING (HOT-DIP). Manager required for old-established galvanizing works in Midlands. General work, malleable castings, tubes, sheet metal work, grey iron castings, wire work, &c. Present output running between 50 and 100 tons per week. The manager would be required to plan production, organize all labour, and supervise the care of all equipment. Man with previous experience preferred; alternatively man with technical qualifications would be considered. Salary, according to qualifications, would be made up of basic monthly salary plus a bonus on profits of the department. Applications will be treated in strictest confidence. Please write in first instance to A. A. Black, Limited, (Dept. Y), 113 Bristol Road, Birmingham, 5.

METALLOGRAPHER for laboratory near Northampton engaged in development of components and materials for the radio, electronic, and radar industries. Work would include investigations of a variety of special metallic and ceramic materials. Few years' experience essential. Salary according to age, qualifications, and experience. Box No. 350, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGIST, with University qualifications, required for investigational work on problems associated with the welding of steels. Some experience in research or industry essential and a knowledge of arc welding an advantage. Excellent salary and prospects for suitable applicant. Apply in writing to Box JIM 205, LPE, 110 St. Martin's Lane, London, W.C.2.

MINISTRY OF SUPPLY, Royal Aircraft Establishment, Farnborough, Hants, requires metallurgist to work in team engaged on titanium alloy development. Qualifications 1st or 2nd class honours degree or recognized equivalent in metallurgy. Preference for and skill in experimental work, especially metallography, required, experience of metallurgical research desirable. Salary within range, Scientific Officer, £417–£675. Women somewhat less. Post unestablished. F.S.S.U. benefits may be available. Application forms from M.L. & N.S. Technical and Scientific Register (K), Almack House, 26 King Street, London, S.W.1, quoting F108/53A. Closing date 15 May, 1953.

THE MOND NICKEL COMPANY, LIMITED, requires a manager for its research department situated at the refinery in South Wales. The position is of high seniority and importance, and a commensurate salary will be paid. A house is available for residence, and the Company operates a retirement pension scheme. Applicants should preferably be not more than 40 years of age, have had good research experience, and be competent to inspire and lead a team investigating physico-chemical problems. Applications, which will be treated as confidential, should be addressed to The Resident Director, The Mond Nickel Company, Limited, Clydach, Swansea.

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THERMODYNAMICS OF ALLOYS

By JOHN LUMSDEN, B.Sc.

(Research Department, Imperial Smelting Corporation, Ltd.)

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

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

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

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

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

Institute Dinner at the Park Lane Hotel, London, 24 March, 1953.

[*Photo: Lawson and Co.*

The President (Professor F. C. Thompson) was in the Chair, and the Principal Guests were Air Commodore F. R. Banks, C.B., O.B.E. (Principal Director of Engine Research and Development, Ministry of Supply) and the Deputy Mayor of the City of Westminster.

MECHANICAL ANISOTROPY IN SOME DUCTILE METALS*

By Professor W. A. BACKOFEN,† S.B., Sc.D., and
B. B. HUNDY,‡ B.Sc., Ph.D., Member

SYNOPSIS

Fracturing test-specimens in tension after prestraining in torsion has shown that a fibrous crack-like structure, causing a considerable degree of mechanical anisotropy, exists in 70 : 30 brass, nickel, Monel metal, and Armco iron. The same programme of testing has also revealed the presence of such a structure in high-purity aluminium, but, for reasons that are not clear, the tensile behaviour of torsionally prestrained commercially pure (2S) aluminium gives no indication of its presence.

I.—INTRODUCTION

Many investigations [1-10] have shown that wrought metals are highly anisotropic and that the anisotropy is not eliminated by annealing. The fracture properties, in particular, have been found to be affected by the direction of testing. The majority of these investigations were carried out on steels or high-strength non-ferrous alloys in which the mechanical anisotropy is usually attributed to alignment in the direction of metal flow of regions of segregation, of cavities, or of various phases in the microstructure. Anisotropy may also be the result of preferred orientation of the grains, and there are examples of directionality of mechanical properties being caused by a combination of both "fibring" and preferred orientation.

Recent work by the present authors [11] has shown that mechanical anisotropy is developed in pure copper by severe hot working or by cold working and annealing. The presence of anisotropy in the metal was demonstrated by fracturing in tension, specimens previously twisted to various degrees of prestrain.§

The tensile fracture stress, ductility, and type of fracture were not affected by small amounts of twisting, but when the specimens were twisted more severely, fracture occurred by abrupt separation across a helical surface rather than in the usual ductile manner. The change in the appearance of the fracture was accompanied by a fall in the fracture stress and ductility. If a specimen was twisted severely and then untwisted, subsequent tension testing showed that prestraining had not significantly altered either the fracture stress or the ductility, and the ductile mode of fracture reappeared. These results seemed to indicate that the material was highly anisotropic before twisting; this conclusion was confirmed by tests on specimens cut in different directions from the original material.

The copper used in these experiments was of high purity, and therefore the results could not be explained by alignment of secondary phases or impurities in the direction of metal flow. An examination of the torsion texture of copper [13] showed that preferred orientation of the grains could not account for the anisotropy, and it was therefore suggested that it resulted from a highly oriented structure of flaws.

These flaws behave very much like cracks, and in the present paper they will be described as micro-cracks, even though they are apparently too small to be visible under normal microscopical examination. It is thought that some micro-cracks originate during solidification of the ingot, and some perhaps are formed during plastic working when they all, regardless of origin, become aligned in the direction of metal flow. They weaken the metal in the transverse direction, but do not affect the longitudinal mechanical properties. Torsional prestraining of a specimen containing micro-cracks aligned in the longitudinal direction causes them to follow a helical path, and in subsequent tension fracture takes place across this path.

Apart from the experiments on copper described above, there is very little published information on mechanical anisotropy in pure ductile metals or in single-phase alloys. The work reported in the present paper arose from a desire to determine whether plastic working of such metals could give rise to mechanical anisotropy and whether such an anisotropy, if found, could be explained by means of the micro-crack hypothesis outlined above.

As in the previous research, the mechanical anisotropy was studied by fracturing in tension, specimens that had been previously twisted to various strains. The experiments show that mechanical anisotropy exists in most of the metals considered and that the anisotropy can be explained by an oriented micro-crack structure.

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* Manuscript received 3 November 1952.

† Massachusetts Institute of Technology, Cambridge, Mass., U.S.A.

‡ British Iron and Steel Research Association, Sheffield.

§ This method was suggested by the work of Swift [12] on the effect of torsional prestrain on the tensile properties of mild steel.

II.—EXPERIMENTAL PROCEDURES

1. Materials

Table I summarizes many of the experimental details, and Fig. 1 gives the dimensions of the torsion specimens and of the tension specimens prepared from them after twisting. All materials were obtained as wrought bars in which the principal direction of metal flow during processing coincided with the axis of the bar. The torsion specimens were machined from these materials in the cold-worked, as-received condition. They were then annealed at the temperatures shown in the table. Both torsion and tension specimens were polished down to 0000 emery paper after machining and annealing, and before any mechanical testing was carried out.

Table I.—*Summary of Experimental Details.*

Material	Processing History	Annealing Treatment of Torsion Specimens after Machining	Testing Procedure
70 : 30 Brass	⅝-in.-dia. cold-drawn rod prepared from an ingot of high-purity components.	30 min. at 950° F. (510° C.)	Torsion—Tension.
Commercially Pure ("A") Nickel	¾-in.-dia. cold-finished rod obtained from a commercial source.	30 min. at 1350° F. (732° C.)	Torsion—Tension.
Monel Metal		30 min. at 1400° F. (760° C.)	Series A : Torsion—Tension. Series B : Torsion—Annealed 30 min. at 1400° F.—Tension.
Armco Iron		20 min. at 1500° F. (816° C.)	Series A : Torsion—Tension. Series B : Torsion—Annealed 20 min. at 1500° F.—Tension.
Commercially Pure (2S) Aluminium : Series A : Series B :	¾-in.-dia. cold finished rod obtained from a commercial source. 1-in.-dia. rod prepared by forging; 87% final cold reduction in cross-sectional area.	60 min. at 600° F. (316° C.)	Torsion—Tension.
High-Purity (99·99%) Aluminium	¾-in.-dia. rod prepared by forging; 75% final cold reduction in cross-sectional area.	60 min. at 650° F. (343° C.)	

Table I shows that two series of Monel metal, Armco iron, and 2S aluminium specimens were prepared. The B series of Monel metal and of Armco iron differed from the A series in the anneal given after twisting and before tensile testing. Both the A and B series of 2S aluminium were tested in tension directly after prestraining, but they differed in processing history. The material of the series-A specimens was obtained from a commercial source, whereas that of the series-B specimens was specially prepared with a final cold reduction in area of 87% by forging.

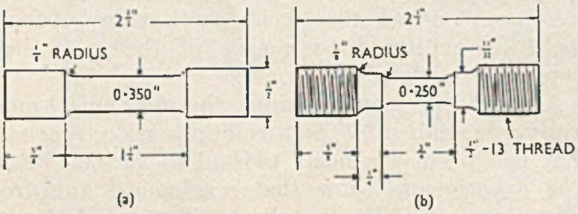

Fig. 1.—Dimensions of the (*a*) Torsion and (*b*) Tension Test-Specimen.

2. Prestraining and Tensile Testing

Both twisting and untwisting were done at a rate of 120°/min. The strain gradient in a twisted round bar is linear; shear strain, γ, increases from zero at the centre to a maximum value at the surface which is given by $\gamma = r\theta/l$, where r is the radius of the bar, θ is the total amount of twist in radians, and l is the length over which twisting takes place. This makes it a simple matter to calculate the shear strain in the surface of the smaller-diameter tension specimen.

An hydraulic machine was used for tensile testing. An average true stress/true strain curve was plotted for each test, and from these curves the tensile fracture stress and the strain to fracture (expressed as a percentage reduction in area) were obtained. Generally, diameter measurements at intervals throughout the test were adequate for calculating strain. But when the specimens became elliptical in cross-section, as did those of unstrained and only slightly prestrained aluminium, the instantaneous area of cross-section required for the strain calculation was determined from measurements of both major and minor axes.

III.—EXPERIMENTAL RESULTS

The relationship between tensile fracturing characteristics and the amount of prestraining for the brass, nickel, Monel metal, and iron specimens, tested directly after prestraining, is presented in Fig. 2. These curves show that the fracture stress and the ductility were not much affected until the shear strain introduced by unidirectional twisting exceeded a value of about 1, after which both, with the exception of the Armco iron fracture stress, underwent an abrupt decrease. The strain at which this marked change in fracturing characteristics begins will be referred to as the critical shear strain. With increasing prestrain, the fracture stress reached a minimum and then rose, whereas the ductility continued to fall.

The trend in the fracture stress data was very pronounced for all metals except Armco iron. But it was only in this respect that iron differed; the other effects of prestraining on ductility and mode of fracture were very much the same for all these metals.

After prestraining in excess of the critical amount, the usual ductile type of tensile fracture was replaced by a helical separation such as had previously been observed in copper. Fig. 6 (Plate LXVII) illustrates the fracture transition by means of typical fractures occurring in the series of nickel specimens. It is apparent from a study of these photographs that the angle α, included between the helical path of fracture in the specimen surface and a reference mark per-

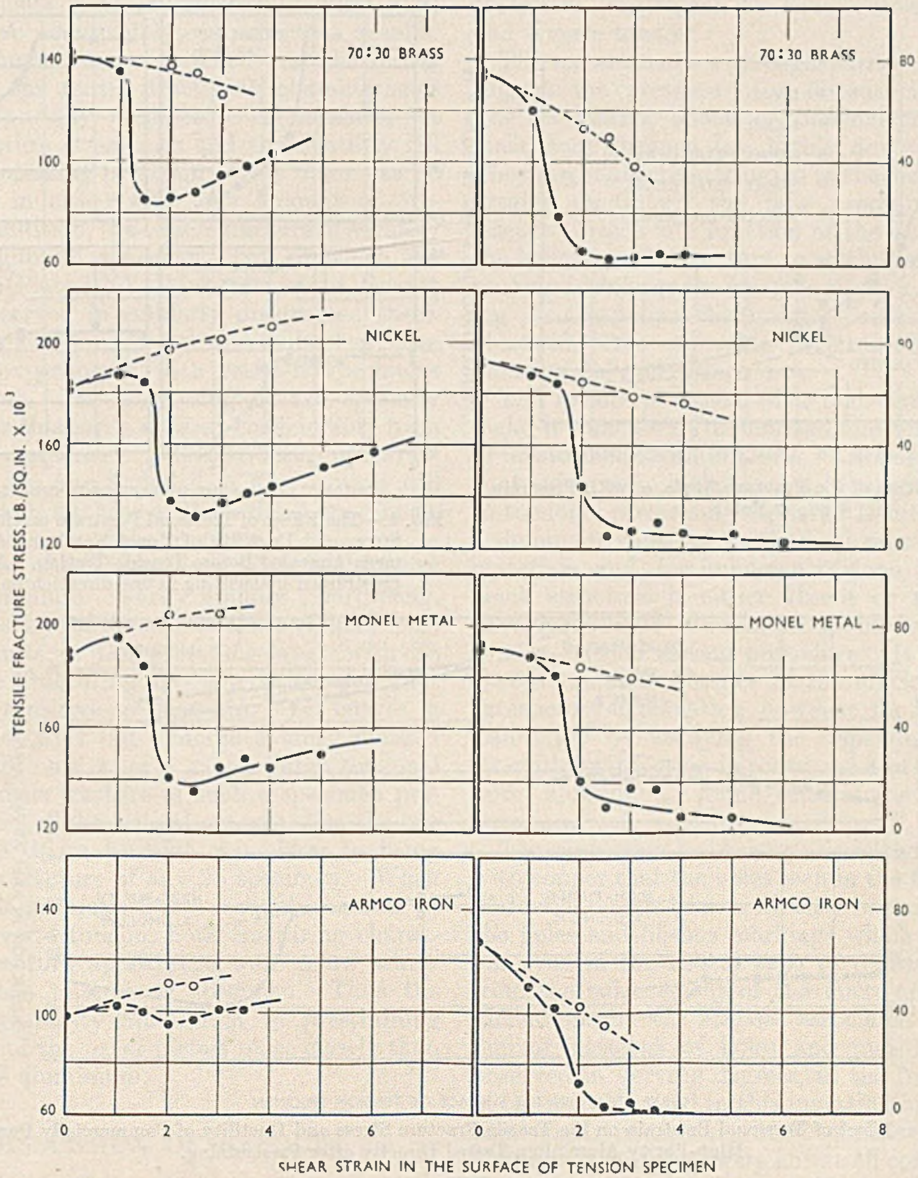

Fig. 2.—The Effect of Torsional Prestrain on the Tensile Fracture Stress and Ductility of Specimens Tested Directly after Prestraining.

Key.

● Specimens twisted to the indicated strain and then tested in tension.

○ Specimens twisted to the indicated strain, then completely untwisted, and tested in tension.

pendicular to the specimen axis, became smaller with increasing prestrain. The measurements plotted in Fig. 3 were made from photographs similar to these.

On the other hand, when prestraining consisted of equal amounts of twisting and untwisting, the tensile fracturing characteristics and fracture appearance were much more like those of specimens not prestrained at all. The values of fracture stress and ductility for specimens prestrained in this way are plotted in Fig. 2 at the maximum shear strain which preceded untwisting. Photographs of the fractures in such specimens are also included in Fig. 6 (Plate LXVII).

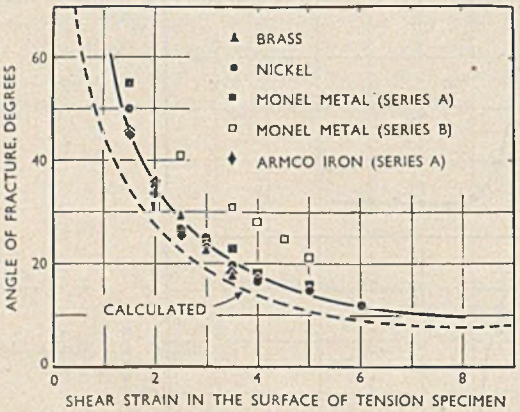

Fig. 3.—The Variation of the Fracture Angle, α, with Prestrain for Various Metals.

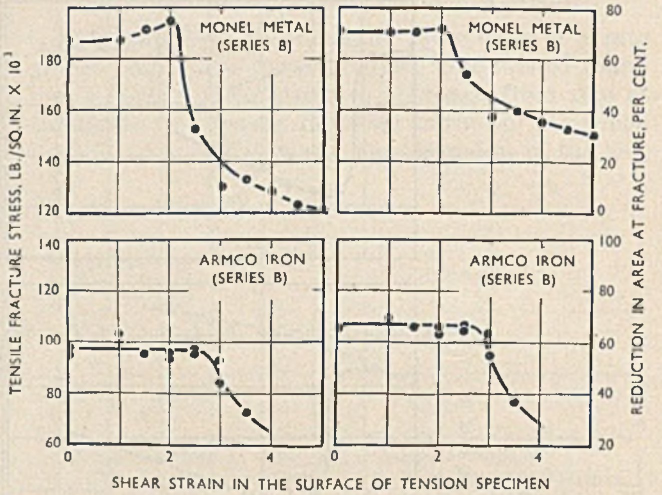

Fig. 4.—The Effect of Torsional Prestrain on the Tensile Fracture Stress and Ductility of Monel Metal and Armco Iron Specimens Annealed Before Tension Testing. All specimens were prestrained by twisting in one direction only.

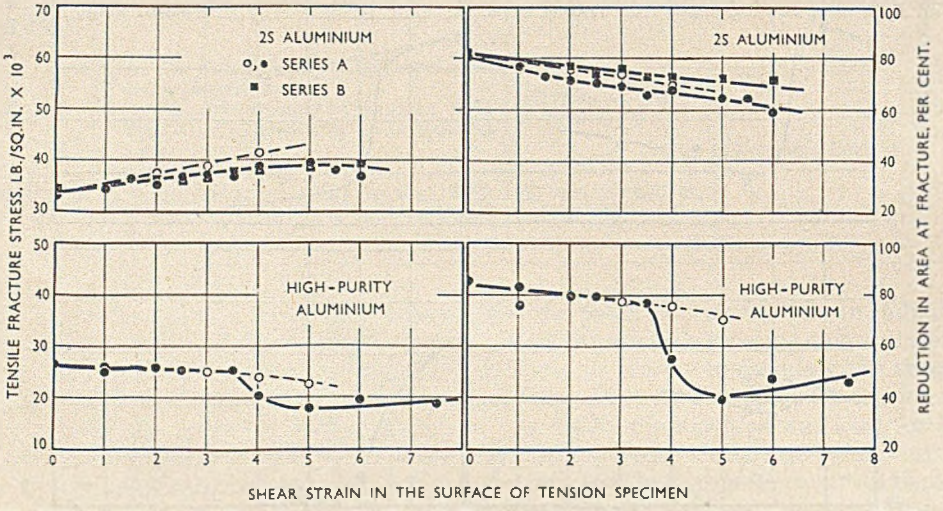

Fig. 5.—The Effect of Torsional Prestrain on the Tensile Fracture Stress and Ductility of Commercially Pure (2S) and High-Purity Aluminium Tested Directly after Prestraining.

Key.

● Specimens twisted to the indicated strain and then tested in tension.
○ Specimens twisted to the indicated strain, then completely untwisted, and tested in tension.

The effect of annealing after prestraining on the relationship between fracturing characteristics and prestrain is illustrated in Fig. 4 for the B series of Monel metal and of Armco iron. Measurements of the fracture angle, α, in Monel metal specimens are plotted in Fig. 3. Comparison of Figs. 2 and 4 shows that an intermediate anneal brought about a general increase in ductility, a continuous decrease in the fracture stress with increasing prestrain, and a higher value of the critical shear strain; there was also an enlargement of the fracture angle of the Monel metal specimens. The variation of fracture stress in the B series of Armco iron specimens showed the most noticeable effect of annealing. The difference between that metal and the others (Fig. 2) was eliminated by removing the cold work due to prestraining. Armco iron specimens could not be twisted to strains much greater than the critical value of series B, however, so that the helical fractures were not sharply defined, and angular measurements could not be obtained for plotting in Fig. 3. The important conclusion to be drawn from these experiments is that annealing does not eliminate the condition created by prestraining which is responsible for the helical fracture and for the reduced fracture stress and ductility.

The experiments with commercially pure (2S) aluminium gave completely unexpected results. Although specimens were prestrained over the widest possible range, the tensile fracturing characteristics remained substantially unaltered. As is shown in Fig. 5, the fracture stress rose and the ductility fell slightly with increasing prestrain. But there was no sudden change in either after any amount of prestraining. In addition, the tensile fracture of severely twisted 2S aluminium specimens, two of which are shown in Fig. 7 (Plate LXVII), was not of the distinct helical type observed in similarly prestrained specimens of other metals. After these results had been obtained, the experiments with series-B specimens were carried out. The processing of the specially prepared rod for this series differed considerably from usual commercial practice. But the fracturing characteristics, which are also included in Fig. 5, were still essentially the same as those of the series-A specimens, as was the appearance of the fractures.

The results of the experiments with high-purity (99·99%) aluminium were equally surprising. Because of its purity, of the near absence of inclusions, and of the results of the preceding work with 2S aluminium, the fracturing characteristics were not expected to be sensitive to prestrain. Yet curves in Fig. 5 show that they did undergo a fairly marked change, although only after a rather large torsional strain. The tensile fracture of such a specimen prestrained in excess of the critical amount, like the one shown in Fig. 8 (Plate LXVII), was closer to being helical than the fracture of any 2S specimen. When prestraining was accomplished by equal amounts of forward and reverse torsion, both fracturing characteristics and fracture appearance were more nearly like those of the unstrained specimen. Thus the response of high-purity aluminium to prestraining resembled that of the other metals more closely than it did that of 2S aluminium.

IV.—DISCUSSION OF RESULTS

A fibrous micro-crack structure, creating a mechanical anisotropy, again provides, as in the previous experiments with copper, an explanation of many of the experimental results. The orientation of such a crack structure in an unstrained specimen is simply represented by a straight line in the surface and parallel to the axis of the specimen. During twisting, this line follows a helical path which may be described by an angle α', similar to the angle α plotted in Fig. 3 (p. 436). After introducing a shear strain, γ, into the specimen surface, α' is equal to $(90° - \tan^{-1}\gamma)$. The agreement in Fig. 3 between the broken curve, describing the calculated variation of the fracture angle with prestrain, and the experimental curve implies, as before, that the helical tensile fracture occurs by separation over a surface of cracks whose directions are governed by the twisting. Only the reorientation of a crack structure by untwisting satisfactorily explains the different tensile behaviour of specimens prestrained by equal amounts of forward and reverse torsion.

The experiments with high-purity aluminium also suggest the presence of a fibrous crack structure. The less-marked effects of torsional prestrain on this metal may mean a less highly developed structure. They might also be attributed to the metal's inherently greater ductility; the flow preceding the tensile fracture, which is a function of the ductility, causing the helical crack structure to straighten slightly, thus tending to enlarge the fracture angle, α. Such reasoning also explains the greater critical shear strain observed after annealing prestrained specimens of Monel metal and Armco iron.

The results obtained with high-purity aluminium make it difficult to understand the tensile behaviour of prestrained 2S aluminium. Although less pure and containing more inclusions, it is relatively insensitive to torsional prestrain. Apparently inclusions are not a dominant cause of mechanical anisotropy in these materials, and, for unknown reasons, the small-scale crack structure is either absent or not sufficiently developed in 2S aluminium to be revealed by the torsion–tension testing procedure. It does not seem that a complete absence of the crack structure is a satisfactory explanation, however, for Jacquesson and Laurent,[14] by studying the dependence of fatigue strength on direction in rolled sheet of a commercially pure aluminium, found indications of a fibrous structure.

The suggestion has been made in the previous work with copper that the holes seen in the fracture surface of ductile metals broken directly in tension, as well as the holes and fibrous markings which are frequently observed in the helical type of fracture, may result from the enlargement of the micro-cracks to visible dimensions by the stresses associated with fracture. Similar patterns of holes and markings were also observed in varying degrees, in the fracture surfaces of the metals tested in this investigation. They were particularly noticeable in specimens of high-purity aluminium, but they were not at all conspicuous in 2S aluminium, as the photographs in Figs. 9 and 10 (Plate LXVII) show. Of course, this is only one difference between the two metals, yet it may reflect the relative intensity of the crack structure and thus help to explain the different tensile behaviour of these highly prestrained specimens.

V.—SUMMARY AND CONCLUSIONS

Fracturing test specimens in tension after prestraining in torsion has shown that a fibrous structure of flaws, which behave like cracks more than anything else, exists in 70 : 30 brass, nickel, Monel metal, and Armco iron. Before the presence of the crack structure in these materials becomes apparent, however, the maximum shear strain introduced by twisting must exceed a critical value that varies from about 1 to 2·5, depending upon the metal and on whether or not prestraining is followed by annealing. If the critical prestrain is exceeded, the fracture strength and ductility decrease with increasing prestrain, and the fracture occurs by separation across a helical surface which defines the position in the twisted specimen of an array of cracks originally aligned parallel to the specimen axis. After prestraining by equal amounts of forward and reverse torsion, the original orientation of the crack structure is unchanged, and this explains why the fracturing characteristics and fracture appearance are not greatly different from those encountered when there is no prestraining. Observations similar to these resulted from the same programme of testing with high-purity (99·99%) aluminium; a crack structure apparently exists in this metal as well as in the others. But for reasons that are not clear, the tensile behaviour of prestrained commercially pure (2S) aluminium gave no indication of the presence of such a structure.

Finding this condition in a number of relatively pure metals and single-phase alloys suggests that there is a definite limit to what can be accomplished in controlling the directionality of wrought materials through the quantity, size, shape, and distribution of extra phases in the microstructure.

ACKNOWLEDGEMENTS

The authors are indebted to the Office of Naval Research for sponsoring this research, and to Mr. R. S. Templin of the Aluminum Company of America for kindly providing the specially prepared rods of commercial and high-purity aluminium.

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THE FORMATION OF INTRACRYSTALLINE VOIDS IN SOLUTION-TREATED MAGNESIUM–ALUMINIUM ALLOYS *

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SYNOPSIS

It has been observed that occasionally the solution-treatment of cast magnesium–aluminium alloys results in the formation of small hexagonal voids in the centre of many grains. These voids have been shown to be orientated with their hexagonal axes parallel to the hexagonal axes of the crystals in which they occur. No explanation has been found for their formation, but it has been shown that in a sample of an alloy that does form cavities, the cavity formation increases to a maximum and then eventually vanishes with increasing homogenization. It has also been shown that the cavities are probably produced on cooling after solution-treatment rather than during the progress of the heat-treatment.

I.—INTRODUCTION

During the war years a considerable amount of research was undertaken with the object of reducing the scatter in the mechanical properties of magnesium-base castings. This work involved the examination of very many microsections of alloys that were then commonly used. It was observed that the alloy Elektron AZ91 (aluminium 9·5, zinc 0·5, manganese 0·3%) in the solution-treated condition often showed well-defined cavities in the central regions of many of the grains. The position of these cavities within the grains, and their well-marked geometric form, clearly distinguished them from other cavities, such as those due to microporosity.

These cavities, which were observed only in alloys that had been solution-treated, were found to be in the form of hexagonal-faced, parallel-sided voids with average dimensions of about 0·015 mm. across the flats of the hexagon and 0·008 mm. in height between the hexagonal faces. The dimensions, especially the height, varied considerably, and cavities up to twice the size mentioned have been observed. In the microsection the cavities were usually seen as rectangles, with occasional triangles, and more occasionally still, perfect hexagons. These shapes were consistent with random sectioning of the hexagonal prism shape already described. In many cases several cavities were observed within a single grain (Fig. 1, Plate LXVIII), and in such cases it was evident that the cavities had a common orientation. By giving specimens a precipitation-treatment at a relatively high temperature, thereby inducing a coarse precipitation in known crystallographic planes, it was possible to establish that the hexagonal voids and the hexagonal lattice of the crystal in which they were situated had the same orientation.[1] Fig. 2 (Plate LXVIII) shows the appearance of one of these cavities in a sample of Elektron AZ91 which had been solution-treated and then precipitation-treated at 250° C. From the appearance of the precipitate, it can be seen that the section had been cut very nearly parallel to the basal planes.

Cavities were never observed in any material that had not been solution-treated. They were observed most commonly in Elektron AZ91, less commonly in Elektron A8 (aluminium 8, zinc 0·5, manganese 0·3%), and in no other commercial alloy. With various experimental alloys such cavities were observed only when a substantial amount of aluminium was present and never in alloys free from aluminium.

Various attempts have been made to account for the formation of these cavities, but no satisfactory explanation has been found. However, in view of the recent observations of similar cavities in other alloy systems, the details of the occurrence of such cavities in magnesium-base alloys assume fresh interest. Thus, Bückle and Blin[2] have described similar cavities produced by the diffusion of aluminium and zinc in copper. Barnes[3] has described cavities produced by the diffusion of nickel in copper, and Brasunas[4] has reported sub-surface cavities in Inconel and 80 : 20 nickel–chromium alloy associated with the loss of chromium from the surface.

II.—EXPERIMENTAL WORK

As the production of these cavities was obviously connected in some way with the solution-treatment, it was decided to investigate the way in which cavity formation was connected with the two main variables in the solution-treatment, viz. time and temperature.

Since cavities were found in only a comparatively

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* Manuscript received 25 November 1952.

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small number of batches of AZ91, it was first necessary to obtain a stock of metal in the "as-cast" condition which was known to produce cavities on normal solution-treatment.

A batch of sand-cast test-bars from a single melt of AZ91 was prepared, and a small test-piece from each bar checked for cavity formation, until a suitable batch was obtained. This batch was of normal commercial purity. Typical impurity contents would be:

Si	$< 0.014\%$	Pb	$< 0.05\%$	Cd	$< 0.01\%$
Cu	$< 0.006\%$	Sn	$< 0.03\%$	Ag	$< 0.005\%$
Fe	$< 0.014\%$	Ca	$< 0.005\%$	Ni	$< 0.005\%$

1. EFFECT OF SOLUTION-TREATMENT TIME

When a suitably checked supply of material had been obtained, the following solution-treatments were carried out in a small air-circulating heat-treatment furnace: 5 hr. at 385° C., followed by heating at 420° C. for 5, 10, 15, 20, 25, 30, 40, 48, and 72 hr. The samples were allowed to cool freely in air. The preliminary treatment for 5 hr. at 385° C. was given in order to prevent any possible formation of liquid phase due to the presence of traces of the ternary eutectic.

On examination of the solution-treated samples, cavities were detected in all except one, which was heat-treated at 385° C. only. The cavities formed showed characteristics that varied with the time of treatment. At short times the cavities were less sharply defined and rather smaller, especially in height, than those seen after the normal 24-hr. treatment. In the samples solution-treated for short times, it was commonly observed that one or both of the hexagonal faces of the void was convex. This was at first thought to be due to distortion effects during polishing, but it was later found to be the true shape of the void, and to be characteristic of cavities formed after a short solution-treatment.

Between 15 and 40 hr. no difference in cavity formation could be detected. However, after 48 hours' treatment a decrease in the number of cavities produced was observed, and after 72 hours' treatment only a few cavities could be detected.

2. EFFECT OF SOLUTION-TREATMENT TEMPERATURE

Using the same "as-cast" material, the following solution treatments were carried out: 24 hr. at 360°, 380°, 400°, 420° C.; 3 hr. at 420° C. + 21 hr. at 440° C.; and 3 hr. at 420° C. + 21 hr. at 450° C.

When these samples were examined, the effect of increasing temperature was found to be very similar to that of increasing time. The sample that had been solution-treated at 360° C. was completely free from cavities, but much undissolved Mg_4Al_3 remained. That treated at 380° C. showed a few small narrow cavities, but solution was still far from complete. At 400°, 420°, and 440° C., a normal cavity formation was obtained. At 450° C. the cavities had decreased considerably in number, but the ones that were observed were rather larger than usual, most of the size increase being in the distance between the hexagonal faces.

These tests gave a plain indication that cavity formation diminished with increase in time or temperature of solution-treatment. Accordingly, combinations of increased time and temperature were tried, and it was found that cavities were completely absent after a treatment of 5 hr. at 385° C. + 16 hr. at 420° C. + 16 hr. at 450° C. This stepped treatment was used in order to prevent any incipient fusion at the higher solution-treatment temperature.

These results indicated that the tendency to produce cavities progressively increased with the degree of solution, and then gradually decreased until eventually no cavities at all were formed.

In view of these results, several tests were made to determine whether cavities already existing in solution-treated AZ91 could be removed by further heat-treatment. It was then discovered that cavities already existing could not be removed by either very long additional solution-treatments, or by solution-treatments at higher temperatures. This rather surprising result suggested that the cavities could not be produced progressively during solution-treatment and then closed up again with further heat-treatment. Instead it seemed most probable that the cavities were not produced at all during the heat-treatment, but were formed during cooling from the solution-treatment temperature, and were due to some condition set up in the metal by the heat-treatment.

3. EFFECT OF COOLING RATE AFTER SOLUTION-TREATMENT

If this supposition were true, cavity formation might be influenced by the rate of cooling from the solution-treatment temperature.

A number of samples of the "as-cast" AZ91 specimens were solution-treated for 8 hr. at 385° C. and 16 hr. at 420° C. and cooled from the solution-treatment temperature at different rates. The various rates of cooling were obtained by water-quenching, oil-quenching, air-cooling, air-cooling in a crucible packed with scrap metal, cooling in a laboratory tube-furnace, and cooling in a large industrial-type air-circulating furnace. In the two latter cases the specimens were allowed to cool in the furnace in which the solution-treatment had been carried out. The times to cool from 420° to 100° C. varied from a few seconds for the quenched samples to about 1 hr. for the large crucible packed with scrap, 6 hr. for the tube furnace, and 20 hr. for the large heat-treatment furnace. All the test-pieces, except those that were furnace-cooled, showed normal cavity formation. The two samples that were furnace-cooled were similar to each other, despite the big difference in their respective rates of cooling, and both showed only a very few cavities.

These were not of the characteristic appearance. They lacked the normal sharp straight outlines and showed rounded corners.

4. Effect of Grain-Size

Using a batch of AZ91 ingots which were known to give sand-cast test-bars that produced many cavities on normal solution-treatment, several chill-cast spectrographic pencils (4 mm. in dia.) and D.T.D. sand-cast test-bars were made. The grain-size of the spectrograph pencil castings was very fine, and that of the material in the heads of the sand-cast test-bars quite coarse. Samples cut from the spectrograph pencils, the test-bars, and the test-bar heads gave a fairly wide range of grain-sizes.

A number of solution-treatments were given to these pieces. In the spectrograph pencils, the solution of the Mg_4Al_3 occurred rapidly, solution being complete in about 6 hr. at 420° C. Samples from these pencils were solution-treated for 4, 6, 8, 16, and 24 hr. at 420° C. No cavities were found in any of these, although test-pieces cut from the sand-cast test-bars showed normal cavity formation after 24 hr. at 420° C.

The coarse-grained pieces cut from the feeder heads contained no cavities after 24 or 48 hr. at 420° C., but showed many abnormally large cavities after being heat-treated for 48 hr. at 440° C.

In addition to these experiments on the effect of grain-size, tests were also carried out on a block cast into a water-cooled mould, similar to that described by Northcott.[5] The walls were of thin steel sheet, and the base was a water-cooled copper plate. Solidification was almost perfectly unidirectional, and the rate of cooling varied over a wide range from bottom to top of the cast block. The structure also varied considerably. At the base of the block of AZ91 cast in this mould, the structure was of a fine-grained solid-solution type, almost free from any Mg_4Al_3. The grain-size gradually coarsened, and the quantities of massive Mg_4Al_3 increased towards the top portion of the block, where there was a structure very similar to that of a normal sand-cast test-bar. Complete vertical sections, approximately $3\frac{1}{2}$ in. long, were cut from these blocks and solution-treated for 8 hr. at 385° C. + 16 hr. at 420° C. Cavities were found only in the top $\frac{3}{4}$ in. of these strips.

5. Effect of Volume Change During Solution-Treatment

A number of tests in connection with other work had been carried out to determine the volume changes that occurred during the heat-treatment of magnesium-base casting alloys. These tests were made by measuring the dimensions of a rod 5 in. long by $\frac{3}{4}$ in. dia. machined from a sand-cast test-bar, before and after heat-treatment.

AZ91 would be expected to show a slight contraction on solution-treatment, but the results obtained were erratic; expansions occurring as often as contractions. When microspecimens were cut from the solution-treated test-pieces, it was found that the bars that had expanded showed considerable cavity formation, whereas the bars that had contracted were free from cavities. So far as it was possible to judge from the examination of microsections, the amount of expansion was roughly proportional to the extent of the cavity formation.

In view of these results several volume-change test-pieces were prepared from the reference batch of AZ91, and solution-treated for 8 hr. at 385° C. + 16 hr. at 420° C. + 24 hr. at 450° C. After this treatment no cavities were present, and the test-bars showed a consistent contraction.

III.—CAVITIES IN ALLOYS OTHER THAN AZ91

During the work, cavities were produced in AZ91, A8, and some experimental alloys containing 10 and 12% aluminium. It was not found possible to produce cavities in AZG (aluminium 5, zinc 3, manganese 0·3%) or AZM (aluminium 6, zinc 0·5, manganese 0·3%), although an old test-piece of AZG was found that showed a small number of cavities. In any alloy the cavity formation was found to vary from melt to melt, even if the structure and analysis were almost identical.

Numbers of scrap castings in solution-treated AZ91 were examined, and it seemed that cavities were produced more frequently in castings made under commercial conditions than in samples of the same alloy prepared under laboratory conditions.

A number of experimental alloys were made under laboratory conditions and attempts made to produce cavities by solution-treatments. The alloys and their solution-treatments were as follows:

Mg–8% Al–2% Zn	Solution-treated 24, 48, and 72 hr. at 350° C.
Mg–6% Al–4% Zn	
Mg–4% Al–6% Zn	
Mg–6% Al–2% Zn	
Mg–6% Al	Solution-treated 8, 16, and 24 hr. at 420° C.
Mg–4% Al	
Mg– 4% Zn	Solution-treated 24, 48, 72, and 96 hr. at 300° C.
Mg– 6% Zn	
Mg– 8% Zn	
Mg–10% Zn	
Mg–12% Zn	

None of these alloys formed cavities.

In general, it seems that cavity formation is associated with high aluminium contents. The effect is reduced by the replacement of aluminium with zinc, and binary magnesium–zinc alloys do not form cavities.

IV.—SUMMARY AND CONCLUSIONS

The work that has been described indicates that there is some relationship between the degree of solution and cavity formation. It is also very probable that the cavities are formed during cooling from solution-treatment rather than during the treatment. It is, however, evident that other factors are involved in the formation of cavities, and no explanation for the wide variation in cavity formation from one batch to another of the same alloy can be offered.

It is suggested that the cavity formation is in some way connected with the aluminium concentration gradient existing in any particular grain at the time that it is cooled from the solution-treatment temperature.

A sand casting in AZ91 contains an appreciable amount of massive Mg_4Al_3 and the average aluminium content of the α-phase crystals will probably be of the order of 4%. A considerable difference in aluminium content must exist in any grain between its centre and a boundary in contact with the massive Mg_4Al_3. Owing to the slow rate of diffusion, this gradient possibly builds up during the early stages of solution-treatment and exists to quite a marked extent after all the massive Mg_4Al_3 has been absorbed. Considerable further time at solution-treatment temperature is required to produce a uniform aluminium content throughout all the α grains.

That such a gradient does exist can be seen readily by precipitating to equilibrium a sample which has been solution-treated just sufficiently to remove the massive Mg_4Al_3. The precipitated Mg_4Al_3 will be seen to be greater in quantity near the grain boundaries, and the centre of the grains will be quite free. This effect can be seen in Fig. 2 (Plate LXVIII). The cavity formation always occurs in these zones which are obviously low in aluminium.

The fact that after solution-treatment for a long time or at a high temperature cavities are not formed will, on this hypothesis, be explained as being due to the reduction or elimination of the concentration gradient by the heat-treatment. This concentration gradient may set up internal stresses within the crystal, which on cooling after solution-treatment, and when some other condition is satisfied, gives rise to cavity formation. There has been no evidence to indicate the nature of this unknown factor, but it may be associated with melting conditions, and the gas content of the metal is an aspect of the problem worth further attention. However, it is unlikely, from the sharply defined geometrical shape of the cavities, that they could be associated with a high internal gas pressure.

A recent paper by Bückle and Blin [2] describes the production of cavities of a very similar nature in brass which had been given a diffusion treatment when in intimate contact with pure copper. The crystallographic relationship of the cavities in the brass to the brass grain, and that of the cavities in the magnesium to the magnesium grain suggest very strongly that a similar mechanism is responsible in both cases.

In the instances described in the present paper the cavities are associated with the diffusion of aluminium *into* magnesium, but in the case of Bückle and Blin's phenomenon it is a case of the diffusion of zinc *out of* brass. It will be noted that both cases have in common the production of a concentration gradient of the solute element, but whereas the solution of aluminium in magnesium causes a contraction of the parent lattice, solution of zinc in copper has the opposite effect.

It is hoped that this description of cavities produced in magnesium–aluminium alloys may help in obtaining a satisfactory explanation of the phenomenon.

ACKNOWLEDGEMENTS

The author is indebted to the Directors of Magnesium Elektron, Ltd., and to Captain A. C. Jessup, Chief Metallurgist, for permission to publish this paper. The work described was carried out several years ago under the general supervision of Dr. F. A. Fox.

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A NOTE ON THE MATHEMATICAL ANALYSIS OF CREEP CURVES*

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

SYNOPSIS

The Andrade equation $L_t = L_0(1 + \beta t^{1/3})\exp Kt$ and an equation containing a logarithmic term have been found to give good, but not perfect, fits of creep curves, obtained from constant-stress tests on a high-purity lead and a lead–1% tin alloy, for times up to 300 hr., although the Andrade equation was the better for longer times up to 5000 hr.

An attempt to correlate the values of the constants in the Andrade equation with the effects of grain-size and stress on creep behaviour was not possible for the high-purity lead owing to relatively large errors involved in the calculation of the constants. For the lead–1% tin alloy, however, it was found that the $\sqrt[3]{K}$ was linearly related to stress and relative grain-boundary areas, although no simple relationship existed between these variables and the β constant.

I.—INTRODUCTION

ANDRADE,[1] in his classical work on the flow of metals in 1910, derived a mathematical equation for creep curves obtained under conditions of constant stress. The constants in this equation are generally believed to have some physical significance,[2, 3] one of them has been associated with transient creep, and another is thought to represent flow of a viscous nature. Since Andrade's early work, many alternative creep equations have been proposed; that most frequently suggested contains a logarithmic function which has been associated, also, with transient creep.[4–6]

At the beginning of a recent and detailed investigation of the effect of constitutional factors on the creep properties of lead and lead alloys,[7] it was considered that mathematical analysis of the creep curves might facilitate interpretation of the results. In consequence, the creep tests were made under conditions of constant stress, instead of at constant load as in most investigations, to enable the creep curves to be so analysed.

In the present paper a comparison is made between the Andrade and the logarithmic equations to determine their relative accuracies in fitting the experimental results from tests of relatively long duration. An attempt is also made to correlate the effect of grain-size and stress on creep behaviour with the numerical values of the constants obtained by analysing the creep curves by the Andrade equation.

II.—EXPERIMENTAL PROCEDURE

The analysis of the high-purity lead (99·999% pure) and an alloy of this basis material with 1% tin, together with the experimental technique, have been reported elsewhere.[7–9] Both materials were extruded at various temperatures, in a laboratory extrusion press, to obtain a range of grain-sizes, the values of which were determined by a comparison method and by the statistical method due to Johnson.[10] The statistical method gives a spatial grain-size number which is inversely proportional to the actual grain-size.

III.—RESULTS OF EXPERIMENTS

1. COMPARISON OF THE ANDRADE AND LOGARITHMIC EQUATIONS

In the Andrade equation $L_t = L_0(1 + \beta t^{1/3})\exp Kt$, where L_t represents the length of the specimen at any time t, and L_0, β, and K are constants; L_0 is similar to the length of the specimen immediately after loading. Andrade himself suggests that the equation is only an approximation and fits a creep curve only for short times; his own tests lasted less than a working day. The present author, however, has shown previously that the equation is a good representation of the experimental results for times at least up to 1000 hr.[9]

The logarithmic equation may be written in a form similar to that due to Andrade, i.e. $L_t = A(1 + B \log t)\exp Ct$, where L_t represents the length of the specimen at any time t, and A, B, and C are constants.

The constants in both equations were calculated for a creep curve by solving three simultaneous equations obtained by substituting in the general equation the length of the specimen at three times separated by equal intervals. The elongation of a specimen at any time was determined to ±0·01%, and this limit of

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* Manuscript received 28 November 1952. The work described in this paper was made available to members of the British Non-Ferrous Metals Research Association in a confidential research report issued in November 1952.

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accuracy produced an error in the constants of the equations the maximum possible value of which was estimated by a method set out in the Appendix (p. 447). In comparing the curves calculated from the two equations with the experimentally determined creep curves in Figs. 1, 2, and 3, the shaded areas for each equation cover a range of values which are possible owing to the errors involved in the determination of the constants.

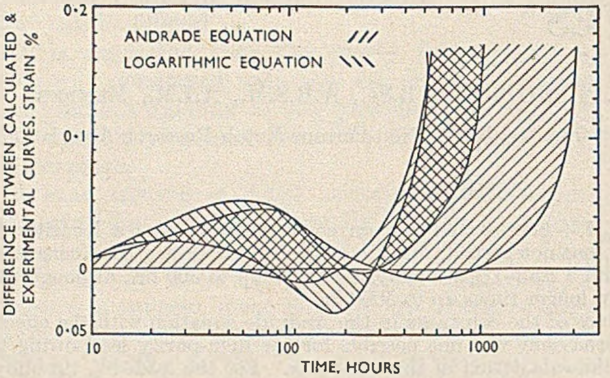

FIG. 1.—Comparison of the Andrade and Logarithmic Equations at Short Times for High-Purity Lead Extruded at 250° C. and Tested at 300 lb./in.^{2}. The constants in both equations were calculated from the values of the strains determined experimentally at 7, 143, and 279 hr. Strain at 1000 hr. was 1·85%.

For short times of test (Fig. 1), both equations are a good, but not perfect, representation of the experimental results for the high-purity lead. However, for longer times of test (Figs. 2 and 3), where the strains involved are approximately 3% for the high-purity lead and 20% for the 1% tin alloy, the Andrade equation is better than the logarithmic equation. In consequence, further work was limited to the Andrade equation.

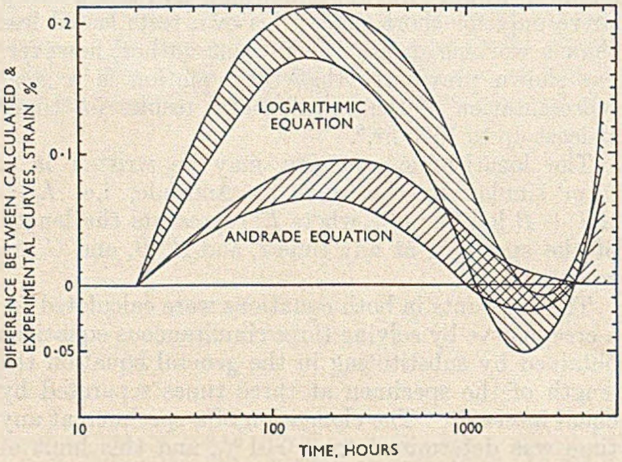

FIG. 2.—Comparison of the Andrade and Logarithmic Equations at Long Times for High-Purity Lead Extruded at 250° C. and Tested at 300 lb./in.^{2}. The constants in both equations were calculated from the values of the strains determined experimentally at 20, 1600, and 3180 hr. Strain at 5000 hr. was 3·25%.

2. VARIATION OF THE CONSTANTS IN THE ANDRADE EQUATION WITH DATA FROM WHICH THE CONSTANTS ARE CALCULATED

As the Andrade equation does not fit an experimental curve perfectly, the values of the constants in the equation depend on the data which are arbitrarily selected from the creep curve at three times, t_1, t_2, and t_3. This is shown in Table I, the effect on both constants being large for the high-purity lead. Table I also shows that the maximum

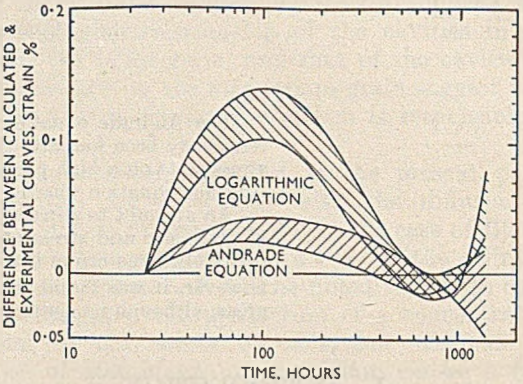

FIG. 3.—Comparison of the Andrade and Logarithmic Equations for Lead–1% Tin Alloy Extruded at 100° C. and Tested at 500 lb./in.^{2}. The constants in both equations were calculated from the values of the strains determined experimentally at 25, 550, and 1075 hr. Strain at 1400 hr. was 19·66%.

TABLE I.—*Variation of Constants in the Andrade Equation, for One Creep Curve, with Data from Which the Constants are Calculated.*

Material	Constants in the Andrade Equation Calculated from Experimental Data at the Following Times, hr.	Value of β, hr.$^{-1/3}$ × 10^4	Maximum Error in Value of β, hr.$^{-1/3}$ × 10^4	Value of K, hr^{-1} × 10^6	Maximum Error in Value of K, hr^{-1} × 10^6
High-purity lead extruded at 250° C. and tested at 300 lb./in.^{2}	$t_1 = 7$ $t_2 = 143$ $t_3 = 279$	10·5	±2·0	5·9	∓3·4
	$t_1 = 7$ $t_2 = 300$ $t_3 = 593$	11·7	±1·3	3·00	∓1·4
	$t_1 = 7$ $t_2 = 603$ $t_3 = 1200$	11·9	±0·9	2·68	∓0·67
	$t_1 = 20$ $t_2 = 1600$ $t_3 = 3180$	15·2	±0·7	0·79	∓0·25
Lead–1% tin alloy extruded at 100° C. and tested at 500 lb./in.^{2}	$t_1 = 25$ $t_2 = 120$ $t_3 = 215$	27·5	±4·2	109	∓6·7
	$t_1 = 25$ $t_2 = 250$ $t_3 = 475$	23·6	±2·1	114	∓2·3
	$t_1 = 25$ $t_2 = 400$ $t_3 = 775$	31·4	±1·5	106	∓1·2
	$t_1 = 25$ $t_2 = 550$ $t_3 = 1075$	31·9	±1·2	107	∓0·9

possible error in the calculation of the constants decreases as t_1, t_2, and t_3 are selected farther apart.

To minimize this effect of t_1, t_2, and t_3 in the correlations between the values of the constants and the factors influencing creep-resistance, the constants for different curves were calculated, as far as possible, from the same values of t_1, t_2, and t_3. To minimize the errors involved in the calculation of the constants, t_1, t_2, and t_3 were selected as far apart as possible on the curves before the inflection which is now generally considered to be due to recrystallization.

3. Effect of Grain-Size and Stress on the Constants in the Andrade Equation

Table II shows the effect of grain-size and stress on the values of the constants β and K for the high-purity lead and the 1% tin alloy. Where the errors in the calculations of a constant are the same for specimens of all grain-sizes tested at one stress, it indicates that the curves of all these specimens have been fitted from the same values of t_1, t_2, and t_3. The magnitude of these errors for the high-purity lead makes it impossible to attach any significance to the values of the constants for this material. The errors are large owing to the small values of t_1, t_2, and t_3 which had to be taken owing to the early recrystallization at all stresses. The errors for the 1% tin alloy were, however, comparatively small.

Table II shows that at each stress the values of the constants for the 1% tin alloy extruded at 300° C. were anomalous, since the values were larger than would be expected from those corresponding to the alloy extruded at lower temperatures. Although the value of β for the 1% tin alloy decreased as the stress was lowered, there was no obvious relationship between the two variables for material of constant grain-size. At a given stress there was no consistent

Table II.—*Variation of* β *and* K *with Grain-Size and Stress.*

Material	Stress, lb./in.^{2}	Temp. of Extrusion, °C.	Mean Grain-Size by Comparison Method, mm.^{2}	Grain-Size No. Johnson Method	β, hr.$^{-1/3}$ × 10^4	Approx. Error in β, hr.$^{-1/3}$ × 10^4	K, hr.$^{-1}$ × 10^6	Approx. Error in K, hr.$^{-1}$ × 10^6
High-purity lead	300	100	0·15	−0·4	6·74	± 2·0	6·61	∓3·4
		150	0·38	−2·0	8·94		4·4	
		200	0·86	−2·9	10·5		2·2	
		250	1·8	−3·4	10·5		15·3	
	500	100	As for stress of 300 lb./in.^{2}		34·7	±15	0·00	∓230
		150			34·7		0·00	
		200			30·9	± 4·6	36·0	∓ 23
		250			40·3	± 4·4	26·8	∓ 16
	750	100	As for stress of 300 lb./in.^{2}		62·8	±38	1220	∓1700
		150			71·3	±19	533	∓ 460
		200			69·4	± 9·4	448	∓ 160
		250			86·3		534	
	1000	100	As for stress of 300 lb./in.^{2}		101	±80	5040	∓4500
		150			125	±40	2780	∓1760
		200			131	±24	2090	∓ 910
		250			168		3870	∓ 910
Lead–1% tin alloy	300	100	0·0024	5·6	11·7	± 0·92	29·7	∓0·67
		150	0·013	4·1	5·28		8·85	
		200	0·023	2·8	3·21		2·84	
		250	0·11	0·2	2·07		1·17	
		300	0·39	−0·17	2·75		5·03	
	500	100	As for stress of 300 lb./in.^{2}		30·3	± 1·2	107	∓0·86
		150			19·6		30·4	
		200			14·6		11·2	
		250			16·8		5·04	
		300			23·1		6·69	
	750	100	As for stress of 300 lb./in.^{2}		65·9	± 2·0	350	∓3·8
		150			53·2		113	
		200			51·7		47·0	
		250			57·3		21·8	
		300			73·6		39·0	
	1000	100	As for stress of 300 lb./in.^{2}		114	± 3·0	1010	∓13·0
		150			130		251	
		200			117		119	
		250			121		30·2	
		300			165		156	

effect of grain-size on the value of β, except, perhaps, at the lowest stress, where the value increased as the grain-size diminished.

Fig. 4 shows that a relationship approximating to a straight line is obtained when $\sqrt[3]{K}$ is plotted against stress for specimens of the 1% tin alloy of constant grain-size. No straight line was obtained, however, for specimens with the largest grain-size.

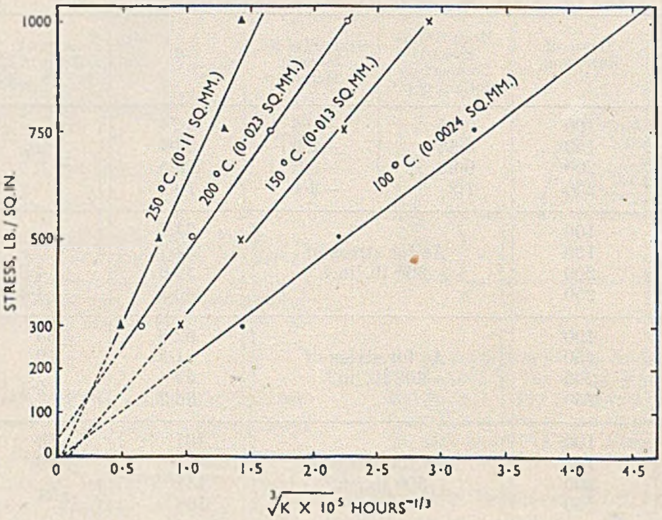

Fig. 4.—Relationship Between $\sqrt[3]{K}$ and Stress for the Lead–1% Tin Alloy Extruded at Different Temperatures. The average grain-sizes of the materials are shown in parentheses.

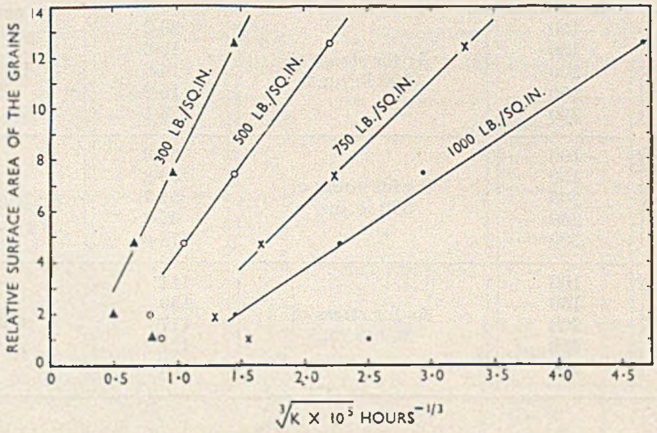

Fig. 5.—Relationship Between $\sqrt[3]{K}$ and the Relative Grain-Boundary Areas of the Lead–1% Tin Alloy Tested at Different Stresses.

In Fig. 5 a straight-line relationship is obtained by plotting $\sqrt[3]{K}$ against the relative grain-boundary areas of the grains per unit volume for the 1% tin alloy tested at a given stress. The values of K corresponding to the specimens with the largest grain-size do not fall on the straight lines, probably because of the anomalous behaviour of these specimens. The relative grain-boundary areas per unit volume were calculated from the grain-sizes determined by the Johnson method, which is more reliable than the comparison method.

III.—DISCUSSION OF RESULTS

In view of the effect of the arbitrary selection of t_1, t_2, and t_3 on the values of the constants in the Andrade equation, and the errors involved in their calculation the results for the effect of grain-size and stress on the constants in the Andrade equation obtained in this work must be accepted with reserve. Only the possible significance of the results is discussed.

Extrapolation of the straight-line relation between $\sqrt[3]{K}$ and stress in Fig. 4 suggests that K becomes zero at zero stress. This supports the general view that K is associated with flow which is of a viscous nature, although K is not linearly related to stress.

However, in Fig. 4 the values of K for specimens of constant grain-size tested at different stresses were not obtained from the same values of t_1, t_2, and t_3. Thus, the straight-line relationship between $\sqrt[3]{K}$ and stress may be in error, which would explain why this result differs from that found by Cottrell and Aytekin [11] and by Andrade and Jolliffe,[12] who showed log K to be directly proportional to stress.

The marked dependence of K on the relative grain-boundary area of the grains per unit volume suggests that K is associated with flow at the grain boundaries. On the other hand, the straight line relating $\sqrt[3]{K}$ and relative grain-boundary area in Fig. 5 also indicates that K is inversely proportional to the volume of the grains, since the grain-boundary area per unit volume is inversely proportional to the diameter of the grains. The straight lines in Fig. 5 cut the abscissa at positive values of K when the relative grain-boundary area of the grains per unit volume is zero, i.e. when the specimens are single crystals. This suggests that K flow can occur within the grains as well as, possibly, at grain boundaries. Cottrell and Aytekin [11] have shown that K flow occurs in single crystals of zinc.

IV.—SUMMARY AND CONCLUSIONS

1. The Andrade and logarithmic equations were found to be a good representation of the experimentally determined creep curve for short times up to about 300 hr. The Andrade equation, however, was the better of the two for longer times up to 5000 hr.
2. As the Andrade equation does not represent the experimental results perfectly for tests of long duration, the values of the constants in the equation depend on the arbitrarily selected experimental data. For this reason and because of the errors involved in the calculation of the values of the constants, the significance of the correlations between the values of the constants with grain-size and stress must be accepted with reserve.
3. The errors involved in the determination of the constants, β and K, in the Andrade equation for the high-purity lead, due to the ease with which this material recrystallizes under test, made it impossible to correlate the calculated values with either grain-size or stress.
4. For the lead–1% tin alloy, β decreased as the stress was reduced, though not in any simple way. There was no consistent effect of grain-size on β. For specimens of constant grain-size, $\sqrt[3]{K}$ was linearly related to stress and, for constant stress, to relative grain-boundary areas of the grains per unit volume.

ACKNOWLEDGEMENTS

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

APPENDIX

Estimation of the Errors in the Calculation of the Constants in the Andrade Equation

Error in β.

The Andrade equation $L_t = L_0(1 + \beta t^{1/3}) \exp Kt$ is fitted to the creep curves by the quadratic method. Solving three simultaneous equations, obtained by substituting in the general equation the length of the specimen at three times (t_1, t_2, t_3), separated by equal intervals, results in the following quadratic in β :

$$\frac{L_2{}^2}{L_1 L_3} = \frac{(1 + \beta t_2{}^{1/3})^2}{(1 + \beta t_1{}^{1/3})(1 + \beta t_3{}^{1/3})}$$

If γ is the strain at time t, then :

$$\frac{(1 + \gamma_2)^2}{(1 + \gamma_1)(1 + \gamma_3)} = \frac{(1 + \beta t_2{}^{1/3})^2}{(1 + \beta t_1{}^{1/3})(1 + \beta t_3{}^{1/3})} \cdot \qquad (1)$$

If γ is small then the left-hand side of (1)

$$\bumpeq \exp (2\gamma_2 - \gamma_1 - \gamma_3)$$

The approximate form of the Andrade equation is :

$$\gamma = \beta t^{1/3} + Kt$$

so that $\gamma \geqslant \beta t^{1/3}$

∴ The right-hand side of (1) can be written $\exp \beta(2t_2{}^{1/3} - t_1{}^{1/3} - t_3{}^{1/3})$

$$\therefore 2\gamma_2 - \gamma_1 - \gamma_3 \bumpeq \beta(2t_2{}^{1/3} - t_1{}^{1/3} - t_3{}^{1/3})$$

Let ϵ be the error in γ at any time t and ϵ_β be the error in β resulting from that in γ. Then the maximum value of ϵ_β is given by :

$$2(\gamma_2 + \epsilon) - (\gamma_1 - \epsilon) - (\gamma_3 - \epsilon) \bumpeq (\beta + \epsilon_\beta)(2t_2{}^{1/3} - t_1{}^{1/3} - t_3{}^{1/3})$$

Then $4\epsilon \bumpeq \pm\, \epsilon_\beta(2t_2{}^{1/3} - t_1{}^{1/3} - t_3{}^{1/3})$.

Error in K.

To find K, the equation $\frac{L_2}{L_1} = \frac{1 + \beta t_2{}^{1/3}}{1 + \beta t_1{}^{1/3}} \exp K(t_2 - t_1)$ is solved.

By reasoning as above,

$$\gamma_2 - \gamma_1 \bumpeq \beta(t_2{}^{1/3} - t_1{}^{1/3}) + K(t_2 - t_1).$$

Let ϵ_K be the error in K resulting from the error in γ and β.

Then ϵ_K is a maximum when

$$\gamma_2 - \gamma_1 + 2\epsilon \bumpeq (\beta + \epsilon_\beta)(t_2{}^{1/3} - t_1{}^{1/3}) + (K + \epsilon_K)(t_2 - t_1)$$

Then $2\epsilon \bumpeq \epsilon_\beta(t_2{}^{1/3} - t_1{}^{1/3}) + \epsilon_K(t_2 - t_1)$

$$\epsilon_K \bumpeq \mp \frac{2\epsilon - \epsilon_\beta(t_2{}^{1/3} - t_1{}^{1/3})}{(t_2 - t_1)}$$

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1463

EQUILIBRIUM RELATIONS AT 460° C. IN ALUMINIUM-RICH ALLOYS CONTAINING 0–7% COPPER, 0–7% MAGNESIUM, AND 0·6% SILICON *

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SYNOPSIS

The equilibrium isothermal at 460° C. is given for quaternary alloys rich in aluminium and containing 0–7 wt.-% magnesium, 0–7 wt.-% copper, and constant (0·6 wt.-%) silicon. The phases encountered are the aluminium-rich solid solution, $CuAl_2$, Mg_2Si, Si, the ternary phase Al_2CuMg (*S*), and the quaternary phase $Al_5Cu_2Mg_8Si_6$ (*Q*). The general form of the isothermal containing 0·6% silicon is similar to that previously reported for alloys containing 1·2% silicon (*J. Inst. Metals*, 1952–53, 81, 209), but the phase fields are all translated towards the aluminium-rich corner.

I.—INTRODUCTION

The present paper contains the results of an investigation into the equilibrium constitution at 460° C. of aluminium-rich alloys which contain copper in the range 0–7%, magnesium 0–7%, and with constant (0·6%) silicon, all compositions being by weight. The results of previous workers, and the experimental methods used in the present work, have been described in a previous paper.[1] The chemical analyses were again conducted by Messrs. Johnson, Matthey and Co., Ltd., and the author's thanks are due to Mr. A. R. Powell for his care and attention to this aspect of the work.

II.—EXPERIMENTAL RESULTS

Fig. 1 shows the phase fields which occur on annealing aluminium-rich alloys containing 0·6% silicon for four weeks at 460° C. In Fig. 1 the point *X* on the vertical axis is taken from the ternary aluminium–copper–silicon system and corresponds to the $(\alpha + \mathrm{Si})/(\alpha + \mathrm{Si} + \theta)$ phase boundary for alloys containing 0·6% silicon. Points *Y* and *Z* on the horizontal axis are taken from the ternary aluminium–magnesium–silicon system, and correspond respectively to the $(\alpha + \mathrm{Si})/(\alpha + \mathrm{Si} + \mathrm{Mg_2Si})$ and the $(\alpha + \mathrm{Si} + \mathrm{Mg_2Si})/(\alpha + \mathrm{Mg_2Si})$ phase boundaries for alloys containing 0·6% silicon.

As in the previous work, the critical alloys were analysed for magnesium and silicon only, since work on this type of system has shown that the synthetic composition with respect to copper may be readily obtained. The points in Fig. 1 which correspond to analysed alloys are denoted by the letter *A* placed above the points. The limits of accuracy of Fig. 1 are: as regards temperature 460° ± 2° C., and as regards composition silicon 0·6 ± 0·03% with 0·016% iron.

If Fig. 1 is compared with the corresponding isothermal for alloys containing 1·2% silicon,[1] it will be seen that in both isothermals the number and general disposition of the phase fields are similar. The chief differences are that the $(\alpha + \mathrm{Si} + \theta + Q)$, $(\alpha + \mathrm{Mg_2Si} + \theta + Q)$, $(\alpha + \mathrm{Si} + Q)$, $(\alpha + \mathrm{Mg_2Si} + Q)$, $(\alpha + \mathrm{Mg_2Si} + \mathrm{Si} + Q)$, and $(\alpha + \mathrm{Si} + \mathrm{Mg_2Si})$ phase fields are narrower in the 0·6% silicon isothermal than in the 1·2% silicon isothermal, and are also displaced to a lower magnesium composition. The phase boundaries of Fig. 1 have been drawn to agree with the amounts of the various phases present, and the symbols used in Fig. 1 are the same as those used in Fig. 4 of the previous paper.[1]

It is convenient to designate the individual points in Fig. 1 by the notation: (*x*, *y*), meaning an alloy containing 0·6% silicon with *x* wt.-% copper and *y* wt.-% magnesium (synthetic compositions); and (*x*, *y*)*A*, meaning that the alloy was analysed for magnesium and the analytical result is used in designating the alloy. Point (1·0, 4·31)*A* is important in fixing the $(\alpha + \mathrm{Mg_2Si})/(\alpha + \mathrm{Mg_2Si} + S)$ phase boundary, since it contains α, Mg_2Si, and a very small amount of the ternary *S* phase. The straight boundaries of the four-phase field $(\alpha + \mathrm{Mg_2Si} + \theta + S)$ appear to be satisfactorily determined by the alloys: (6·0, 2·00)*A*, which contains a minute trace of *S* existing as very small crystals; (4·0, 2·11)*A*, which contains a trace of *S* existing as occasional large crystals; (6·0, 3·03)*A*, which contains a trace of θ uniformly distributed throughout the alloy; and (4·0, 2·65)*A*, which contains a trace of θ distributed around the edge of the specimen only. This is

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* Manuscript received 19 November 1952.

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probably because the surface of the specimen loses magnesium by volatilization during annealing. The surface layers of all specimens are removed before analysis; hence point (4·0, 2·65)*A* has been plotted as an ($\alpha + Mg_2Si + S$) alloy near to the ($\alpha + Mg_2Si + \theta + S$) boundary. The boundaries of the four-phase field ($\alpha + Mg_2Si + \theta + S$) shown in Fig. 1 appear to interpolate well between the corresponding boundaries in the 1·2% silicon section and the ($\alpha + Mg_2Si + S$) boundaries of the ternary aluminium–magnesium–copper system at 460° C. Evidence for the existence of the narrow ($\alpha + Si + \theta + Q$) and ($\alpha + Mg_2Si + \theta + Q$) phase fields is given by alloys (6·0, 0·31)*A* and (4·0, 0·25), both of which contain minute amounts of silicon, and alloys (7·0, 1·0), (6·0, 1·09)*A*, and (3·0, 1·07)*A*, of which the first two contain minute traces of *Q* distributed throughout the alloy, whilst the third contains small crystals of *Q* at the rim of the specimen only.

The boundaries of the ($\alpha + Q$) phase field have been drawn to coincide in size and shape (but not position) with the ($\alpha + Q$) phase field of the 1·2% silicon section, and it is satisfactory to note how well the boundaries so drawn agree with the experimental results. No experimental evidence has been obtained for the existence of the ($\alpha + Mg_2Si + Si + Q$) phase field, which is very small but which must exist in the 0·6% silicon section in order to satisfy the geometrical requirements of the previous work.

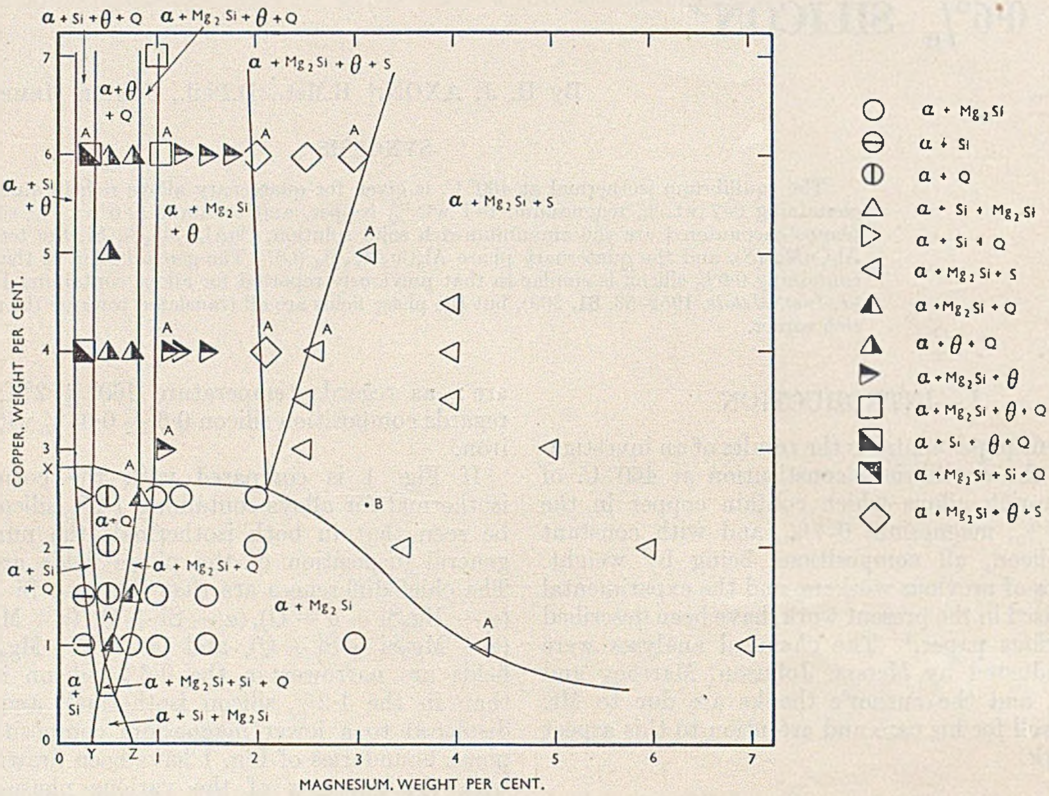

Fig. 1.—The 460° C. Isothermal for Aluminium-Rich Alloys Containing Magnesium, Copper, and Silicon. All alloys in this isothermal contain 0·6 wt.-% silicon. The point *B* indicates the apex of the ($\alpha + Mg_2Si + \theta + S$) phase field for alloys containing 1·2 wt.-% silicon. *Note added in proof.*—The attention of the reader is drawn to the fact that the scale of this figure is slightly different from that used for the figures in the earlier paper.[1] This arose from factors outside the control of the author, who hopes inconvenience will be minimized by this note.

ACKNOWLEDGEMENTS

The author thanks Professor F. C. Thompson for research facilities and continued encouragement, and The Royal Society for a grant in aid of the research.

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THE CONSTITUTION OF NICKEL-RICH ALLOYS OF THE NICKEL–CHROMIUM–ALUMINIUM SYSTEM*

1464

By A. TAYLOR,† Ph.D., F.Inst.P., Member, and
R. W. FLOYD,‡ B.Sc., A.I.M., Member

SYNOPSIS

The equilibrium relationships in nickel–chromium–aluminium alloys containing more than 50 at.-% nickel have been studied over the temperature range 750°–1150° C. It has been shown that the phase fields of the nickel primary solid solution, γ, and of the β solid solution based on NiAl both contract as the temperature falls, whereas that of the γ' phase based on Ni_3Al extends. The equilibrium between the γ and β phases which exists from the solidus breaks down at about 1000° C., giving rise to equilibrium between γ' and α-chromium by a four-phase reaction: $\beta + \gamma \rightleftharpoons \alpha + \gamma'$. The ordering of the γ phase is enhanced by the presence of aluminium atoms, and for alloys with more than 10 at.-% aluminium the ordering temperature is over 1150° C. The equilibrium between the γ and γ' phases is such that over a range of compositions the lattice parameters of the phases are almost identical.

I.—INTRODUCTION

This investigation of the nickel–chromium–aluminium system was undertaken as part of a more extensive survey of nickel-base alloys containing chromium, titanium, and aluminium. Accounts have already been given of the nickel–chromium–titanium [1] and nickel–titanium–aluminium [2] systems, with which the nickel–chromium–aluminium system links to form the boundaries of the nickel corner of the quaternary nickel–chromium–titanium–aluminium system. Revisions of the binary nickel–chromium and nickel–aluminium systems which have been described in these papers have been incorporated in the present work.

Nickel and chromium form a simple eutectiferous system with primary solid solutions which are extensive at the eutectic temperature of 1345° C. (Fig. 1). As the temperature falls, so the solubility of chromium in the nickel decreases steadily from 50 at.-% at the solidus to 38 at.-% at 750° C. In the neighbourhood of Ni_3Cr the face-centred cubic nickel solid solution, γ, is ordered below 540° C.[3] The solubility of nickel in body-centred cubic α-chromium decreases rapidly from 32 at.-% at the solidus to 2 at.-% at 750° C. The modifications to the phase diagram proposed by Bloom and Grant [4] have not been substantiated by observations made by the present authors.

The binary nickel–aluminium phase diagram is shown in Fig. 2. The solubility of aluminium in the nickel solid solution falls from 21 at.-% at 1385° C. to 11·9 at.-% at 750° C. The ordered face-centred cubic Ni_3Al phase, γ', forms via a peritectic reaction,[5, 6] and its phase field widens to 23·3–27·3 at.-% aluminium at 750° C. The phase field of the ordered body-centred cubic NiAl phase (β) which has a wide range at 1400° C. becomes restricted at lower temperatures. The boundary with the ($\beta + \gamma'$) region lies at 30 at.-% aluminium at 1350° C. and 38 at.-% aluminium at 850° C.[7] The other phases occurring in the system have no bearing on the part of the nickel–chromium–aluminium system which is described here.

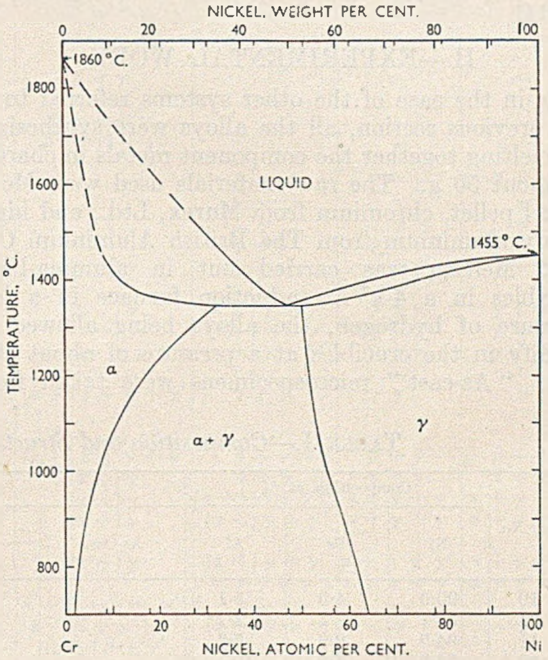

Fig. 1.—The Nickel–Chromium Phase Diagram.

The only phase of the chromium–aluminium system

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* Manuscript received 8 November 1952.

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‡ The Mond Nickel Co., Ltd., Birmingham.

which is involved in the present work is the chromium solid solution (α). Bradley and Lu [8] have shown that the solubility of aluminium in chromium is about 45 at.-% above 900° C. and 28 at.-% below 800° C.

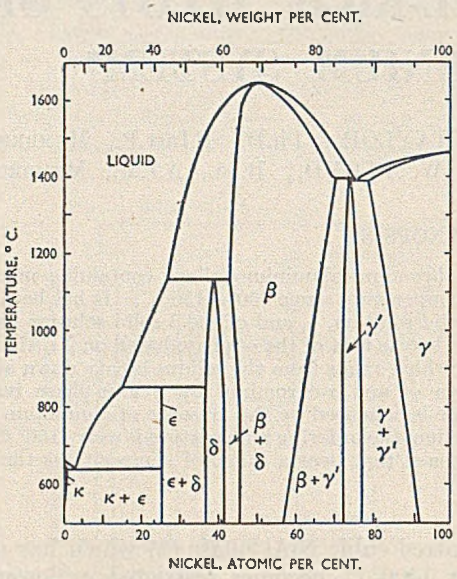

Fig. 2.—The Nickel–Aluminium Phase Diagram.

II.—EXPERIMENTAL WORK

As in the case of the other systems referred to in the previous section, all the alloys were synthesized by melting together the component metals in charges of about 50 g. The raw materials used were Mond nickel pellet, chromium from Murex, Ltd., and high-purity aluminium from The British Aluminium Co., Ltd. Melting was carried out in alumina-lined crucibles in a 4-kVA. induction furnace in a low pressure of hydrogen, the alloys being allowed to solidify in the crucibles at a pressure of about 10^{-3} mm. "As-cast" microspecimens were taken from the ingots, and then the remainder of each ingot was annealed at 1150°–1250° C. for 4 days to promote homogeneity. The majority of the alloys were analysed chemically for the main elements, and a few were checked for minor impurities, which could in all cases be accounted for by those present in the initial charge. With few exceptions good agreement was obtained between actual and nominal compositions.

The system was first explored by a study of the X-ray-diffraction patterns of powder samples slowly cooled from 900° C. over 2 weeks. Subsequently X-ray and microscopic techniques were employed on specimens isothermally heat-treated as follows:

	Temp., °C.	Time
X-ray samples . .	750	3 weeks
,, ,, . .	850	3 weeks
Microspecimens . .	850	3 weeks
,, . .	1000	4 days
,, . .	1150	2 days

The X-ray-diffraction patterns were obtained in a 9-cm. Debye–Scherrer camera, using manganese $K\alpha$ radiation, and accurate lattice parameters were obtained by the usual extrapolation method.[9] Microspecimens were prepared by grinding to 600 grit, followed by polishing on γ-alumina or diamond dust. After trying out a number of etching reagents, it was found that a 5% hydrofluoric acid and 10% glycerine aqueous solution, used electrolytically, satisfactorily revealed the phases in alloys over a wide range of composition without staining or pitting the surfaces of the specimens.

For studying the system at the lower temperatures the X-ray and microscopic methods of examination were used together. In the first place the arrangement of the phase fields was mapped out from a qualitative estimate of the X-ray evidence. Examination of the microspecimens was of particular value in elucidating the structure of the alloys consisting of the two face-centred cubic phases, γ and γ', which could not in all cases be differentiated by visual examination of

Table I.—*Composition and Structure of Nickel–Chromium–Aluminium Alloys.*

Alloy No.	Composition, at.-%			Observed Structure					
					Quenching Temperature				Slowly Cooled
	Ni	Cr	Al	As-Cast	1150° C.	1000° C.	850° C.	750° C.	
10	90·0	4·9	5·1	...	...	...	γ	γ	γ
34	85·1	12·2	2·7	...	...	...	γ	γ	γ
11	84·9	9·9	5·2	...	...	...	γ	γ	γ
98	84·3	5·2	10·5	...	...	γ	γ + γ'	γ + γ'	γ + γ'
99	82·8	9·6	7·6	...	...	γ	γ	γ	γ
100	79·9	14·8	5·3	...	...	...	γ	γ	γ
101	79·4	12·8	7·8	...	...	γ	γ + γ'	γ + γ'	γ + γ'
12	79·3	9·9	10·8	...	...	...	γ + γ'	γ + γ'	γ + γ'
102	76·6	9·9	13·4	...	γ	γ + γ'	γ + γ'	γ + γ'	γ + γ'
103	76·6	5·2	18·2	(β +) γ + γ'	γ	γ + γ'	γ + γ'	γ + γ'	γ + γ'
35	74·6	22·4	3·0	...	...	...	...	γ	γ
13	74·9	19·8	5·2	...	...	...	γ	γ	γ
74	75·0	17·2	7·8	...	...	γ	γ + γ'	γ + γ'	γ + γ'
146	73·9	18·2	7·9	...	...	...	γ + γ'	γ + γ'	γ + γ'
73	74·9	14·8	10·3	...	...	γ	γ + γ'	γ + γ'	γ + γ'

TABLE I.—*continued.*

Alloy No.	Composition, at.-%			Observed Structure					
	Ni	Cr	Al	As-Cast	Quenching Temperature				Slowly Cooled
					1150° C.	1000° C.	850° C.	750° C.	
72	74·5	12·2	13·3	...	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
71	74·6	9·8	15·6	...	...	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
36	73·9	7·5	18·6	...	γ	$\gamma + \gamma'$	γ'	γ'	γ'
14	74·8	4·9	20·3	...	$\gamma + \gamma'$	γ'	γ'	γ'	γ'
219	75·1	2·8	22·1	$\beta + \gamma + \gamma'$	$(\gamma +)\,\gamma'$	$\gamma + \gamma'$	...	...	...
218	74·5	2·7	22·8	$\beta + \gamma + \gamma'$	γ'	γ'	...	...	...
209	73·6	5·0	21·4	...	γ'	γ'	...	...	...
37	73·8	2·8	23·4	...	...	γ'	γ'	γ'	γ'
251	71·7	17·3	11·0	...	...	γ'	$\gamma + \gamma'$	$\gamma + \gamma'$	...
250	72·1	15·0	12·9	...	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$	...
38	71·9	9·4	18·7	...	γ	$\gamma + \gamma'$	γ'	γ'	γ'
294 *	72·5	7·5	20·0	$\beta + \gamma + \gamma'$	$\gamma + \gamma'$	γ'	...	...	...
210	71·8	4·9	23·4	...	γ'	γ'	...	...	...
217	72·1	2·4	25·5	...	$(\beta +)\,\gamma'$	γ'	...	...	...
39	69·9	27·4	2·7	...	...	...	γ	γ	γ
15	69·9	24·7	5·4	...	...	...	γ	γ	γ
108	69·7	21·7	8·6	...	γ	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
149	68·4	22·0	9·6	...	...	...	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
107	69·0	20·3	10·7	...	γ	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
40	68·6	17·5	13·9	...	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	γ'	γ'
106	68·9	15·1	16·0	...	γ	$\gamma + \gamma'$	γ'	γ'	γ'
147	69·8	12·6	17·6	...	γ	$\gamma + \gamma'$	γ'	γ'	γ'
299 *	70·0	10·0	20·0	$\beta + \gamma$	$(\beta +)\,\gamma + \gamma'$	$(\beta +)\,\gamma + \gamma'$	...	...	...
105	68·8	10·2	21·0	...	$\gamma + \gamma'$	$\gamma + \gamma'$	γ'	γ'	γ'
295 *	70·0	7·5	22·5	$\beta + \gamma + \gamma'$	γ'	γ'	...	...	...
211	71·2	5·1	23·7	$\beta + \gamma'$	γ'	γ'	...	...	...
298 *	70·0	5·0	25·0	$\beta + \gamma + \gamma'$	$\beta + \gamma'$	$\beta + \gamma'$	...	...	...
216	70·9	2·5	26·6	$\beta + \gamma + \gamma'$	$\beta + \gamma'$	$\beta + \gamma'$	...	...	...
300 *	67·5	12·5	20·0	$\beta + \gamma$	$\beta + \gamma + \gamma'$	$\beta + \gamma + \gamma'$	...	...	...
296 *	67·5	7·5	25·0	$\beta + \gamma + \gamma'$	$\beta + \gamma + \gamma'$	$\beta\,(+ \gamma) + \gamma'$	...	...	...
16	68·5	4·9	26·6	...	...	...	$\beta + \gamma'$	$\beta + \gamma'$	$\beta + \gamma'$
41	64·7	32·5	2·8	...	...	...	γ	γ	γ
17	64·9	29·3	5·8	...	...	...	γ	γ	γ
113	65·1	27·2	7·7	...	γ	γ	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
145	65·5	24·0	10·5	...	γ	$(\alpha +)\,\gamma$	$\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
112	64·3	24·9	10·8	...	...	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
111	65·0	22·6	12·4	...	γ	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
110	64·6	20·2	15·2	$\beta + \gamma$	γ	$(\alpha +)\,\gamma + \gamma'$	$(\alpha +)\,\gamma + \gamma'$	$\gamma + \gamma'$	$\gamma + \gamma'$
302 *	65·0	17·5	17·5	$\beta + \gamma$	$\beta + \gamma$	$\beta + \gamma + \gamma'$	...	...	...
109	64·3	16·7	19·0	$\beta + \gamma$	$(\alpha +)\,\beta + \gamma$	$(\alpha +)\,\beta + \gamma + \gamma'$	$\alpha + \gamma'$	γ'	γ'
301 *	65·0	15·0	20·0	$\beta + \gamma$	$\beta + \gamma$	$\beta + \gamma + \gamma'$	...	...	...
42	64·9	9·9	25·2	...	...	$(\alpha +)\,\beta + \gamma + \gamma'$	$\beta + \gamma'$	$\beta + \gamma'$	$\beta + \gamma'$
43	60·1	34·8	5·1	γ	...	$(\alpha +)\,\gamma$	$\alpha + \gamma$	$\alpha + \gamma$	$\alpha + \gamma$
117	60·1	31·7	8·2	...	γ	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma\,(+ \gamma')$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
148	61·5	29·0	9·5	...	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma + \gamma'$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
116	60·6	29·0	10·4	γ	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma$	$(\alpha +)\,\gamma + \gamma'$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
115	59·8	24·9	15·3	$\beta + \gamma$	...	...	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
303 *	60·0	25·0	15·0	$\beta + \gamma$	$(\alpha +)\,\beta + \gamma$	$\alpha + \beta + \gamma\,(+ \gamma')$	...	...	...
44	62·6	16·1	21·3	...	$(\alpha +)\,\beta + \gamma$	$\alpha + \beta + \gamma + \gamma'$	$(\alpha +)\,\gamma + \gamma'$	γ'	γ
297 *	62·5	7·5	30·0	$\beta + \gamma'$	$\beta + \gamma'$	$(\alpha +)\,\beta + \gamma'$	...	...	...
88	59·1	19·7	21·2	$\beta + \gamma$	$\beta + \gamma$	$\beta + \gamma$	$\alpha\,(+ \beta) + \gamma'$	$\alpha + \gamma'$	$\alpha + \gamma'$
114	58·2	10·8	31·0	β	$(\alpha +)\,\beta$	$(\alpha +)\,\beta + \gamma'$	$\alpha + \beta + \gamma'$	$\alpha + \beta + \gamma'$	$\alpha + \beta + \gamma'$
118	55·5	34·2	10·3	...	...	...	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
151	55·0	34·8	10·2	$\alpha + \gamma$	...	$\alpha + \gamma$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$	$\alpha + \gamma + \gamma'$
304 *	55·0	30·0	15·0	$\beta + \gamma$	$\alpha + \beta + \gamma$	$\alpha + \beta + \gamma$	...	...	...
305 *	52·5	37·5	10·0	$\alpha + \gamma$	$\alpha\,(+ \beta) + \gamma$	$\alpha\,(+ \beta) + \gamma$	...	...	...
119	50·3	25·6	24·1	$\alpha + \beta + \gamma$	$\alpha + \beta + \gamma$	$\alpha + \beta + \gamma + \gamma'$	$\alpha + \beta + \gamma'$	$\alpha + \beta + \gamma'$	$\alpha + \beta + \gamma'$
223 *	45·0	40·0	15·0	$\alpha + \beta + \gamma$	...	...	...	...	...
174	34·0	34·2	31·8	$\alpha + \beta$	...	...	$\alpha + \beta$	$\alpha + \beta$	...
175	25·4	54·6	20·0	$\alpha + \beta$	...	...	$\alpha + \beta$	$\alpha + \beta$	...
176	16·2	69·1	14·7	$\alpha + \beta$	...	...	$\alpha + \beta$	$\alpha + \beta$	...

* Composition nominal.

the diffraction patterns. Similarly, microscopical examination provided a ready means of identifying the body-centred cubic α and β phases, especially in alloys in which they were present in such small amounts that the superlattice lines characterizing the β phase would be below the visibility limit. The accurate location of the boundaries of the γ and γ' phase fields was obtained from lattice-parameter data which also enabled the corners of the three-phase triangles to be located.

TABLE II.—*Lattice-Parameter Data.*

(a) *Alloys in the γ Phase Field at 750° C.*

Alloy No.	Composition, at.-%			Heat-Treatment	Lattice Parameter, kX
	Ni	Cr	Al		
...	100	...	...	...	3·5168
1	90·2	9·8	...	SC [a] 750° C.	3·5273 3·5270
10	90·0	4·9	5·1	SC 750° C.	3·5302 3·5302
2	84·8	15·2	...	SC 750° C.	3·5315 3·5317
34	85·1	12·2	2·7	SC	3·5314
11	84·9	9·9	5·2	SC 750° C.	3·5342 3·5352
99	82·8	9·6	7·6	SC 750° C.	3·5388 3·5396
3	80·2	19·8	...	SC 750° C.	3·5374 3·5380
100	79·9	14·8	5·3	SC 750° C.	3·5412 3·5419
4	75·1	24·9	...	SC 750° C.	3·5441 3·5437
35	74·6	22·4	3·0	SC	3·5442
13	74·9	19·8	5·3	SC 750° C.	3·5471 3·5483
5	70·2	29·8	...	SC 750° C.	3·5509 3·5518
39	69·9	27·4	2·7	SC	3·5519
15	69·9	24·7	5·4	SC 750° C.	3·5533 3·5543
6	65·2	34·8	...	SC 750° C.	3·5582 3·5591
41	64·7	32·5	2·8	SC	3·5569
17	64·9	29·3	5·8	SC 750° C.	3·5582 3·5610

(b) *Alloys in the ($\gamma + \gamma'$) Phase Field at 750° C.*

Alloy No.	Composition, at.-%			Heat-Treatment	Lattice Parameter, kX	
	Ni	Cr	Al		γ	γ'
98	84·3	5·2	10·5	SC 750° C.	3·5362 3·5382	3·545 3·547
101	79·4	12·8	7·8	SC 750° C.	3·5411 3·5423	3·547 ...
12	79·3	9·9	10·8	SC 750° C.	3·5399 3·5406	3·549 3·555
102	76·7	9·9	13·4	SC 750° C.	3·5128 ...	3·5506 3·5510
103	76·6	5·2	18·2	SC 750° C.	... ...	3·5514 3·5508
74	75·0	17·2	7·8	SC 750° C.	3·5480 3·5513	... ...
146	73·9	18·2	7·9	SC	3·5509	...
73	74·9	14·8	10·3	SC 750° C.	3·5497 3·5186	
72	74·5	12·2	13·3	SC 750° C.	... 3·5478	3·5517
71	74·6	9·8	15·6	SC 750° C.	... ...	3·5521 3·5515
251	71·7	17·3	11·0	750° C.	3·5516	
250	72·1	15·0	12·9	750° C.	3·5532	
108	69·7	21·7	8·6	SC 750° C.	3·5527 3·5546	
149	68·4	22·0	9·6	SC	3·5555	
107	69·0	20·3	10·7	SC 750° C.	3·5558 3·5558	
113	65·1	27·2	7·7	SC 750° C.	3·5604 3·5644	3·553 3·555
145	65·5	24·0	10·5	SC	3·5611	3·5549
112	64·3	24·9	10·8	SC 750° C.	3·5627 3·5657	3·5561 3·5560
111	65·0	22·6	12·4	SC 750° C.	3·5620 3·5656	3·5552 3·5568
110	64·6	20·2	15·2	SC 750° C.	3·5613 ...	3·5544 3·5554

(c) *Alloys in the γ' Phase Field at 750° C.*

Alloy No.	Composition, at.-%			Heat-Treatment	Lattice Parameter, kX
	Ni	Cr	Al		
36	73·9	7·5	18·6	SC	3·5519
14	74·8	4·9	20·3	SC 750° O.	3·5532 3·5537
37	73·8	2·8	23·4	SC	3·5579
22	74·0	...	26·0	SC	3·5600
38	71·9	9·4	18·7	SC	3·5515
40	68·6	17·5	13·9	SC	3·5537
106	68·9	15·1	16·0	SC 750° C.	3·5556 3·5559
147	69·8	12·6	17·6	SC	3·5564
105	68·8	10·2	21·0	SC	3·5593
109	64·3	16·7	19·0	SC 750° C.	3·5570 3·5569
44	62·6	16·1	21·3	SC	3·5585

(d) *Alloys Containing α or β Phase at 750° C.*

Alloy No.	Composition, at.-%			Heat-Treatment	Structure	Lattice Parameter, kX		
	Ni	Cr	Al			γ	γ'	α
16	68·5	4·9	26·6	SC 750° C.	$\beta + \gamma'$ $\beta + \gamma'$	... ...	3·5639 3·5653	... ...
42	64·9	9·9	25·2	SC	$\beta + \gamma'$	...	3·5659	...
7	60·7	39·3	...	SC 750° C.	$\alpha + \gamma$ $\alpha + \gamma$	3·5657 3·5646	... ...	... ...
43	60·1	34·8	5·1	SC	$\alpha + \gamma$	3·5620	...	...
117	60·1	31·7	8·2	SC	$\alpha + \gamma + \gamma'$	3·5604	3·554	2·8752
148	61·5	29·0	9·5	SC	$\alpha + \gamma + \gamma'$	3·5645	3·5570	...
116	60·6	29·0	10·4	SC	$\alpha + \gamma + \gamma'$	3·5615	3·5545	2·8762
115	59·8	24·9	15·3	SC	$\alpha + \gamma + \gamma'$	3·559	3·5535	2·878
88	59·1	19·7	21·2	SC	$\alpha + \gamma'$	...	3·5604	2·876
8	55·5	44·5	...	SC 750° C.	$\alpha + \gamma$ $\alpha + \gamma$	3·5656 3·5646	... ...	... ...
118	55·5	34·2	10·3	SC	$\alpha + \gamma + \gamma'$	3·5607	3·5539	2·8764
						β		
114	58·2	10·8	31·0	SC	$\alpha + \beta + \gamma'$	2·5634	3·5660	2·877
119	50·3	25·6	24·1	SC	$\alpha + \beta + \gamma'$	2·8616	3·5651	2·877

[a] SC = Slowly cooled from 900° C. in 2 weeks.

The form of the phase diagram at higher temperatures was determined from micro-examination alone, as the characteristics of the constituents had already been established and difficulty was anticipated in satisfactorily heat-treating filings at the higher temperatures for X-ray examination.

The chemical composition of the alloys examined and their structures are given in Table I. Table II details the lattice-parameter data from which the 750° C. isothermal was determined.

III.—THE EQUILIBRIUM AT 750° AND 850° C.

The greater part of the investigation was devoted to establishing the 750° C. isothermal for alloys containing more than 50 at.-% nickel, with an accuracy of about $\pm\frac{1}{2}$%, and in particular the relationship between the γ and γ' phases. The diagram, shown in Fig. 3, is based on the findings on alloys annealed at 750° C. and on those slowly cooled from 900° C. to room temperature in 14 days, which were very similar. The region denoted by γ consists of the primary solid solution of chromium and aluminium in nickel and has a face-centred cubic structure which can become ordered in the neighbourhood of Ni$_3$Cr.[3] The γ' phase is based on the structure of Ni$_3$Al, which has an ordered face-centred cubic lattice in which nickel atoms occupy cube faces and aluminium and chromium atoms occupy cube corners. The γ' phase field is quite extensive, and the structure will absorb 19 at.-% chromium in place of aluminium and nickel.

Between these single-phase fields lies the two-phase $(\gamma + \gamma')$ region, in which the tie-lines link together the phase structures which are in equilibrium with each other at 750° C. A two-phase equilibrium also exists between the γ phase and the body-centred cubic α phase based upon chromium. A three-phase $(\alpha + \gamma + \gamma')$ region links the three structures together, followed by the two-phase $(\alpha + \gamma')$ region. The γ' phase also co-exists in equilibrium with the body-centred cubic β phase, based upon NiAl, which has a CsCl-type superlattice. The two-phase $(\beta + \gamma')$ region is indicated in the diagram in addition to the $(\alpha + \beta + \gamma')$ region, but an accurate determination of the β phase boundary and of the two- and three-phase regions linked with it lay outside the scope of the present investigation.

A more extended region covering the triangle Ni–Cr–NiAl is shown in Fig. 4 to display the general relationship of the nickel-rich alloys with the main body of the system.

Observations on alloys quenched at 850° C. show small changes in the equilibrium relationships from those obtaining at 750° C. As shown in Fig. 5, the γ phase takes into solution approximately 0·5% more aluminium and 2% more chromium than alloys slowly cooled or quenched from 750° C. The boundary at the chromium-rich end of the γ' phase recedes towards the composition Ni_3Al, narrowing the $(\alpha + \gamma')$ two-phase region and widening the $(\gamma + \gamma')$ region.

It has been shown by previous workers [5, 10, 11] that in the binary nickel–aluminium system the phase field of the γ solid solution is separated from that of

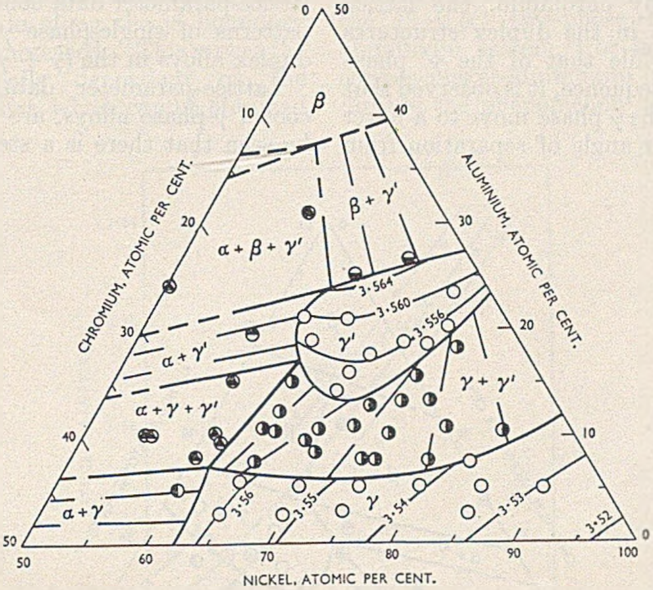

Fig. 3.—The Nickel–Chromium–Aluminium Phase Diagram : Isothermal Section for 750° C.

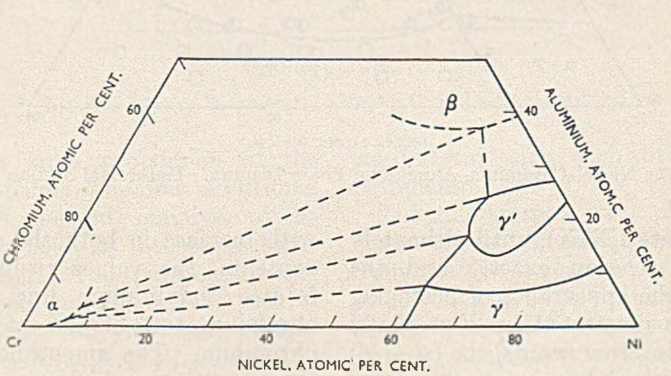

Fig. 4.—The Nickel–Chromium–Aluminium Phase Diagram : Isothermal Section for 750° C., Showing Relationship Between Nickel-Rich Alloys and the α Phase.

the γ' solid solution by a two-phase $(\gamma + \gamma')$ region. The lattice parameters of the co-existing structures in alloys isothermally treated at 750° C. ($a = 3{\cdot}54$ kX for γ and $a = 3{\cdot}56$ kX for γ') are sufficiently different to produce complete resolution of the two sets of high-angle (311) $K\alpha$ doublets in Debye–Scherrer photographs taken with manganese radiation (Fig. 13 (*a*), Plate LXIX). However, if the aluminium level is maintained in the region of 12·5 at.-% and then nickel is replaced by chromium, the lattice parameter of the γ phase in the duplex structures increases quite rapidly, while that of the γ' phase changes but little. In consequence, it is observed that the (311) $K\alpha$ doublets of the γ phase move to a lower Bragg angle, reducing their angle of separation from those of γ' (Fig. 13 (*b*), Plate LXIX), until ultimately the doublets appear to overlap exactly and the diffraction patterns give the appearance of belonging to single-phase alloys, as illustrated by Fig. 13 (*c*) and (*d*) (Plate LXIX). In other words, the $(\gamma + \gamma')$ gap of the nickel–aluminium binary system appears to be enclosed by a binodal curve, and the γ phase of the nickel–chromium system seems to stretch across the composition triangle as far as Ni_3Al. Thus γ' would appear to be merely the ordered portion of the γ phase field, as it appeared to be in the nickel–iron–aluminium system [12] until it was re-examined by classical metallographic methods.[7]

At still higher chromium contents, the group of end-doublets again breaks up into two pairs having progressively increased separation as the chromium content increases (Fig. 13 (*e*), Plate LXIX). This would seem to indicate the existence of a second $(\gamma + \gamma')$ two-phase field, in which it is the γ' constituent that has the lower lattice parameter, and this two-phase region is bounded by a side of the $(\alpha + \gamma + \gamma')$ three-phase triangle. An alternative explanation of the sequence of diffraction patterns described above would be that γ and γ' form two quite distinct phase fields separated by a $(\gamma + \gamma')$ two-phase region in which the lattice parameters of the co-existing phase structures become equal over a restricted range of composition. Of these alternatives the second can be shown to be correct from a consideration of the lattice-parameter data computed from the diffraction patterns of single-phase γ and γ' alloys, and from duplex alloys in the $(\gamma + \gamma')$ phase field.

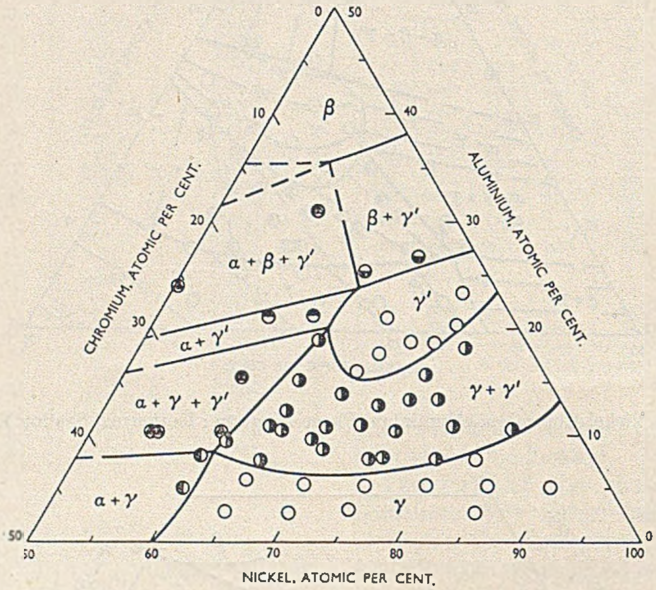

FIG. 5.—The Nickel–Chromium–Aluminium Phase Diagram : Isothermal Section for 850° C.

Lattice-parameter data for a series of slowly cooled γ-phase alloys, are plotted in Fig. 6. It will be seen that there is a steady increase in parameter with increase in both the chromium and aluminium contents, the values rising continuously along the 5 at.-% aluminium line from 3·525 kX at zero chromium to 3·56 kX at approximately 30 at.-% chromium. The smoothness of the curves is consistent with the single-phase character of the region. Sections across the diagram from Ni_3Cr to Ni_3Al and along a parallel direction at the 70 at.-% nickel level (Fig. 7) show that the lattice parameter of the ordered face-centred cubic γ' phase based on Ni_3Al, decreases steadily as aluminium is replaced by chromium, but, in each case, there are two well-defined discontinuities before the portion relating to the γ phase field is reached. These discontinuities clearly mark the boundaries of the γ and γ' phase fields. The fact that the diffraction pattern of alloy No. 108 (Fig. 13 (*d*), Plate LXIX) has a single-phase appearance yet gives a parameter which does not lie on either the γ or γ' branch of the curve is evidence that this alloy is, in fact, two-phase $(\gamma + \gamma')$, but with the co-existing structures having closely similar lattice parameters.

Further evidence of the complete separation of the γ and γ' phase fields by a $(\gamma + \gamma')$ field is to be found in the isoparametric contours drawn in Fig. 3. The intersection of these contours with the phase boundaries gives the lattice parameters of the ends of the tie-lines linking the co-existing phase-structures. It is evident from the diagram that the lattice parameters increase much more rapidly along the boundary of the γ phase than they do along the γ' boundary as the chromium content is increased. This makes the tie-line directions principally dependent on the parameter of the γ constituent, which can be obtained quite satisfactorily for a large proportion of the alloys in the two-phase field. The few additional tie-lines whose directions depend mainly on the parameter of the γ' phase field, fit into the scheme in a self-consistent manner. The parameter data show that the tie-lines start parallel to the nickel–aluminium side of the system and pivot round Ni_3Al until they lie askew across the $(\gamma + \gamma')$ field. The alloys of these skew tie-lines are the ones which are apparently single-phase, and it is found that their lattice parameters are a weighted mean of the parameters of the two co-existing phases at their extremities. The differences between the γ and γ' parameters are of the order of one part in 1000, so that resolution of their individual reflections is impossible; but the two-phase structure of these alloys is confirmed beyond doubt by their microstructures (Fig. 16, Plate LXIX). Careful inspection of the line shapes indicates a slight degree of unsharpness which does not normally occur with homogeneous, well-annealed, single-phase alloys. Beyond this region, the tie-lines close up fan-wise about the low-nickel end of the $\gamma/(\gamma + \gamma')$ boundary.

A feature of a number of the diffraction patterns, which adds complications to their interpretation by visual inspection alone, is the fact that slowly cooled alloys lying well within the γ single-phase field all show faint but well-defined superlattice lines based on a highly ordered Ni_3Cr structure, though with a few chromium atoms replaced by aluminium. These superlattice lines overlap those of the γ' phase, so that their presence on a powder photograph is no evidence of the alloy being duplex, as is the case in the $(\beta + \beta')$ two-phase field of the iron–nickel–aluminium system,[12] or in the $(\alpha + \alpha_1)$ field of the copper–nickel–aluminium system,[13] both of which display the phenomenon of two similar structures in equilibrium with each other having equal lattice parameters.

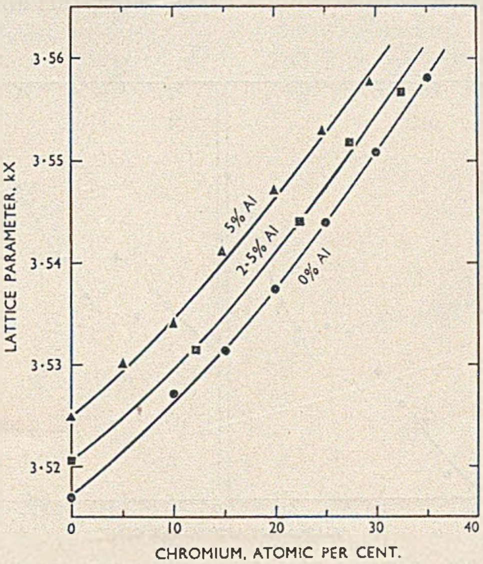

Fig. 6.—Lattice Parameters of Slowly-Cooled γ-Phase Alloys.

The micrographic work places the phase boundaries in practically the same positions as those determined from the X-ray parameter data, provided that due attention is given to the changes taking place between the melting point and the temperature of annealing. This is particularly necessary in the case of alloys at the chromium-rich end of the region in which are to be found small amounts of α phase persisting either from the melt or from the breakdown of $(\beta + \gamma)$ eutectic (discussed below).

The microstructures of alloys in the $(\gamma + \gamma')$ region, annealed at 850° C., vary considerably with composition. In all cases the one phase appears as a precipitate within the other, but the form and distribution of the precipitate differ from alloy to alloy. The γ' phase appears to be formed by precipitation from the γ matrix in all the two-phase alloys examined.

Examples of the types of structure observed in the $(\gamma + \gamma')$ phase field are shown in Figs. 14–17, (Plate LXIX). The dispersed globular precipitation of

γ' in γ represented in Fig. 14 is the most common structure. In some of the alloys the dispersed precipitate lies in regular rows along crystallographic planes of the matrix and may also be present in the form of needles (Fig. 15). The fine precipitate in the structure shown appears to lie on the octahedral (111) planes of the matrix, a relatively massive form occurring at the grain boundaries. An unusual appearance is presented by alloys at the middle of the $(\gamma + \gamma')$ field (Fig. 16), in which the precipitate assumes a rectangular shape and appears to form on the cube faces (100) of the matrix structure. There are only a few alloys in which this type of structure has been observed at this temperature, and in no case is the difference in spacings of the two phases sufficient to distinguish the diffraction pattern from that of a single phase. The growth of one face-centred cubic phase on the (100) planes of another is rather unexpected, since precipitation normally occurs on the (111) planes.

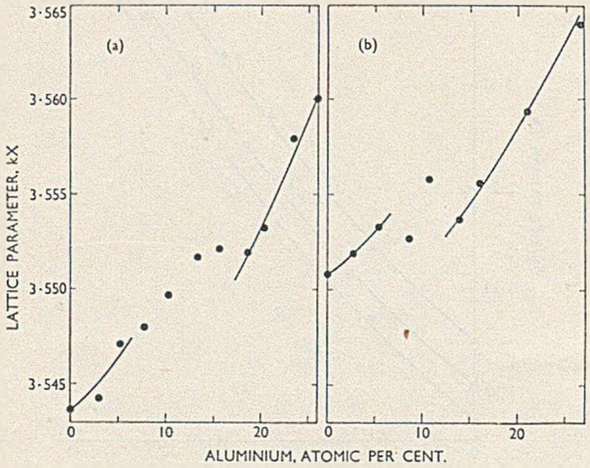

FIG. 7.—Lattice Parameters of Slowly-Cooled Alloys Containing (*a*) 75 at.-% Nickel and (*b*) 70 at.-% Nickel.

A further variation in the $(\gamma + \gamma')$ structures is exhibited by alloy No. 103 (Fig. 17). From the positions of the boundaries and tie-lines of this two-phase field, the predominating phase must be γ', and what appears to be the infilling between the grains is really the primary solid solution from which the γ' has precipitated.

On the aluminium side of the γ' phase lies the two-phase $(\beta + \gamma')$ region. From microscopical examination, it is evident that the β phase extends over a much greater range of composition at temperatures near the solidus than at the lower temperatures. The β phase in an alloy such as No. 16 has a striated structure in the "as-cast" state and, on annealing at lower temperatures, the γ' phase separates out as plates lying parallel to each other within the darker-etching pools of β phase (Fig. 18, Plate LXX). The $(\beta + \gamma')$ region is limited by the appearance of the α phase when the chromium content is more than about 9%. That the α phase precipitates from β on cooling in the same way as does γ' is shown in Fig. 19 (Plate LXX), where it can be seen as small round particles unaffected by the etching reagent, while the γ' phase appears as a middle tone. At higher chromium contents, lakes of α phase formed from the melt can be seen in addition to α globules precipitated from the β phase (Fig. 20, Plate LXX).

Adjacent to the $(\alpha + \beta + \gamma')$ phase field lies a two-phase $(\alpha + \gamma')$ region which should include alloy No. 88, according to the X-ray evidence. The microstructure of this alloy (Fig. 21, Plate LXX) obviously contains more than two phases, the dark pools consisting of β phase which has not completely transformed. This is not unexpected, for, as discussed later, this alloy solidifies as primary dendrites of β with $(\beta + \gamma)$ eutectic (Fig. 26, Plate LXXI), both phases transforming at lower temperatures. The proportion of β phase remaining after annealing at 850° C. is less than might be judged at first sight, as the pools of β contain large amounts of α and γ'. It would appear that on further annealing the β disappears entirely, as no trace can be detected in the X-ray-diffraction pattern, presumably because filings for X-ray examination attain equilibrium more rapidly than solid microspecimens on account of the heavy cold work imparted to them, and in this particular case the X-ray data are more reliable.

The $(\alpha + \gamma')$ region is separated from the $(\alpha + \gamma)$ region by an area in which the alloys consist of three phases: $\alpha + \gamma + \gamma'$. Microstructures of two alloys in this region, Nos. 115 and 151 (Figs. 22 and 23, Plate LXX), show distinct differences in the appearance of the constituents. In both structures the α phase appears white, but in the one alloy its boundaries are clearly defined, whereas in the other they are quite indistinct. The appearance of the γ and γ' phases also shows marked differences. To explain these structures, it is necessary to refer to the original "as-cast" structure of the alloys. Alloy No. 115 solidified as primary γ dendrites with an infilling of $(\beta + \gamma)$ eutectic (Fig. 24, Plate LXXI), and the pattern of the structure was not changed on annealing at 1150° C. Because of this, annealing at 850° C. resulted in the β phase breaking down to the α and γ' phases, as in alloy No. 88 (Fig. 21, Plate LXX), while the greater part of the γ phase transformed to γ'. The $(\gamma + \gamma')$ pattern obtained is similar to that exhibited by alloy No. 103 (Fig. 17, Plate LXIX), the finely divided constituent resembling a "precipitate" being in fact the remnants of the original γ phase. On the other hand, alloy No. 151 in the "as-cast" state consisted mainly of $(\alpha + \gamma)$ eutectic with excess primary γ. Annealing this at 850° C. did not change the amount of α to a great extent, but the γ split up into the γ and γ' phases in approximately equal proportions. Much of the γ' phase occurs as well-formed particles, but the appearance of parts of the matrix suggests that some is there in a very finely divided form, probably accompanied by some α phase. Consideration of these two structures, along with a number of others, has indicated that, in the presence of α phase, the γ phase may sometimes be differentiated from γ' by the lack of definition of the α/γ interface, as compared with α/γ' interface.

In the iron–nickel–aluminium system[7, 12] a single-phase field extends across the composition triangle to link the β-iron phase with the ordered β-NiAl phase. This suggested that in the nickel–chromium–aluminium system the α-chromium phase would likewise link with β-NiAl to give a single-phase field which would form a natural limit to investigations of the system starting from the nickel corner. Consequently, a brief survey was made along the line Cr–NiAl, and it was found that the microstructures of the cast alloys (Fig. 29, Plate LXXI) corresponded to those of a eutectiferous binary system with the eutectic point lying near 30 : 40 : 30 at.-% Ni–Cr–Al and that the duplex structures persisted in alloys annealed at temperatures down to 750° C. This result indicated that the half of the ternary system defined by the corners Ni, Cr, and NiAl could be regarded from a practical viewpoint as self-contained.

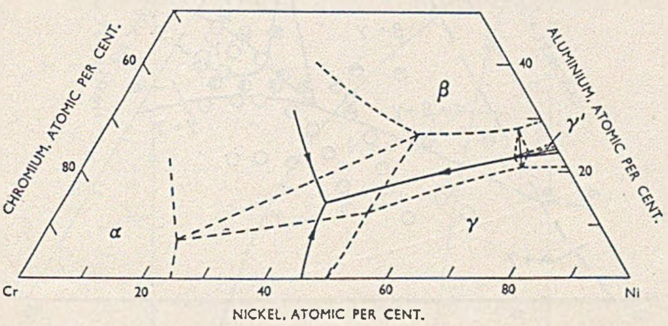

FIG. 8.—Tentative Diagram of Fields of Primary Crystallization.

At 750° C. both α and β phase fields extended well into the diagram, and lattice-parameter measurements indicate that in the two-phase $(\alpha + \beta)$ alloy No. 175 the parameters of the two phases do not differ by more than one part in 400, which is much less than in alloys in the three-phase $(\alpha + \beta + \gamma')$ region.

IV.—THE SOLIDIFICATION DIAGRAM

From the foregoing observations on the structures of alloys at 750° and 850° C., it will be seen that to interpret the microstructures of alloys annealed at 850° C. it was necessary from time to time to refer to the structures of the "as-cast" alloys. This led to a fairly extensive examination of alloys in the cast state, from which it was possible to map out tentatively the fields of primary crystallization within the composition range Ni–Cr–NiAl as shown in Fig. 8. Subsequently the equilibria at 1150° and 1000° C. were determined in order to relate the phase relationships at the solidus with those at 750° C.

Reference has already been made to alloys which in the cast state consist of either γ or β phase with an infilling of a $(\beta + \gamma)$ eutectic. No two-phase equilibrium exists between the γ and β phase in the binary nickel–aluminium system, although it has been observed in the copper–nickel–aluminium[14] and iron–nickel–aluminium[7] systems. In the nickel–aluminium system there occur two three-phase reactions,

$$\beta + \text{liq.} \rightleftharpoons \gamma' \text{ at } 1395^\circ \text{ C.}$$

and

$$\text{liq.} \rightleftharpoons \gamma + \gamma' \text{ at } 1385^\circ \text{ C.}$$

involving liquid of almost the same composition.[5] As chromium is added to the nickel–aluminium alloys, the temperatures of the above reactions fall and at the same time approach each other, as do also the compositions of the participating liquids, until both become identical at about 8 at.-% chromium. At this temperature (1340° ± 10° C.) there is a four-phase invariant plane in which the following reaction takes place :

$$\text{liq.} + \gamma' \rightleftharpoons \gamma + \beta$$

Below this plane there are the three-phase equilibria ($\beta + \gamma + \gamma'$) and (liq. + $\beta + \gamma$), which give rise to the eutectic reaction:

$$\text{liq.} \rightleftharpoons \beta + \gamma$$

Although the four-phase reaction given above is in accord with the findings of Bradley [7] on the iron–nickel–aluminium system, Alexander [14] regards the reaction as untenable, since he considers it improbable that the liquid and the γ' phases would react to form β in view of the peritectic reaction of the binary system by which γ' is formed from liquid and β. However, his proposed reaction

$$\text{liq.} \rightleftharpoons \beta + \gamma + \gamma'$$

is even more improbable, as it represents a ternary eutectic.

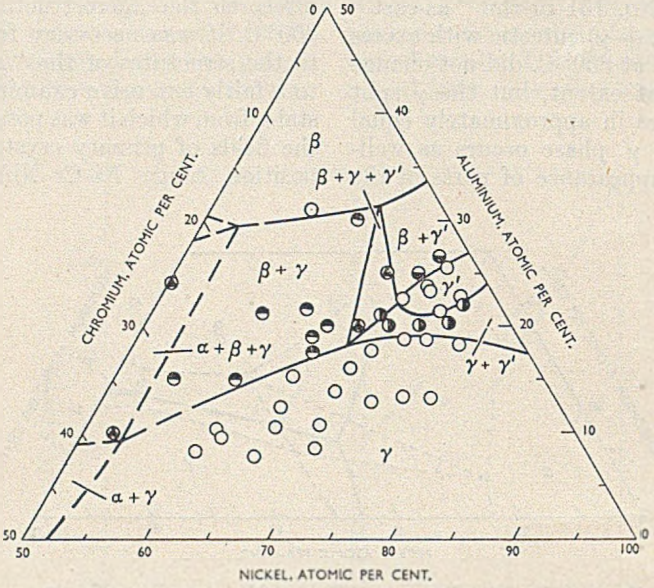

Fig. 9.—The Nickel–Chromium–Aluminium Phase Diagram: Isothermal Section for 1150 °C.

It was at first thought that the ($\beta + \gamma$) eutectic might run into the ($\alpha + \gamma$) eutectic, forming a trough in the liquidus surface across the nickel corner. However, when it was found that the α and β phases existed separately from the melt, it was realized that this construction was impossible. The existence of the three binary eutectics,

$$\text{liq.} \rightleftharpoons \alpha + \gamma$$

$$\text{liq.} \rightleftharpoons \beta + \gamma$$

and

$$\text{liq.} \rightleftharpoons \alpha + \beta$$

suggested that they might run together to form a ternary eutectic,

$$\text{liq.} \rightleftharpoons \alpha + \beta + \gamma$$

That this was in fact the case was confirmed by the structures of alloys lying near the centre of the Ni–Cr–NiAl triangle (Fig. 30, Plate LXXI). The composition of the ternary eutectic point is in the neighbourhood of 45 : 40 : 15 at.-% Ni–Cr–Al at 1320° ± 10° C.

From Fig. 8 (p. 459) it will be seen that most of the nickel-rich alloys solidify as γ or β or as a mixture of the two phases and that the composition range within which the γ' phase occurs is quite small. When the γ' phase does appear in the structures it is usually in the form of envelopes separating the β phase from γ (Fig. 27, Plate LXXI). The α phase appears in cast alloys only if the nickel content is below about 52 at.-% (Fig. 28, Plate LXXI).

V.—THE 1150° C. ISOTHERMAL SECTION

Apart from changes in the positions of the phase boundaries, the phase relationships remain the same as the temperature falls from the solidus to 1150° C. (Fig. 9). The γ and β phase fields contract towards the nickel corner and NiAl, respectively, and the γ' phase field extends slightly, moving the ($\beta + \gamma + \gamma'$) three-phase triangle away from the nickel–aluminium edge of the composition triangle. Alloys within the ($\beta + \gamma$) phase field remain little changed in appearance, apart from the spheroidizing of the eutectic and the occurrence of small globules of α-chromium associated with the β phase (Fig. 32, Plate LXXII). The ($\gamma + \gamma'$) phase field is very narrow, the γ and γ' phase fields being separated by a gap of less than 2 at.-% aluminium at one point. Alloys containing 70–80 at.-% nickel and lying in the γ phase field between the 10 at.-% aluminium line and the $\gamma/(\gamma + \gamma')$ boundary are highly ordered, and at the boundary have practically the same lattice parameter as the γ' alloys, on the $\gamma'/(\gamma + \gamma')$ boundary. In consequence the X-ray-diffraction patterns of γ, $(\gamma + \gamma')$, and γ' alloys are almost indistinguishable from each other. It has been mentioned earlier that, in some $(\gamma + \gamma')$ alloys in which the parameters of the two phases were nearly identical after annealing at 750° C., the γ' precipitate assumed a cubic form instead of the more usual rounded shape. Similar cubic precipitates were also observed in alloys annealed at 1150° and 1000° C. (Figs. 31 and 35, Plate LXXII). This seems to confirm the idea that this type of precipitate occurs in preference to the spherical type whenever the lattice parameters are, within close limits, the same. Similar structures have been shown by Bradley[15] in alloys within the $(\beta + \beta')$ phase field of the iron–nickel–aluminium system, where two body-centred cubic phases with equal lattice parameters co-exist in equilibrium.

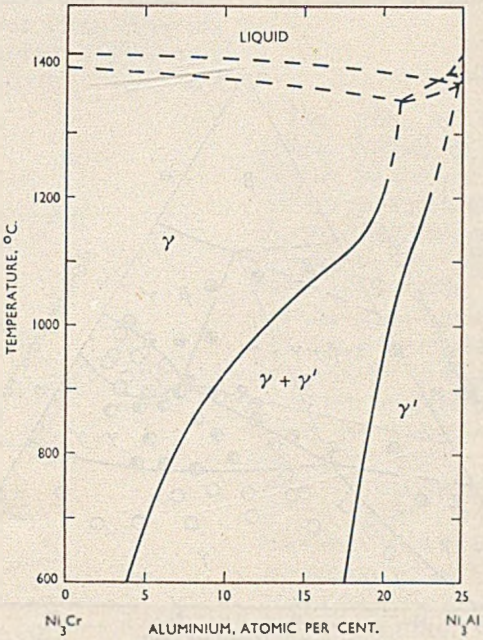

Fig. 10.—Vertical Section at 75 at.-% Nickel.

By comparison of the 1150° and 750° C. isothermals (Figs. 9 and 3) and reference to the vertical section (Fig. 10) it will be seen that alloys within a range of composition at the aluminium-rich side of the γ phase field at 1150° C. transform to γ' on cooling to 750° C. The γ' phase is ordered throughout its whole range with aluminium atoms preferentially occupying the cube-corner positions. The γ phase is ordered around the Ni_3Cr composition at low temperatures, with chromium atoms at the cube corner sites, and as the solubility of aluminium in γ increases with temperature, so the aluminium atoms tend to replace the chromium atoms and the ordering temperature rises rapidly. As a result of this an alloy in the γ phase field at 1150° C. may be in the γ' phase field at 750° C. and have precisely the same structure as at the higher temperature. At intermediate temperatures it consists of a mixture of the two phases of which γ' is always ordered while the γ is probably disordered at the lower temperatures. Although it has not been confirmed experimentally, on account of the difficulty in obtaining satisfactory parameter measurements on alloys quenched from very high temperatures, the run of the isoparametric contours determined for the γ and γ' phase fields at 750° C. suggests that the lattice parameter of such an alloy in the γ' state is higher than in the γ state.

VI.—THE 1000° C. ISOTHERMAL SECTION

The structures of alloys annealed at 1000° C. reveal a distinct change in phase relationships from those existing at higher temperatures. The phase diagram for the nickel-rich alloys is shown in Fig. 11. The γ and β phase fields are both smaller and the γ' field larger than at 1150° C. The $(\beta + \gamma')$ and $(\gamma + \gamma')$ two-phase regions persist and become considerably more extensive. Many of the alloys which at 1150° C. lay in the $(\beta + \gamma)$ and $(\beta + \gamma + \gamma')$ phase fields contain appreciable amounts of α phase at 1000° C. and γ' phase appears in alloys previously in the $(\alpha + \beta + \gamma)$ field. This evidence points to the existence of a four-phase invariant plane at or near 1000° C. in which α, β, γ, and γ' are in equilibrium together. Consideration of the phase fields at temperatures above and below 1000° C. shows that such a four-phase plane must exist. At 1150° C.

the four single-phase fields are linked by the two three-phase triangles ($\alpha + \beta + \gamma$) and ($\beta + \gamma + \gamma'$) and the two-phase region ($\beta + \gamma$) (Fig. 12 (*a*)). The two-phase fields ($\alpha + \gamma$), ($\alpha + \beta$), ($\beta + \gamma'$), and ($\gamma + \gamma'$) persist throughout the temperature range from the solidus to 750° C. and are not involved in the transition. At 750° and 850° C. the single-phase fields are linked by the two three-phase triangles ($\alpha + \beta + \gamma'$) and ($\alpha + \gamma + \gamma'$) and the two-phase field ($\alpha + \gamma'$) (Fig. 12 (*c*)). The disappearance of the ($\beta + \gamma$) phase field and formation of the ($\alpha + \gamma'$) phase field can be achieved only by an invariant four-phase transition plane on which they are represented by the diagonals (Fig. 12 (*b*)). The transition may be expressed in the form :

$$\beta + \gamma \rightleftharpoons \alpha + \gamma'$$

This reaction bears some resemblance to both a eutectoid and a peritectoid, but since all four phases persist through the four-phase plane, the only terms by which it is adequately described are " four-phase transition " and " 2/2 inversion ". As far as the authors are aware, no similar instance of such a transition involving four solid phases in an actual alloy system has been recorded in the literature.

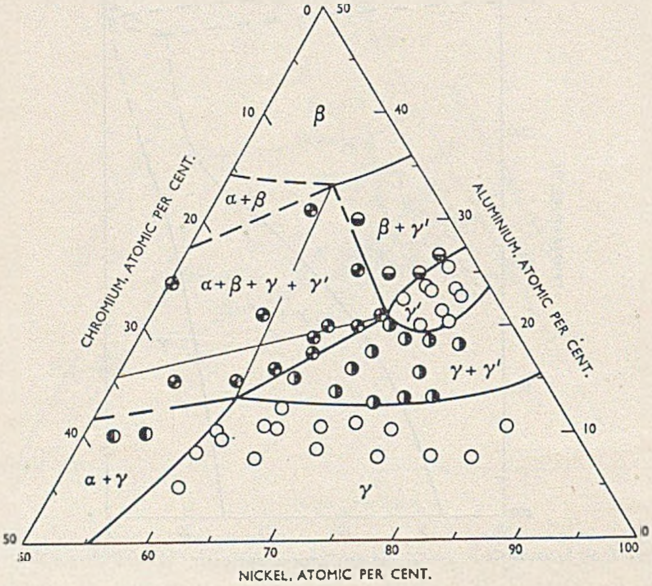

Fig. 11.—The Nickel–Chromium–Aluminium Phase Diagram : Isothermal Section for 1000° C.

The microstructures of alloys within the four-phase plane vary considerably, as would be expected from the various changes in structure taking place on cooling to 1000° C. (Figs. 36 and 37, Plate LXXII). The amount of α phase visible in the structures is small, but is in accord with the low solubility of nickel in chromium at this temperature, which places the α corner of the phase field near the chromium corner of the composition triangle. An exception among the microstructures is that of alloy No. 88, which contains very little α phase at 1000° C. and apparently no γ'. At 850° C. some β remains in the microstructure, although the X-ray diffraction pattern of a filed sample annealed at 750° C. shows only α and γ'. The reason for this may be associated with the fact that the composition of the alloy lies very near the β–γ diagonal of the four-phase plane, so that the equilibrium between the β and γ phases is more stable than in alloys with more or less chromium. This has been taken into consideration in placing the boundaries in the 1000° C. diagram.

Mention has been made in earlier sections of the presence of small amounts of α phase in the microstructures of some of the annealed alloys in which it would not be expected. It has been found that in almost every alloy examined in which the β phase crystallizes from the melt, angular globules of α phase form on annealing either in or adjacent to the β phase. These are readily recognized microscopically since they are a brilliant bluish white, whereas the β phase is light brown (Fig. 32, Plate LXXII). Even when the original β has disappeared as a result of structural changes such as the formation of γ', the α remains and cannot be accounted for by the phase diagram. Similarly, where α occurs in the cast structure, β is to be found in the same parts of the structure on annealing. In the case of an alloy such as No. 305 (Figs. 28, Plate LXXI, 33, and 38, Plate LXXII) there is no evidence to indicate that the β phase should not be present at 1150° C., but the run of the boundaries at 1000° C., based on the evidence from other alloys, places the alloy in the ($\alpha + \gamma$) field, although the β persists in the structure. These structures may well be caused by lack of homogeneity in the alloys, for, while an alloy consisting of a single phase at temperatures near the solidus can be made homogeneous provided sufficient time of annealing is allowed, an alloy with a polyphase structure at the solidus will consist of more than one phase, however long it is annealed. Hence, on annealing at lower temperatures, the phases transform independently of each other, except at their interfaces. Because of this, traces of eutectic are to be found around the primary dendrites in many of the alloys, and it is in these regions that the anomalous α–β changes occur. It may be inferred from the readiness with which the one phase forms from the other that there is a very marked reduction of solubility of β in α and of α in β as the temperature falls, particularly along the side of the ($\alpha + \beta$) region nearest the nickel–chromium edge of the composition triangle. In order to derive a phase diagram which represents the probable equilibrium relationships, small amounts of α or β in the structures of some of the alloys have had to be ignored.

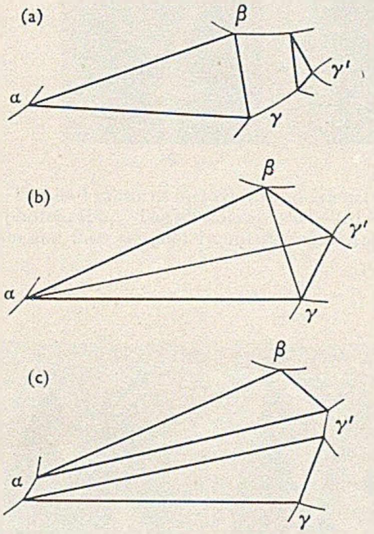

FIG. 12.—The Transition from β–γ to α–γ' Equilibrium.

Some of the alloys which lie in the ($\beta + \gamma + \gamma'$) three-phase region at temperatures above 1000° C. present a structural feature which is worth noting. The microstructure of an alloy such as No. 300 at 1150° C. (Fig. 34, Plate LXXII) shows γ' precipitated within the γ phase and envelopes of γ' surrounding the β phase, reminiscent of the γ' formed by the peritectic reaction between the liquid and β (Fig. 27, Plate LXXI). Since the alloy solidified from the melt as ($\beta + \gamma$) another explanation has to be sought. In fact, the formation of these γ' envelopes at temperatures below the solidus is due to an interaction between β and γ which might be described as a peritectoid, for, as the temperature falls, the ($\beta + \gamma + \gamma'$) triangle moves away from the composition of the last-formed γ', depositing γ' of a composition intermediate between those of β and γ. Whereas the β/γ' and γ'/γ interfaces in the structures resulting from the peritectic reaction are normally strongly dissimilar in appearance under the microscope, owing to the fact that the γ phase was liquid,[6] there is no marked difference between them in structures produced by annealing the solid alloys in which the peritectoid reaction takes place. This is attributable to the long time during which diffusion can take place and to the approximately equal rates of diffusion in the solid alloys.

VII.—SUMMARY

The part of the nickel–chromium–aluminium system defined by the composition triangle Ni–Cr–NiAl is in essence a self-contained ternary eutectic system in which the participating solid phases are the solid solution based on nickel (γ), NiAl (β), and chromium (α). The temperature of the ternary eutectic is 1320° ± 10° C. The solidification of low-chromium alloys containing 20–30 at.-% aluminium is complicated by the formation of the γ' phase, based on Ni_3Al, via a peritectic reaction.

As the temperature falls, the γ and β phase fields, which are extensive at the solidus, contract towards the nickel corner and NiAl, respectively. At the same time, the γ' phase field extends in the direction of the nickel–chromium edge of the composition triangle, chromium atoms replacing both nickel and aluminium in the Ni_3Al structure. From the solidus to a temperature in the neighbourhood of 1000° C. the γ and β phases are in equilibrium with each other, the two-phase field diminishing in extent until it disappears in a four-phase plane representing the isothermal reaction:

$$\beta + \gamma \rightleftharpoons \alpha + \gamma'$$

The two-phase ($\alpha + \gamma'$) field which is formed as a result of this reaction widens as the temperature falls.

The face-centred cubic γ and γ' phases have separate identities throughout the temperature range explored. The equilibrium between them is such that their lattice parameters in some of the ($\gamma + \gamma'$) alloys are almost identical, giving rise to microstructures in which the precipitated γ' assumes a cubic form.

Although specific-heat and resistivity measurements have shown that the γ phase is ordered below 540° C. at the Ni_3Cr composition, superlattice lines cannot be seen in Debye–Scherrer photographs of this alloy. However, they show clearly in diffraction patterns of similar alloys containing 5 at.-% aluminium after slow cooling to room temperature. Diffraction patterns of alloys quenched from 1150° C. show that ordering persists to this temperature when the aluminium content is more than 10 at.-%.

ACKNOWLEDGEMENTS

The authors wish to thank The Mond Nickel Co., Ltd., for permission to publish this paper and Mr. H. W. G. Hignett and Mr. E. H. Bucknall for their interest and helpful discussions.

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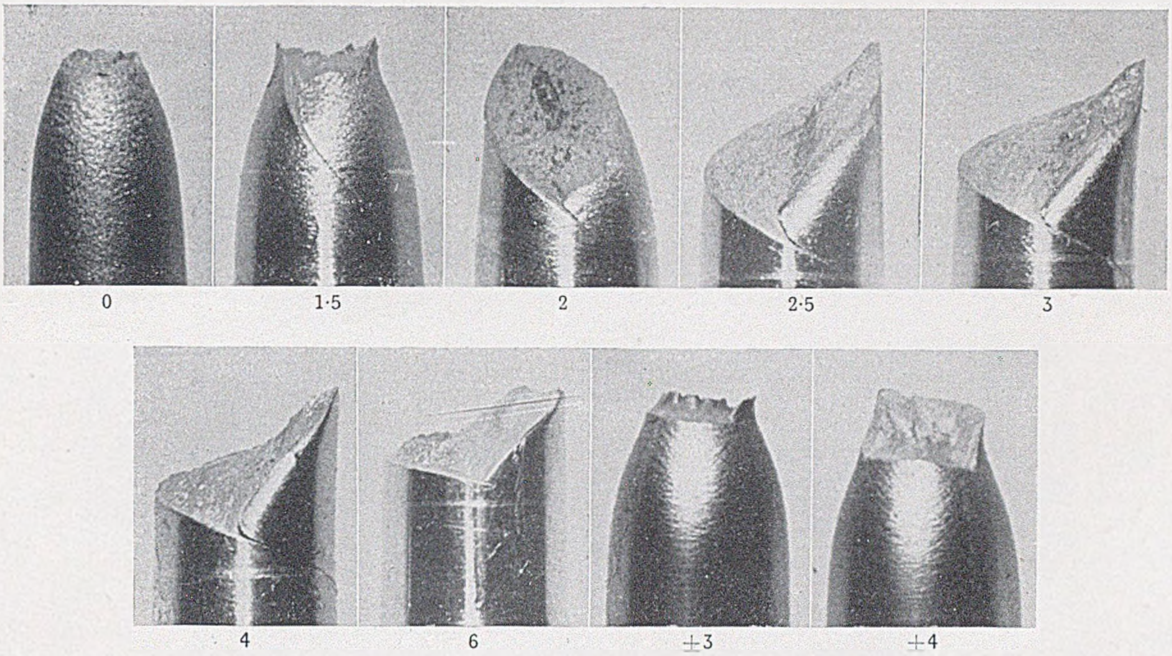


FIG. 6.—Tensile Fractures in Prestrained Nickel Specimens. The shear strain in the surface of each specimen is indicated below the photograph. The photographs labelled with numbers preceded by ± show tensile fractures after twisting to the indicated strain and then completely untwisting. Approx. × 4.

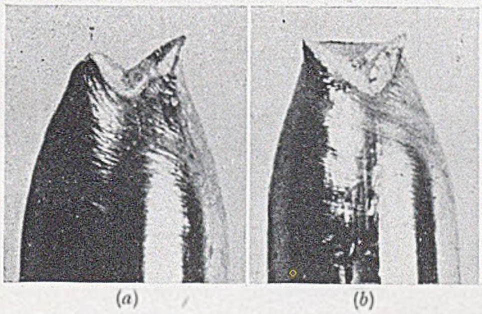


FIG. 7.—Tensile Fractures in Commercially Pure (2S) Aluminium Specimens Twisted to Shear Strains of (a) 4 and (b) 5. Approx. × 4.

FIG. 8.—The Tensile Fracture in a High-Purity Aluminium Specimen Twisted to a Shear Strain of 6. Approx. × 4.

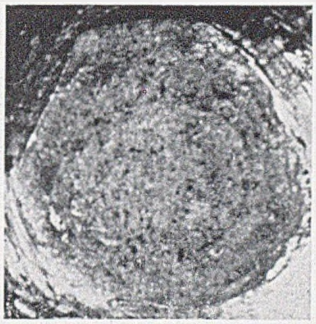


FIG. 9.—The Tensile Fracture in a Commercially Pure Aluminium Specimen Tested Without Prestraining. Approx. × 16.

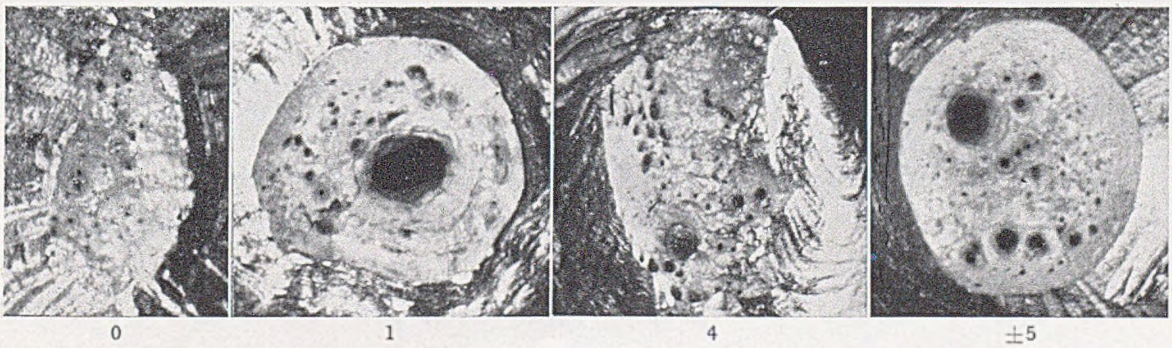


FIG. 10.—Tensile Fractures in High-Purity Aluminium. The shear strain in the surface of each specimen is indicated below the photograph. The ± sign has the meaning indicated in the caption to Fig. 6. Approx. × 16.

MAGNESIUM–ALUMINIUM ALLOYS.

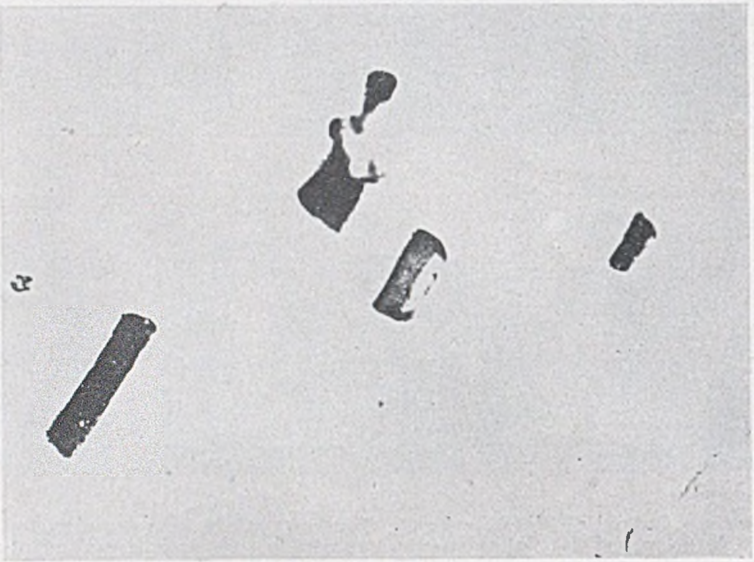


FIG. 1.—Several Cavities Within One Large Grain. × 750.

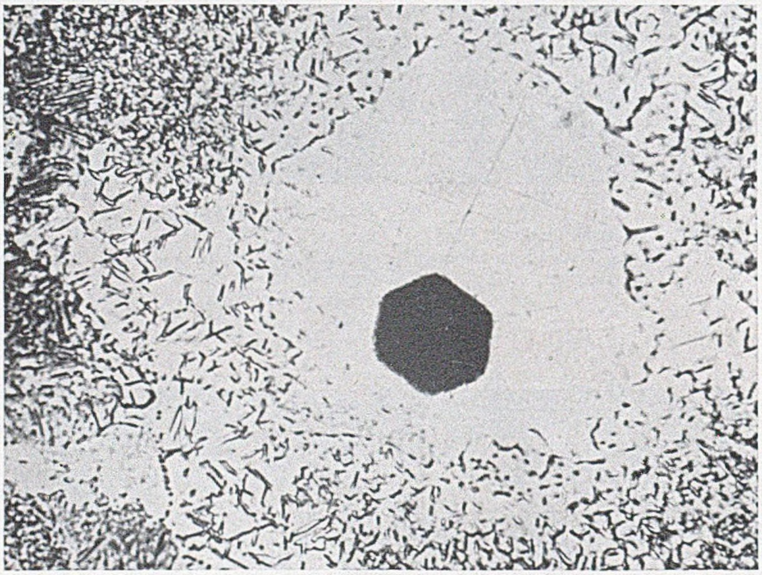


FIG. 2.—Cavity in a Grain Sectioned Nearly Parallel to the Basal Planes. × 750.

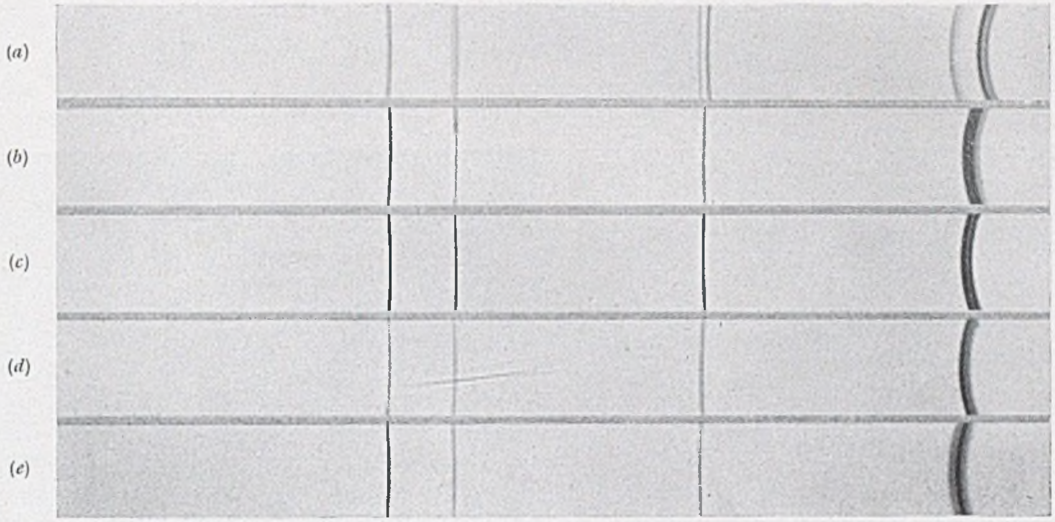
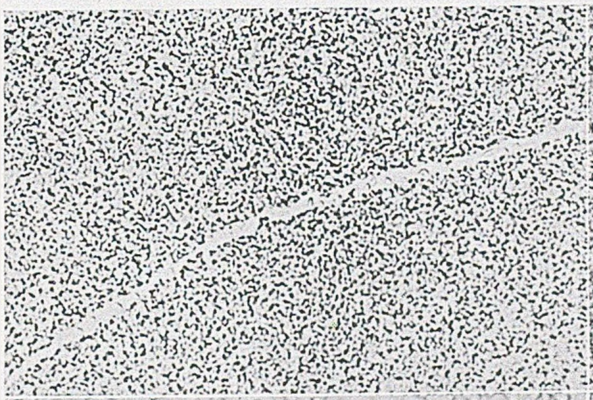


FIG. 13.—X-Ray-Diffraction Patterns of Alloys in the $(\gamma + \gamma')$ Phase Field at 750° C. (Mn $K\alpha$ Radiation).

(a) 83·9/16·1 at.-% Ni–Al.
(b) 76·7/9·9/13·4 at.-% Ni–Cr–Al.
(c) 72·1/15·0/12·9 at.-% Ni–Cr–Al.
(d) 69·7/21·7/8·6 at.-% Ni–Cr–Al.
(e) 65·0/22·6/12·4 at.-% Ni–Cr–Al.

MICROSTRUCTURES OF NICKEL–CHROMIUM–ALUMINIUM ALLOYS ANNEALED AT 850° C.

14



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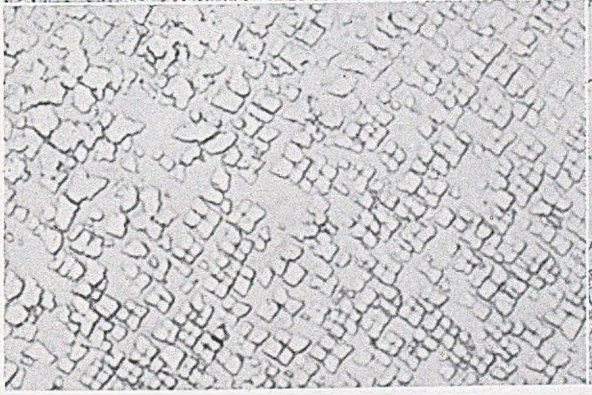

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FIG. 14.—Alloy No. 71. 74·6/9·8/15·6 at.-%. $\gamma + \gamma'$. × 500.
FIG. 15.—Alloy No. 107. 69·0/20·3/10·7 at.-%. $\gamma + \gamma'$. × 500.
FIG. 16.—Alloy No. 251. 71·7/17·3/11·0 at.-%. $\gamma + \gamma'$. × 1500.
FIG. 17.—Alloy No. 103. 76·6/5·2/18·2 at.-%. $\gamma + \gamma'$. × 500.

MICROSTRUCTURES OF NICKEL–CHROMIUM–ALUMINIUM ALLOYS ANNEALED AT 850° C.

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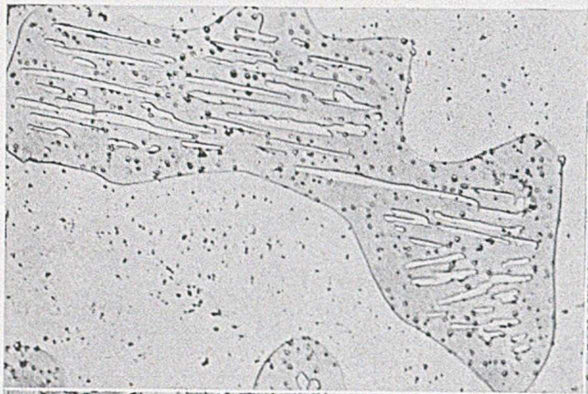

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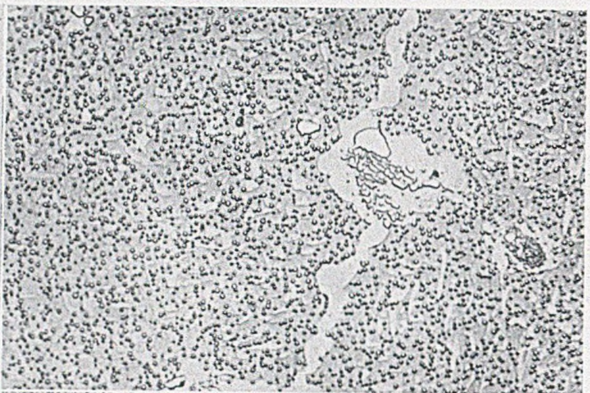

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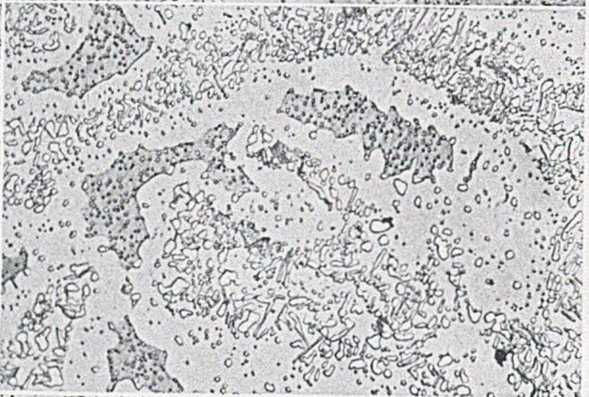

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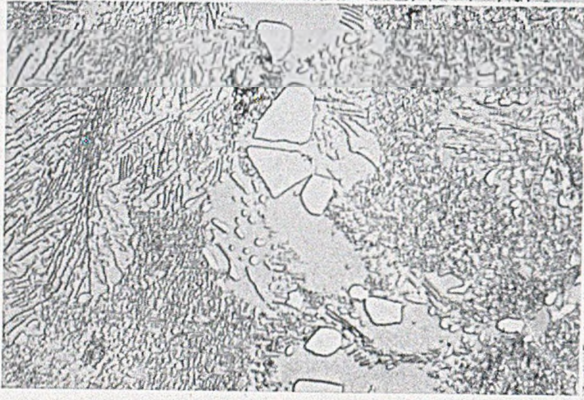

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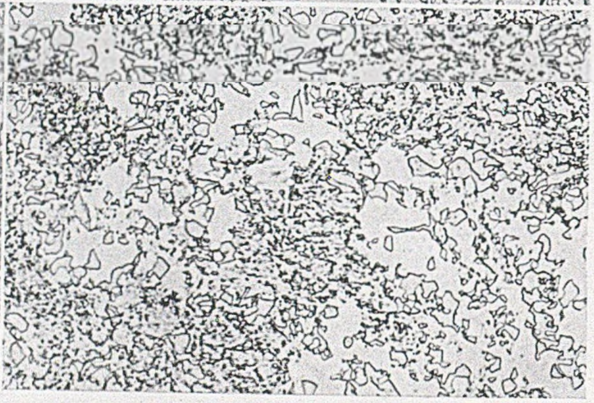
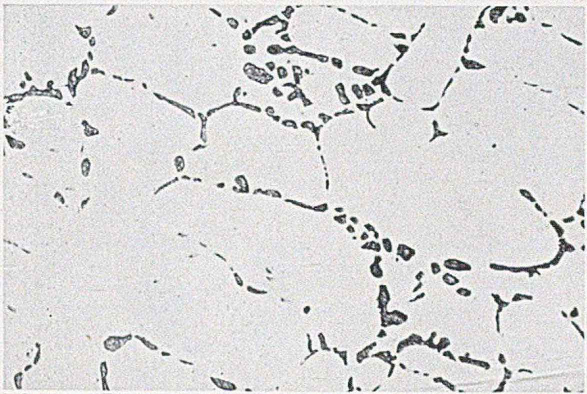


FIG. 18.—Alloy No. 16. 68·5/4·9/26·6 at.-%. $\beta + \gamma'$. × 500.
FIG. 19.—Alloy No. 114. 58·2/10·8/31·0 at.-%. $\alpha + \beta + \gamma'$. × 500.
FIG. 20.—Alloy No. 119. 50·3/25·6/24·1 at.-%. $\alpha + \beta + \gamma'$. × 1000.
FIG. 21.—Alloy No. 88. 59·1/19·7/21·2 at.-%. $\alpha + \beta + \gamma'$. × 500.
FIG. 22.—Alloy No. 115. 59·8/24·9/15·3 at.-%. $\alpha + \gamma + \gamma'$. × 500.
FIG. 23.—Alloy No. 151. 55·0/34·8/10·2 at.-%. $\alpha + \gamma + \gamma'$. × 500.

MICROSTRUCTURES OF AS-CAST NICKEL-CHROMIUM-ALUMINIUM ALLOYS.

24



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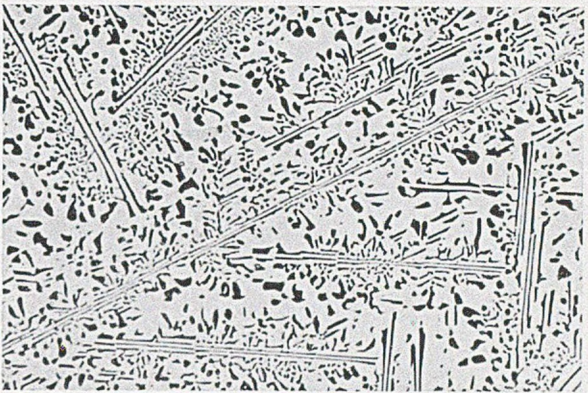

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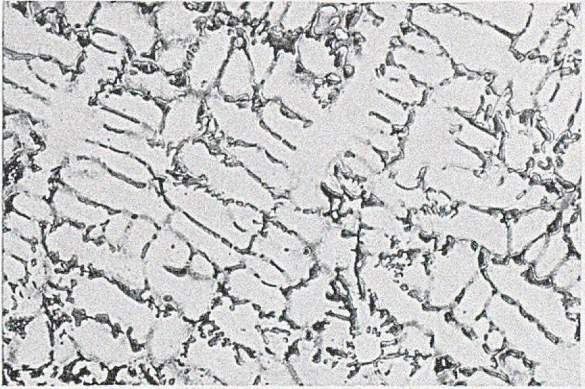

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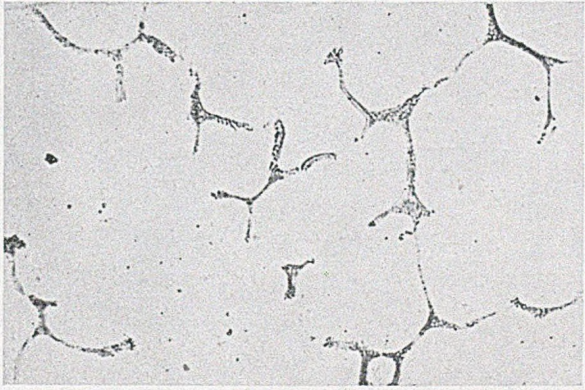

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FIG. 24.—Alloy No. 115. 59·8/24·9/15·3 at.-%. $\beta + \gamma$. × 150.

FIG. 25.—Alloy No. 300. 67·5/12·5/20 at.-%. $\beta + \gamma$. × 200.

FIG. 26.—Alloy No. 88. 59·1/19·7/21·2 at.-%. $\beta + \gamma$. × 200.

FIG. 27.—Alloy No. 295. 70/7·5/22·5 at.-%. $\beta + \gamma + \gamma'$. × 200.

FIG. 28.—Alloy No. 305. 52·5/37·5/10 at.-%. $\alpha + \gamma$. × 200.

FIG. 29.—Alloy No. 175. 25·4/54·6/20 at.-%. $\alpha + \beta$. × 200.

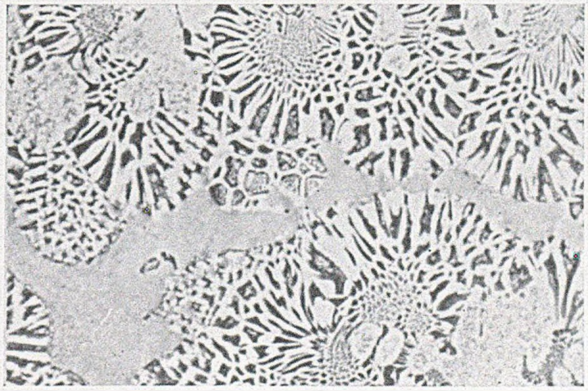
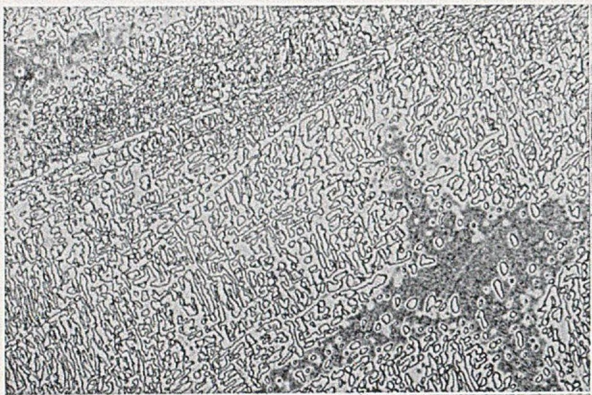


FIG. 30.—Alloy No. 223. 45/40/15 at.-%. $\alpha + \beta + \gamma$. × 200.
Unetched; phase contrast.

MICROSTRUCTURES OF NICKEL–CHROMIUM–ALUMINIUM ALLOYS.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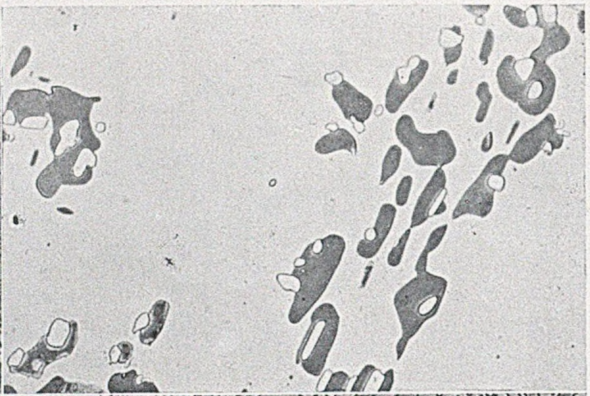
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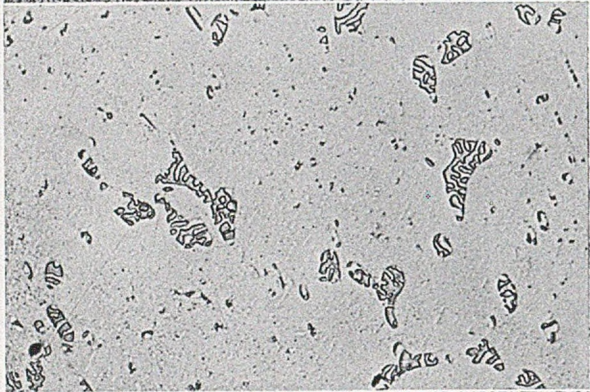
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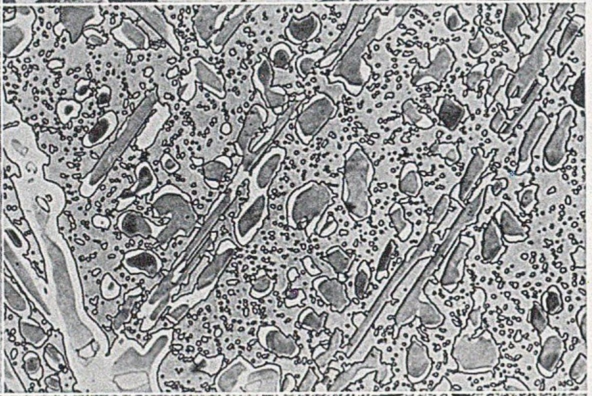
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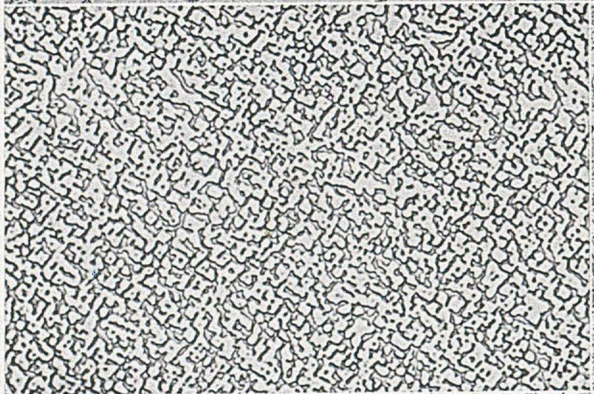
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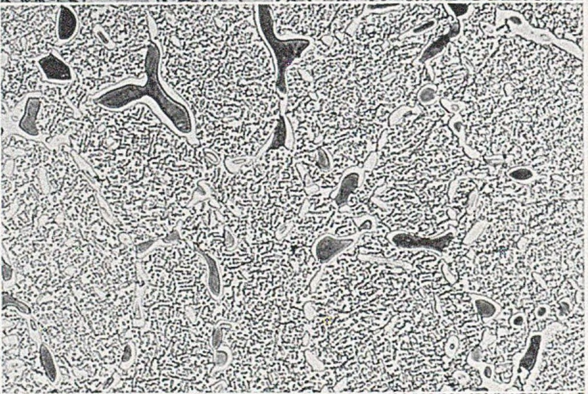
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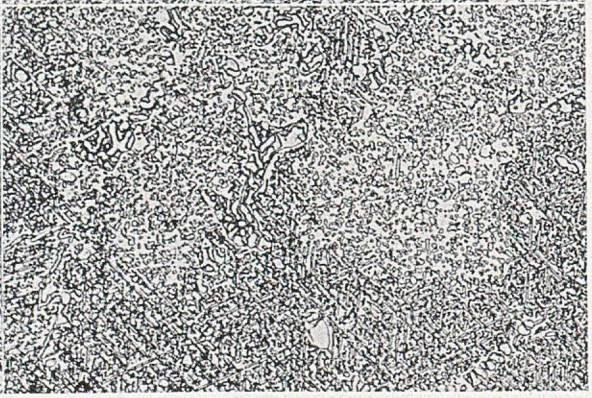
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FIGS. 31–34.—Annealed at 1150° C.

FIG. 31.—Alloy No. 14. 74·8/4·9/20·3. $\gamma + \gamma'$. × 200.
FIG. 32.—Alloy No. 109. 64·3/16·7/19·0. $\alpha + \beta + \gamma$. × 200.
FIG. 33.—Alloy No. 305. 52·5/37·5/10. $\alpha + \beta + \gamma$. × 200.
FIG. 34.—Alloy No. 300. 67·5/12·5/20. $\beta + \gamma + \gamma'$. × 200.

FIGS. 35–38.—Annealed at 1000° C.

FIG. 35.—Alloy No. 147. 69·8/12·6/17·6. $\gamma + \gamma'$. × 500.
FIG. 36.—Alloy No. 109. 64·3/16·7/19·0. $\alpha + \beta + \gamma + \gamma'$. × 200.
FIG. 37.—Alloy No. 303. 60/25/15. $\alpha + \beta + \gamma + \gamma'$. × 200.
FIG. 38.—Alloy No. 305. 52·5/37·5/10. $\alpha + \beta + \gamma$. × 200.

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 20 MAY 1953 PART 9

1—PROPERTIES OF METALS

***Creep in Metals [High-Purity Aluminium].** A. D. Schwope, F. R. Shober, and L. R. Jackson ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (2618), 52 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 427).—The report deals with creep tests on single crystals of high-purity Al, with particular ref. to the shape of the strain/time curve during the very early stages of creep. Various theories which would predict the shape of this curve have been studied. Creep tests were carried out for a few hundred hr., but only that strain which occurred during the first hr. was compared with theory. The best agreement, especially at the lower stresses, was obtained when the test curves were compared with the theoretical curve of Mott and Nabarro (*Phys. Soc.: Rep. Conf. on Strength of Solids*, 1948, 1; *M.A.*, **16**, 260), which has the form $\varepsilon = AT^{2/3}(\ln \nu t)^{2/3}$, where ε is strain, T is temp., t is time, and A and ν are const. It was found that the critical resolved shear stress for the high-purity Al used was between 20 and 30 lb./in.^{2}, and that the creep was more directly related to the resolved shear stress than to the tensile stress.

***The Specific Heat of Small Particles [of Aluminium] at Low Temperatures.** George Jura and Kenneth S. Pitzer (*J. Amer. Chem. Soc.*, 1952, **74**, (23), 6030–6032).—To illustrate the conclusions reached from development of a theory for the sp. heats of very finely divided solids at low temp., numerical values obtained for 100-Å. cubes of Al and rectangular parallelepipeds of the same vol. (i.e. 10^{-18} cm.^{3}) in the range 1°–6° K. are presented and discussed. As expected, the heat capacity of the small crystals is larger than that of the infinite crystals, and it is shown that the contribution of the gross motion decreases with increase in particle size.
—J. R.

***Nuclear Magnetic Resonance Line Width Transition in Aluminium.** E. F. W. Seymour (*Proc. Phys. Soc.*, 1953, [A], **66**, (1), 85–88).—The breadth of the magnetic resonance line in ^{27}Al has been measured up to 450° C. A progressive decrease is found, with an activation energy of 21,000 cal./mole. Although this is rather lower than the suggested activation energy of 33,000 cal./mole for self-diffusion, a correlation between the two phenomena is suggested.—E. O. H.

***Thermal Conductivities of Single-Crystal and Polycrystalline Bismuth.** Shigeka Nishioka (*Mem. Coll. Sci., Univ. Kyoto*, 1949, [A], **25**, 148–154; *C. Abs.*, 1953, **47**, 17).—[In English]. An apparatus consisting of water-jacketed heating and cooling blocks is described, and details are given for temp. and elect. power corrections. The thermal conductivity, found at 14° C. (in cal./cm./sec./°C.) was: for polycryst. Bi, 0·01914; for a single crystal ∥ the trigonal axis, 0·01269; for a single crystal ⊥ the trigonal axis, 0·02252. Linder's theory of the relation between the thermal conductivities of single- and polycryst. Bi was verified by applying the above values (L., *Phys. Rev.*, 1927, [ii], **29**, 554; see *J. Inst. Metals* (Abstracts), 1927, **37**, 414). Measurements were made also on several specimens in the form of rectangular plates of a single crystal, of which the orientation of the trigonal axis with respect to the normal to the plate ranged from 0 to 89°. The relation between the orientation and the thermal conductivity was linear, thus confirming the validity of the Voigt–Thomson law.

***The Mean Free Path of Conduction Electrons in Bismuth.** A. B. Pippard and R. G. Chambers (*Proc. Phys. Soc.*, 1952, [A], **65**, (11), 955–956).—A letter. Sondheimer (*ibid.*, (7), 561; *M.A.*, **20**, 618) has suggested theoretically that the mean free path of electrons in metals which have few conduction electrons should be abnormally long. This quantity has been measured, using skin effect, and the value is 2–4 μ at room temp., thus confirming the theory.
—E. O. H.

***A Critical Study of the Asymmetrical Temperature Gradient Thermoelectric Effect in Copper and Platinum.** N. Fuschillo (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 896–904).—A method is described by which metallic wires can be subjected to easily measurable but very pronounced asymmetrical temp. gradients without straining the wire. The e.m.f. developed by the appn. of symmetrical and asymmetrical gradients to annealed "thermopure" Pt wire and specially prepared cables of Cu wire, at both high and low temp., have been measured at many points along considerable lengths of wire. Precautions to minimize or eliminate spurious effects enabled measurements significant to 0·01 μ to be taken. The results provide no justification for the supposition that an asymmetrical temp. gradient thermoelect. effect exists, under the conditions of investigation, which cannot be explained in terms of inhomogeneity effects. The nature of these effects is discussed, and some evidence is provided which indicates that the variation in the Thomson thermoelect. potential with strain and impurity is significant.—AUTHOR.

***Creep of Copper under Deuteron Bombardment.** Warren F. Witzig (*J. Appl. Physics*, 1952, **23**, (11), 1263–1266).—Deuteron bombardments at an energy of 10 MeV. and an intensity of $\sim 10^{12}$ deuterons/cm.^{2}/sec. produced no alteration in the creep rate of polycryst. O.F.H.C. Cu at 10^4 lb./in.^{2} at 260° C., within the reproducibility limits of $\pm 12\%$. The rate of creation of large dislocation rings from vacancies is estimated to be negligible, while the extra resistance to dislocation motion by displaced atoms and the extra thermal energy associated with the latter is calculated to be negligible in terms of creep. It is remarked that Andrade's experiment on the acceleration of creep of Cd crystals (*Nature*, 1945, **156**, 113; *M.A.*, **12**, 348) cannot be taken to be of general validity for radiation effects on creep.—R. W. C.

***Distribution Coefficient of Indium in Germanium on Crystallization.** J. J. Dowd and R. L. Rouse (*Proc. Phys. Soc.*, 1953, [B], **66**, (1), 60–61).—A letter. The distribution of radioactive In in a single crystal of Ge has been measured. If diffusion in the solid phase is neglected, the concentration

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

should obey the relation $c/c_0 \cong k/(1 - g)$, where c and c_0 are the impurity concentrations in the ingot and melt, g is the fraction of melt solidified, and k is the equilibrium concentration of impurity in solid and liquid on either side of the growing interface. Agreement with experiment is fairly good. —E. O. H.

***Resistance of Germanium Contacts.** J. B. Gunn (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 908–909).—A letter. It is shown that at point contacts of Ge, very high c.d. exist, which cause non-linear mobility effects. Making certain assumptions about the variation of mobility with field, G. deduces an approx. current/voltage relationship, and the deductions are compared with other experimental investigations. Agreement is reasonably good.—E. O. H.

***Area Contacts on Germanium.** J. W. Granville and H. K. Henisch (*Proc. Phys. Soc.*, 1952, [B], **65**, (8), 650–651).—A letter. G. and H. describe experiments on Au contacts evaporated on to n-type Ge. In the reverse direction, the current for a fixed voltage is directly proportional to contact area; in the forward direction it is not. Variations of peak back voltage and power with contact area are also given. A comparison is made with formed and unformed W point contacts, and the results are briefly discussed. —E. O. H.

***Contact Properties of *p*-Type Germanium.** J. W. Granville, H. K. Henisch, and P. M. Tipple (*Proc. Phys. Soc.*, 1952, [B], **65**, (8), 651–652).—A letter. Graphs are shown of current/voltage curves for p- and n-type Ge, with either etched or polished surfaces, using unformed W contacts, and also p-type Ge contacts. G., H., and T. conclude that the interpretation requires assumptions which differ from those applied to n-type material.—E. O. H.

***Transparent Conducting [Gold] Films.** E. J. Gillham and J. S. Preston (*Proc. Phys. Soc.*, 1952, [B], **65**, (8), 649).—A letter. Experiments are described on Au films. The resistance of the films, 60 Å. thick, is much higher than that of the bulk material, and the transmitted light is bluish, implying a high absorption. The properties are influenced by the process of cleaning the glass. Oxide substrates also influence the resistance and opt. absorption. The effect of heating the films is described, and briefly discussed theoretically.—E. O. H.

***Measurements on Thin Iron Films.** A. van Itterbeek, L. de Greve, and F. Heremans (*Appl. Sci. Research*, 1951, [B], **2**, (4), 320–324).—[In English]. Curves are determined for the sp. resistance of Fe as a function of thickness of the film. The shape of the curves obtained for the sputtered films differs completely from that of the condensed ones. This difference is attributed to the more amorphous nature of the structure of sputtered as compared with condensed films. The structure of the films was investigated by means of electron microscopy and electron diffraction.—S. Ma.

***Influence of a Magnetic Field on the Electrical Resistance of Iron Films.** B. Franken, A. van Itterbeek, G. J. van den Berg, and D. A. Lockhorst (*Physica*, 1952, **18**, (10), 771–779).—[In English]. A plot of the change of resistance in Fe films against magnetic field shows max. for thin films, and min. for thick films. It is considered that the max. and min. occur at field strengths equal to the coercive force. Studies were made at various temp., from 1·3° K. upwards, and it is considered that the changes observed are in reasonable agreement with theory.—E. O. H.

***The Influence of Thallium on the Creep of Lead.** (Gifkins). See col. 624.

Properties of Molybdenum. Robert M. Parke (*Metal Progress*, 1951, **60**, (1), 80-B).—A data sheet.—S. R. W.

Properties of Electrodeposited Nickel. —— (*Metal Ind.*, 1953, **82**, (3), 47–48).—Abridged from *Plating*, 1952, **39**, 865; *M.A.*, **20**, 324.—J. H. W.

***The Temperature Variation of the Magnetization [of Nickel] in Low and Moderate Fields.** R. S. Tebble, J. E. Wood, and J. J. Florentin (*Proc. Phys. Soc.*, 1952, [B], **65**, (11), 858–871).—Experiments are described on Ni rods and ellipsoids, in which the changes in magnetization accompanying change in temp. are measured. At high fields (>50 Oe.), the effect is mainly reversible, and can be correlated directly with the "inverse" magneto-caloric effect. At low fields, however, not all the effect is reversible, and a sepn. of reversible and irreversible phenomena can be made. The reversible part is discussed in terms of the theory of Stoner and Rhodes (*Phil. Mag.*, 1949, [vii], **40**, 481; *M.A.*, **17**, 9), and leads to the conclusion that reversible rotational processes occur. —E. O. H.

***Changes in the Wetting of Platinum Under the Influence of Various Gases.** I. N. Plaksin and S. I. Vladimirov (*Doklady Akad. Nauk S.S.S.R.*, 1951, **78**, (5), 933–936; *C. Abs.*, 1953, **47**, 23).—[In Russian]. The action of certain gases on the surface of Pt was studied for the purpose of establishing a characteristic flotation-active surface. The method used was that of measurement of the contact angle at the three-phase boundary in the system water–air–metal. O, N, CO_2, A, H, and air were the gases studied. Interaction of the gas with the Pt surface was continued for 3 hr., the contact angle being measured every 15 min. In the case of O, a considerable increase of the hydrophobicity was noted. N, A, and CO_2 gave identical wettability curves. The activity of O dissolved in water was also tested. The aq. medium contg. dissolved O had a greater effect on the Pt surface than did gaseous O. The effect of H on Pt surfaces was found to be negligible.

***Breakdown in Selenium Rectifiers.** R. Cooper (*Proc. Phys. Soc.*, 1952, [B], **65**, (6), 409–414).—Experiments were carried out to determine how the breakdown voltage varied with temp., counter-electrode metal, and the thermal dissipation const. of the rectifier. Breakdown voltage decreased with decreasing temp., and was independent of the counter-electrode at lower temp., but some dependence was observed at room temp. Reducing the thermal dissipation caused breakdown to occur at a higher voltage. The results are discussed on the basis of image force theory and thermal instability theory, but there is no agreement with either. —E. O. H.

***Temperature Variation of the Rectification Characteristics at Selenium Contacts.** E. W. Saker (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 990–991).—A letter. A comparison is made between two theories of barrier rectification due to Mott and to Landsberg, and experimental results on reverse current–voltage characteristics are reported. Both theories gave reasonably good fit at room temp. and above, and it is concluded that no choice can be made between the two models. Some anomalies appear below 0° C., and their implications are briefly discussed.—E. O. H.

***Positive-Current Creep in Selenium Rectifiers.** R. Cooper and J. Harrington (*Proc. Phys. Soc.*, 1952, [B], **65**, (4), 303–304).—A letter. When a const. reverse voltage is applied to a Se rectifier, the value of the current varies with time. The experiments described here indicate that this positive-current creep is not, as has been suggested, due to power dissipation in the barrier layer.—E. O. H.

***The Influence of Mercury Vapour on Selenium Rectifiers and Selenium Photoelements.** P. Selényi (*Proc. Phys. Soc.*, 1952, [B], **65**, (7), 552; corrigendum, (9), 742).—A letter. The blocking resistance of Se rectifiers is known to be reduced by Hg vapour. Se photoelements are also affected. S. suggests that this is the result of the formation of mercuric selenide, a semi-conductor of high conductivity. The effects on the blocking current/voltage curve, and the forward current/voltage curve are described, and the theoretical consequences are briefly discussed.—E. O. H.

***Silver Wire: Its [Property Changes on] Cold Working and Annealing.** M. Balicki (*Wire and Wire Products*, 1952, **27**, (12), 1291–1296, 1313, 1315–1316).—Contains a typical picture of property changes that occur in the work-hardening/annealing cycle in wire-making. The wire tested was "pure" Ag contg. Cu 0·001, Te 0·0001%, and traces of Pb, Si, Ca, Mg, and Mn. The results of cold-drawing and annealing are analysed math. Curves of the change of activation energy and the frequency coeff. of primary recrystn. due to the degree of cold-working are given, and an equation describing the course of changes encountered on annealing of work-hardened wires is obtained. This is applicable to metals other than Ag, and is of practical value as a time-saver. It is now possible to calculate how long a time a given material in its work-hardened state is capable of performing at a given temp. before it loses 5 or 10% of its stress-carrying capacity. 6 ref.—C. P. F.

***Adsorption of Oxygen on Silver.** F. H. Buttner, E. R. Funk, and H. Udin (*J. Phys. Colloid Chem.*, 1952, **56**, (5), 657–660).—Measurements were made of the surface tension of solid Ag in He–O mixtures between 870° and 945° C. Using the Gibbs adsorption isotherm with these data indicates that O chemisorbs on Ag and that at 932° C. there are ~1·4 atoms of O adsorbed/Ag atom at the surface.—K. C. R.

***The Effect of Cyclotron Bombardment on Self-Diffusion in Silver.** R. D. Johnson and A. B. Martin (*J. Appl. Physics*, 1952, **23**, (11), 1245–1254).—Tracer techniques were used to measure D for self-diffusion of Ag both with and without proton bombardment during the anneal. The protons were of 10-MeV. energy and of 0·5–5 μ amp./cm.^{2} beam intensity. No significant variation of D due to irradiation was noted at any temp. Least-square values for D_0 and Q are $0{\cdot}11 \pm 0{\cdot}05$ cm.^{2}/sec. and $40{,}800 \pm 700$ cal./mole, resp. The ineffectiveness of the irradiation would be expected theoretically.
—R. W. C.

***The Cellular Method of Determining Electronic Wave-Functions and Eigenvalues in Crystals, with Applications to Sodium.** D. J. Howarth and H. Jones (*Proc. Phys. Soc.*, 1952, [A], **65**, (5), 355–368).—The wave-functions of electrons are considered at certain specified points in the Brillouin zone. By expanding the wave equation in spherical harmonics eigenvalues can be found for the electrons whose wave vectors lie at the ends of 2-, 3-, and 4-fold axes. In particular, the energy gap at the centre of a zone face can be calculated, and is found to be 0·65 eV. for Na.—E. O. H.

***Experiments on the Magnetic Transition from the Supraconducting State to the Normal State [in Tin].** M. P. Garfunkel and B. Serin ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 43–44).—Transition of the bulk of a metal, of finite length, from the supraconducting to the normal state can take place by growth from an already established nucleus at the end of the sample, due to non-zero demagnetization of the ends. An experiment carried out to study the transition under conditions such that the transition occurs in single crystals of Sn without the creation of an intermediate state, is described. No indication was obtained of the passage of the central section of the crystals into the normal state when the magnetic field over that part exceeded the critical field.
—J. S. G. T.

***Kinetics of the Supraconducting Phase Transition [in Tin].** T. E. Faber ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 47–50).—Two macroscopic aspects of the process whereby a rod of Sn in a longitudinal magnetic field becomes supraconducting are discussed, viz. (1) what governs the field strength at which the growth of one phase at the expense of the other is just able to start, and (2) in what direction and how fast the growth then proceeds. The theory is supported by experimental results obtained with rods of Sn ~ 20 cm. long and 1–4 mm. in dia. The transition is set off at one end of a rod that alone is supercooled and the voltage pulses produced in search coils are recorded photographically. The supraconducting phase propagates most readily along the surface and tends to form a supraconducting sheath that locks in the flux in the interior. The thickness, d, and velocity, v, of the supraconducting layer are shown to be given by $d = \Delta[(H_c - H)/H_c]^{-1}$ and $v = 8\pi C[(H_c - H)/H_c]^2/\sigma\Delta$, where σ is the elect. conductivity of the normal metal and C is a dimensionless const. The theory, therefore, predicts the observed variation of d and v with $(H_c - H)$, H_c denoting the critical field strength and H the applied strength of field. It is suggested that the theory requires amendment, possibly to take account of mean free path effects.—J. S. G. T.

***Studies on Oscillating Supraconducting [Tin] Spheres.** R. H. Pry, A. L. Lathrop, and W. V. Houston ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 81–84).—Torsional oscillations of a Sn sphere at liq.-He temp. were studied in horizontal magnetic fields of strengths 10^{-4}–100 Oe. As the conductivity or angular frequency of the sphere was increased, the restoring torque increased steadily to a max., but the damping torque increased to a max. and then decreased to zero. The experimental results, including those obtained with a vertical magnetic field, are in accordance with London's theory. In view of Bardeen's recent articles on supraconductivity, suggesting an effective superelectron mass of 10^{-4} of the free-electron mass, the mass of the electron in a supraconductor, as here measured, is of the same order of magnitude as the free-electron mass.—J. S. G. T.

Hardness Conversions for Titanium and the Relation Between Hardness and Tensile Strength. J. B. Sutton, E. A. Gee, and W. B. DeLong (*Metal Progress*, 1950, **58**, (5), 720-B).—A data sheet.—S. R. W.

Titanium. F. R. Morral (*Nucleo* (*Madrid*), 1952, **7**, 369–375; *C. Abs.*, 1953, **47**, 86).—The past, present, and future of Ti in the U.S. is considered from the point of mining, extraction, casting, and fabrication. Properties of Ti and its alloys and their uses are given. In a table are listed the characteristics of binary Ti alloys so far studied; these are considered in the light of the Hume-Rothery generalizations.

***Dependence of the Compressibility of Cast Zinc Upon the Angle Between the Axis of Compression and the Principal Direction of Crystal Growth.** Rolf-Bodo Denker, Gunter Falkenhagen, and Wilhelm Hofmann (*Abhandl. Braunschweig. wiss. Ges.*, 1952, **4**, 87–93).—The compressibility of cast Zn plates was measured at 20°–400° C. by the drop-hammer method. The resistance to deformation measured ⊥ the plane of the plates has a value many times that measured ‖ or inclined at 45° to the plane of the plates, and this is attributed to crystal translation and twinning in the cast crystal texture of the plates.—J. S. G. T.

***Creep of Zinc Single Crystals Under Direct Shear.** Elliot H. Weinberg (*J. Appl. Physics*, 1952, **23**, (11), 1277).—A letter. Creep of single crystals of pure Zn under pure shear ‖ basal plane always followed the time law: creep strain $s = at^m$, where $m \sim 0{\cdot}6$, as long as no lattice bending occurred. Rapid recovery on annealing took place, with the same proviso. A marked Bauschinger effect for creep rates was observed.
—R. W. C.

***Creep of Zinc Single Crystals.** Donald O. Thompson (*J. Appl. Physics*, 1952, **23**, (11), 1277–1278).—A letter. Creep in tension of pure Zn crystals obeys an equation of state: $s = at^m$, where s is strain and t is time. $m \simeq 0{\cdot}5$ at room temp., reproducibly. The index m depends on temp. exponentially, with an activation energy ~3400 cal./mole. Values of m in the range 0·2–3·3 have been observed; $m = 1$ at ~60° C. The critical stress for creep is temp.-independent.
—R. W. C.

***Some Properties of High-Purity Zirconium and Dilute Alloys with Oxygen.** R. M. Treco (*Amer. Soc. Metals Preprint*, 1952, (**38**), 19 pp.).—An investigation was made of the mech. and phys. properties of high-purity Zr and of a series of Zr–O alloys. The rolled Zr bars used contained <0·2% impurities, the O content being ~0·025%. The Zr–O alloys were prepared by heating Zr specimens in O at 1055° or 1280° C., followed by heating *in vacuo* for a time sufficient to produce a homogeneous alloy. The following changes in properties were observed as the O content increased from 0·025 to 0·415%. The elect. resistivity of Zr increases linearly with increasing O content, from ~40 to 56·74 ohm-cm.; the thermal conductivity decreases from 0·400 to 0·0282 cal./°C./cm./sec. The Rockwell A hardness increases from ~20 to 59 as the O content increases. At the low-O end of the range the tensile properties are 0·1% P.S. ~11,000 lb./in.^{2}, U.T.S. ~32,000 lb./in.^{2}, elongation (on 1 in.) ~30%; corresponding figures for the other end of the O content range are 0·1% P.S. 45,400 lb./in.^{2}, U.T.S. 73,100 lb./in.^{2}, elongation 2·6%. The microstructure of Zr is not affected by up to 0·4% O. Alloys contg. up to 0·25% O can be cold-rolled without undue difficulty. The recrystn. of Zr is unaffected by the presence of O. The diffusion const. for the diffusion of O in Zr is calculated to be at least $5{\cdot}25 \times 10^{-6}$ cm.^{2}/sec. Tensile tests made at temp. up to 400° C. showed that the strength of high-purity Zr at elevated temp. is much less and that the ductility is higher than hitherto supposed. High-purity Zr has a U.T.S. of 13,000 lb./in.^{2} at 400° C. and an elongation of $\sim$90%. The presence of O increases the elevated temp. strength, but the influence of this element is less than at room temp. 14 ref.—R. W. R.

The Variability of Metal. C. O. Lloyd (*Pract. Eng.*, 1952, **26**, (652), 264–265).—A general account of certain properties of metals.—E. J.

Physical Properties of Some Liquid Metals. R. R. Miller (*Metal Progress*, 1951, **59**, (1), 80-B).—A data sheet.—S. R. W.

***On the Temperature Dependence of Poisson's Ratio for Metals.** F. Vitovec (*Österr. Ing.-Arch.*, 1952, **6**, (2), 132–134; *Appl. Mechanics Rev.*, 1952, **5**, 431).—From the temp. dependence of Young's modulus, which has been found experimentally, and the temp. dependence of the compressibility modulus, calculated from the lattice energy, it is found that Poisson's ratio increases with temp. For Al, for example, Poisson's ratio increases from 0·333 at 0° K. to 0·400 at the m.p.

***Scabbing of Metals Under Explosive Attack: Multiple Scabbing.** John S. Rinehart (*J. Appl. Physics*, 1952, **23**, (11), 1229–1233).—A relation has been discovered, and explained by elastic theory, between the number of transverse fractures occurring in a steel plate as an elastic stress wave passes through it, and the "critical normal fracture stress".
—R. W. C.

Creep of Non-Ferrous Metals. K. Löhberg (*Metal Ind.*, 1953, **82**, (1), 3–6).—Read before the Fourth International Mechanical Engineering Congress, Stockholm, 1952. See *M.A.*, **20**, 233.—J. H. W.

Factors Bearing on the Creep of Metals and Alloys. J. J. de Jong (*Ingenieur*, 1952, **64**, (28), 0.92–0.101; *Appl. Mechanics Rev.*, 1952, **5**, 514).—[In Dutch]. A critical review is given of the effect of internal and external structural factors on the creep of metals. Some points, such as effect of grain boundaries and grain-boundary segregation, structural changes of phases (the similar effect of carbide transformations on creep and on temper-brittleness is especially emphasized), formation of phases from supersaturated soln., superficial and internal oxidation, and diffusion of other gases are dealt with in more detail. The importance, as regards high creep-resistance, of lattice deformations of high stability as arising from age-hardening, internal oxidation, and mech. deformation is particularly stressed.

Fatigue and Fracture of Metals. (Murray). See col. 667.

***Fatigue of Metals.** M. Roš (*Z. angew. Math. u. Mech.*, 1952, **32**, (4/5), 130–145; *Appl. Mechanics Rev.*, 1952, **5**, 469).—R. reports the results of a careful study of the fatigue of metals, with notched and unnotched samples, and under homogeneous and inhomogeneous, uniaxial and non-uniaxial stress. Alternating stress produces anelastic and plastic disarrangements of the lattice, which increase with the number of stress reversals and the value of the stress amplitude. These disarrangements locally decrease the lattice cohesion, loosen the structure, and finally produce fissures. Thus, the various phases of rupture by fatigue are: (*a*) formation of first fissure; (*b*) growth of fissure without appreciable deformation; (*c*) forced rupture of reduced and weakened cross-section. Exhaustion by fatigue does not differ fundamentally from that due to repeated static stressing; however, each new stressing finds the conditions of the sample altered in consequence of local structural changes. Resistance to fatigue is not affected by time rate of stressing unless the behaviour of a sample is influenced by a work-hardened or otherwise treated surface layer. The Coulomb–Mohr hypothesis has been found generally valid.

***The Processes [Which Occur] in Metals Subjected to Tensile and Fatigue Stresses. III.—The Practical Relationship Between the Fatigue Stress and the Stress Concentration in Smooth and Notched Specimens.** Cord Petersen (*Z. Metallkunde*, 1951, **42**, (6), 161–170).—Cf. Thum and P., *ibid.*, 1942, **34**, 39; *M.A.*, **12**, 275. Heywood's formula (*Aircraft Eng.*, 1947, **19**, 81; *M.A.*, **15**, 143) for the relationship between the fatigue stress and the stress concentration in smooth and notched test-pieces is discussed. It is shown that the expression can be simplified somewhat, and that this modified form can be derived, also, from consideration of Morkovin and Moore's "equivalent notch" theory (*Proc. Amer. Soc. Test. Mat.*, 1944, **44**, 137). The practical implications of the results are discussed, especially in relation to C steels, for which there is excellent agreement between theory and experiment. 23 ref.—E. N.

A Survey of "Statistical Effects" in the Field of Material Failure. Waloddi Weibull (*Appl. Mechanics Rev.*, 1952, **5**, (11), 449–451).—A short review, with 63 ref.—N. B. V.

***The Influence of the Coefficient of Work-Hardening of Metals Upon the Profile of the Mound Surrounding the Impression Obtained in Hardness Testing.** M. V. Yakutovich, B. A. Vandyshev, and E. E. Surikova (*Zavod. Lab.*, 1948, **14**, (3), 338–343).—[In Russian]. Tensile tests on samples of Al, Duralumin, Cu, Armco Fe, and two steels were carried out to determine the Y.P. and the coeff. of strain-hardening tan α = slope of rectilinear portion of curve relating $s = P/A_y$ (P = load, A_y = area at yield) and ψ, the % reduction of area. Finely ground faces of specimens were impressed by a WC indenter of 90° apical angle under a previously determined load to give a 2-mm.-dia. impression in all cases. The profile of the mound was obtained by measurement of the height at definite points by a contact method. The mound vol. computed from the profiles so obtained agreed with the vol. of the impression to within 3–5%. The Y.P. (σ_s) calculated by Tammann and Müller's (*Z. Metallkunde*, 1936, **28**, 49; *M.A.*, **3**, 268) formula $\sigma_s = L/\pi a_0^2$ (where L = load (kg.) and a_0 = outside radius of the deformed area) differ from the observed ones by 4–5%, except for the steels, where 40–60% deviation was observed; this discrepancy can be eliminated by taking the value of $a_0 = a$ at $h = 5\%\ h_{\text{max.}}$, where $h_{\text{max.}}$ in the max. height of the mound. The relation between $h_{\text{max.}}$ and tan α is linear to within ±8%.—D. M. P.

Recent Views on Bearing Properties. Børge Lunn (*Metal Ind.*, 1953, **82**, (4), 63–65).—L. gives a review of some experiments in boundary lubrication. Recent work indicates that boundary lubrication is the primary and fluid lubrication the useful secondary effect. The general hypothesis is that the function of a sleeve bearing is conditioned by the formation of adherent films that prevent metallic contact between the rubbing surfaces. This hypothesis is applied to the action of bearing metals and to lubricants. A full account of the experiments is given in *Trans. Dan. Acad. Tech. Sci.*, 1952, (2).—J. H. W.

***Radioactive Tracers Reveal Friction and Wear of Metals.** J. T. Burwell and C. O. Strang (*Metal Progress*, 1951, **60**, (3), 69–74).—By making one of any two rubbing surfaces radioactive, the metal transfer or "microwelding" on rubbing under controlled conditions can be measured. Results are given of a series of tests to determine the effect of lubricant, surface hardness, and material compn. on the transfer of metal using Cr and hardened steel, Nitralloy, and steels. Examples of appn. of this radiographic technique are described briefly.—S. R. W.

***On the Change of the Internal Friction Due to the Fatigue of Metal.** Hideo Miki (*Bull. Eng. Research Inst., Kyoto Univ.*, 1952, **1**, (March), 15–19).—[In Japanese, with tables and graphs in English]. It is pointed out that the log decrement due to the internal friction of a metal is influenced not only by the internal state of the material but also by the frequency used for the measurement. Study of the log decrement measured over a wide range of frequencies is necessary to determine the internal state of the metal. During rotating–bending fatigue tests, the log decrements of 0·1% C steel (cold-drawn wire rods, 3·55 mm. dia., in the drawn and in the annealed states) have been measured by the damping method in a frequency range of 56–2250 c./s., and the mechanism of the fatigue fracture has been studied from the change of the internal friction. M. concludes that the intercryst. strain gradient induced within the metal as a result of repeated stresses plays an important role and the imperfection of the structure plays a secondary role in fatigue failure.—AUTHOR.

***Diffusion in Solid Metals.** W. Seith and A. Kottmann (*Angew. Chem.*, 1952, **64**, (14), 379–391).—A study is made of the relationship between the partial diffusion coeff. as determined by radioactive methods and the chem. partial diffusion coeff. as determined by the movement of markers in the test-pieces. Changes of shape and dimensions, and the appearance of voids through diffusion, were studied in the systems Ni–Cu, Ag–Au, Ni–Au, Ag–Pd, and Fe–Ni. On the basis of experimental results the problem of the diffusion phenomena in solid soln. is discussed. 21 ref. —W. F. H.

***Development of Porosity During Diffusion in Substitutional Solid Solutions.** R. W. Balluffi and B. H. Alexander (*J. Appl. Physics*, 1952, **23**, (11), 1237–1244).—Porosity associated with diffusion was detected metallographically after three distinct processes: (1) dezincification of α-brass by heating, and similar removal of Cu from Cu–Ni and Ag from Ag–Pd or Ag–Au. (2) Introduction of Ag into Au from the vapour phase; Ni exposed to Cu vapour and Cu exposed to Zn vapour did not give porosity. (3) Diffusion in bimetal sandwiches, of Cu/brass, Ag/Au, and Cu/Ni. The porosity developed in the brass, Ag, and Cu, resp. The total porosity developed in brass by methods (1) and (3) was compared with the max. possible (that expected if all the lost Zn were replaced by pores). The actual porosity was always a large fraction of the possible (e.g. 40% at 10% wt. loss). For a given wt. loss the porosity increased slightly as temp. of diffusion decreased. Except for the development of porosity in Ag–Au by process (2), these observations are all in consonance with previous observations in this field. The large amount of porosity is emphasized, and the consequent inaccuracy in measured D values is pointed out; it seems impossible to apply a correction for this.—R. W. C.

***Sorption of Gases on Metal Powders and Subsequent Change in Metal Reactivity at Room Temperature.** Hung Li Wang and Norman Hackerman (*J. Phys. Colloid Chem.*, 1952, **56**, (6), 771–774).—The amounts of Cl, NO_2, O, and CO taken up by powders of stainless steel, Ni, and Mo were measured at room temp. Only CO was truly adsorbed on stainless steel and Ni powders, forming a chemisorbed monolayer. Reactivities of metal powders in terms of H evolution in dil. acid soln. in the absence of air were determined for reduced stainless steel, Ni, and Cr powders when they were bare, with CO adsorbed or when covered with sorbed O.—K. C. R.

***Determination of the Surface Energy of a Metal by Molecular Orbitals.** G. R. Baldock (*Proc. Phys. Soc.*, 1953, [A], **66**, (1), 2–12).—Calculations of the electronic binding energy are used to determine the surface energy of a simple cubic lattice of monovalent atoms. Finally, the energies of {100} and {110} faces are determined for the alkali metals, but a comparison with experimental results from Na and K reveals that the calculated values are too low by at least a factor of 2. The reasons for this discrepancy are discussed. —E. O. H.

***The Polarization of Acoustic Waves in Cubic Crystals.** A. E. Fein and Charles S. Smith (*J. Appl. Physics*, 1952, **23**, (11), 1212–1213).—The spontaneous polarization of acoustic waves due to the elastic anisotropy has been calculated for Ni as a function of the direction of propagation. The max. polarization angle (i.e. deviation of the angle between direction of propagation and direction of displacement from that which would obtain in an isotropic medium) is 11½°. Cu, Ag, Au, and Fe should be similar in magnitude; Al (more isotropic) should have a smaller max. polarization angle. The polarization should be greater for β-brass and the alkali metals.—R. W. C.

***The Velocity of Second Sound in Various Media.** R. B. Dingle (*Proc. Phys. Soc.*, 1952, [A], **65**, (12), 1044–1050).—The transmission of second sound takes place in He II primarily by fluctuations in temp. and entropy. D. applies the appropriate formula to discuss the possibility of second sound propagation in other media, also taking into account the possibility of attenuation. Second sound is then not possible in ideal gases, and in non-ideal gases attenuation is large. Propagation should be possible in insulators at very low temp., and in metals the velocity would be $c_2 = \pi kT/p_0$, where p_0 is the momentum at the Fermi surface. A discontinuous jump in c_2 by a factor of $\sqrt{3}$ is predicted at the critical temp. in supraconductors. Finally, propagation in ferromagnetics is considered.—E. O. H.

***Infra-Red Absorption by Metals at Low Temperatures.** K. G. Ramanathan (*Proc. Phys. Soc.*, 1952, [A], **65**, (7), 532–540).—Since the absorption coeff. of a metal is a function of the elect. conductivity on classical theory, supraconductors should be perfect reflectors. Experiments were carried out on Sn, Pb, and Sn–Bi alloys using Cu to check the constancy of the source. No change was found, and it is suggested that other absorption mechanisms are present, which operate at frequencies greater than kT_c/h, where T_c is the transition temp., k Boltzmann const., and h Planck const. The absorption of other alloys, brass, Eureka, and Staybrite, was also measured, and the results compared with current theory. —E. O. H.

A Note on the Theory of Conduction in Metals. E. H. Sondheimer (*Proc. Phys. Soc.*, 1952, [A], **65**, (7), 561–562).—A letter. It is shown that laws of conservation of energy and momentum restrict the number of modes of lattice vibration with which conduction electrons interact. This makes many alterations in expressions for relaxation time, and the variation of elect. and thermal conductivities with temp. The new theory is applied to the case of Bi, and it is shown that the mean free path of the conduction electrons should be abnormally long, in accord with experiment. —E. O. H.

***The Thermal Conductivities of Metals at Low Temperatures.** E. H. Sondheimer (*Proc. Phys. Soc.*, 1952, [A], **65**, (7), 562–564).—A letter. S.'s modification to the theory of metals is applied to thermal conduction. It is shown that agreement with experiment is unsatisfactory. Certain modifications are suggested, but it is considered that some more exact theory is essential to clear up the discrepancies.—E. O. H.

***The Two-Band Effect in Conduction.** R. G. Chambers (*Proc. Phys. Soc.*, 1952, [A], **65**, (11), 903–910).—The properties of a model consisting of two non-interacting conduction bands are considered, and general formulæ are developed for the various galvanomagnetic and thermoelect. effects in such a model. The effects are classified in two groups, one (the isothermal case) when the transverse temp. gradient is zero, the other (the isothermal case) when the transverse heat flow is zero. The former is the simpler case. The formulæ derived are applicable with any degree of degeneracy present, and in the discussion the results are correlated with other theoretical work.—E. O. H.

***Mean Free Path of Electrons in Metals and Semi-Conductors.** E. Justi, M. Kohler, and G. Lautz (*Abhandl. Braunschweig. wiss. Ges.*, 1951, **3**, 34–43).—In general, the sp. elect. resistance, ρ, of a conductor does not depend upon the geometrical form of the conductor. At low temp., however, the mean free path of the electrons in sufficiently thin sheets or wires becomes comparable with the geometrical dimensions of the conductor. Theories proposed by Fuchs (*Proc. Cambridge Phil. Soc.*, 1938, **34**, 100; *M.A.*, **5**, 83) to cover this condition in the case of thin sheets, and by Dingle for the case of thin wires (*Proc. Roy. Soc.*, 1950, [A], **301**, 545; *M.A.*, **18**, 13) are discussed, and extended to the case of semi-conductors. It is shown that the formulæ of Fuchs and of Dingle do not give results in agreement with experimental results of van den Berg (*Physica*, 1949, **15**, 65; 1950, **16**, 907; *M.A.*, **17**, 248; **18**, 643) relating to the values of ρ for thin Au wires at 4·2° K., and those of Justi (hitherto unpublished) relating to the values of ρ for Au crystals, a few mm. thick, at very low temp. The authors, to explain this discrepancy, introduce into the theory the notion of an increase of residual resistivity with cold working of the surface layer, and the associated increase of the effective surface area. This assumption of a "surface resistivity" gives theoretical results in good agreement with some of the experimental results obtained by Dingle. Values of ρ for Mylius-Au (*Z. anorg. Chem.*, 1911, **70**, 203), possibly the

purest metal ever produced, at liq.-He temp. are tabulated. For cylindrical Mylius-Au wires of dia. 0·3–3·75 mm., the residual resistance decreases from 0·1% to the uniquely low value of ~0·01% of the resistance at 0° C. 18 ref.
—J. S. G. T.

***Hall Effect and Electronic Configuration in the Noble Metals.** E. Justi and M. Kohler (*Abhandl. Braunschweig. wiss. Ges.*, 1951, **3**, 44–48).—Values of the Hall coeff., A_H, at 293° and 20° K., of the noble metals of the 4th Period, show a steady decline in value from the high positive value for Ru (44) through Rh (45) to negative values for Pd (46) and Ag (47); the same characteristic decrease of A_H from an initial high positive value to a low negative value characterizes the sequence of metals of the 5th Period, from Re (75) through Os (76) and Ir (77), followed by negative values for Pt (78) and Au (79). These sequences of values are satisfactorily explained in terms of Mott's two-band (*s* and *d*) atomic model (*Proc. Phys. Soc.*, 1935, **47**, 571; *M.A.*, 2, 416), and enable deductions to be made as to electronic configurations in the free and cryst. states of these metals. The abnormally high positive values of A_H of Re and Ru are explained on the assumption that even holes in the upper band may act as defect electrons.—J. S. G. T.

The Variation with Time of the Temperature in Electrical Contacts. M. Kohler and G. Ziolasek (*Abhandl. Braunschweig. wiss. Ges.*, 1952, **4**, 117–126).—Holm ("Technische Physik der elektrischen Kontakte", Berlin, **1941**) has given a solution of the variation with time in an elect. contact, after switching-on; he assumed spherical symmetry of the effective contact area, a condition which cannot be realized. K. and Z. now give a soln. for the case of cylindrical symmetry of the contact area, which corresponds better with tech. conditions. The temp. immediately after switching-on is found, by K. and Z., to be only ~40% of that found by Holm.
—J. S. G. T.

***Electrode Phenomena in Transient Arcs.** J. M. Somerville, W. R. Blevin, and N. H. Fletcher (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 963–970).—The formation of hot spots and the resulting marks left on anode and cathode during arcing have been studied. The arcs were of short duration and with currents up to 200 amp. With moderate currents, a single crater is left on the anode, while the cathode marks are less regular. At high currents, both anode and cathode show multiple craters. The growth of these marks is discussed, and at the anode it is shown to be due to heat conduction, while the cathode marks arise from motion of the emitting area. Metals studied were Sn, Cu, and Al.—E. O. H.

***The Behaviour of Thermoelectric Effects at Very Low Temperatures.** Max Kohler (*Abhandl. Braunschweig. wiss. Ges.*, 1951, **3**, 49–54).—The three principal laws of thermodynamics are used to discuss the thermoelect. properties of electronic conductors in the neighbourhood of 0° K. In the limit as $T \to 0$° K., the product of the elect. conductivity, σ, and the differential thermoelect. power, de_{ab}/dT, should tend to the limiting value 0. It follows, therefore, that only in cases where $\underset{T \to 0}{\mathrm{Lt}}\,\sigma \neq 0$ can one conclude that $\underset{T \to 0}{\mathrm{Lt}}\, de_{ab}/dT = 0$. If the Thomson coeff., μ, is given by $\mu = \beta \,.\, T^n$ in the neighbourhood of $T = 0$, Nernst's theorem gives, for the case $\underset{T \to 0}{\mathrm{Lt}}\,\sigma \neq 0$, the condition $n > 0$. The accepted condition hitherto, for this case, has been $n > 1$. It is claimed that the present discussion explains reported discrepancies between the thermodynamic and electronic treatment of thermoelect., effects at low temp.—J. S. G. T.

***Approximate Calculations of the Surface Impedance of a Metal in the Anomalous Region [in Which Deviations Occur from the Behaviour Predicted by Classical Skin-Effect Theory.]** P. M. Marcus ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 265–272).—An approximate theory of the surface impedance of a metal in the anomalous region, which contemplates the motion of electrons in a field as decreasing exponentially with depth from the surface, and then modifies and improves the theoretical results by introducing refinements of this concept, is developed.—J. S. G. T.

***Occurrence of Supraconductivity Below 1° K.** B. B. Goodman ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 71–72).—Al, Cd, Ga, Zn, Ru, and Os become supraconducting in the region of ~1° K., the resp. critical temp., T_c, being Al, 1·197°, Cd, 0·560°, Ga, 1·103°, Zn, 0·905°, Ru, 0·47°, and Os, 0·71° K. Values of the critical field H_0 are tabulated. The various results could all be accurately represented by parabolic relations of the form $H_c = H_0[1 - (T/T_c)^2]$. The following metals were found to be not supraconducting down to the temp. stated: Li, 0·08°; Na, 0·09°; K, 0·08°; Ba, 0·15°; Y, 0·10°; Ce, 0·25°; Pr, 0·25°; Nd, 0·25°; Mn, 0·15°; Co, 0·12°; Pd, 0·10°; Ir, 0·10°; and Pt, 0·10° K. Values of the electronic sp. heats for the supraconductors Ru and Os are low compared with the values for the five elements Fe, Co, Ni, Pd, and Pt, which range from 12×10^{-4} to 33×10^{-4} cal. mole^{-1} deg^{-2}.—J. S. G. T.

***[Studies in] Supraconductivity at the Clarendon Laboratory [Oxford, England].** K. Mendelssohn ([*U.S.*] *Nat. Bur. Stand. Circ.*, 1952, (**519**), 33–36).—The following resp. values of the transition temp., T_c, in zero magnetic field, $T_cM^{0\cdot5}$, and R_4/R_{90}, the ratio of the resistivity at He temp. (4° K.) and at 90° K. have been found for three Sn isotopes of mass M (116·2, 120·0, and 123·6) and for natural Sn of at. wt. 118·7: M, 116·2; T_c, 3·764°; $T_cM^{0\cdot5}$, 4·057; R_4/R_{90}, 5×10^{-3}; 120·0, 3·710°, 4·064, $1\cdot5 \times 10^{-3}$; 123·6, 3·653°, 4·061, 2×10^{-3}; natural Sn, 118·7, 3·727°, 4·061, $2\cdot5 \times 10^{-3}$. According to theory, the product $T_cM^{0\cdot5}$ is found to be const. Experiments on the transition of the elect. resistance in magnetic fields in the case of Pb, Sn, and Hg have shown that all these metals exhibit a tendency, towards lower temp., to have the transition spread over a finite interval of temp.; the susceptibility transition, on the other hand, was found to be confined to a sharply defined field and temp. The thermal conductivities, K, of Pb–Bi alloys contg. 0·02–10% Bi have been determined at 3°–11° K. The alloys contg. 0·02–0·1% Bi show a behaviour similar to that of pure Pb, but the curves for the supraconducting and normal metal gradually approach as the impurity content is increased. The alloy contg. 0·2% Bi shows a crossing over of the conductivity/temp. curves of the supraconducting and normal metal at a temp. below the transition temp. Another interesting feature is the gradual shift of the max. value of K to higher temp. with increasing % of Bi.—J. S. G. T.

2 — PROPERTIES OF ALLOYS

***Investigations on Cast Aluminium Alloys for Cylinder Heads.** H. Vossküler (*Aluminium*, 1952, 28, (4), 99–106).—V. describes tests leading to the development of Hydronalium 51 (Mg 5, Si 1, Mn 0·2%, balance Al) in which this alloy and various modifications thereof were compared with Hy 71 (Mg 7, Si 1, Mn 0·2%, balance Al) and three Cu- or Ni-bearing alloys (R.R.53, Y-alloy, and Al G3) in regard to hot tensile and creep properties. It is concluded that Hy 51 can readily be substituted for the usual cylinder-head alloys, and it has the advantage of structural stability. 16 ref.—M. A. H.

***The Effect of Dispersions on the Tensile Properties of Aluminium-Copper Alloys.** R. B. Shaw, L. A. Shepard, C. D. Starr, and J. E. Dorn (*Amer. Soc. Metals Preprint*, 1952, (37), 23 pp.).—The plastic properties of high-purity rolled, soln.- and pptn.-treated Al–Cu alloys were correlated with the mean free path between the $CuAl_2$ particles dispersed in the α-solid-soln.-phase. The alloys contained 3, 4, or 5% Cu. The mean free path between the $CuAl_2$ particles was calculated math. from counts of the number of particles observed in unit area of a polished microspecimen. True stress/true strain curves were determined on specimens in which the mean free path of the precipitate particles varies widely. Tests were made at temp. in the range 78°–295° K. P.S. and rate of strain-hardening were found to increase uniformly with decreasing mean free path. The deformation strength at low temp. is related to the mean free path by $\sigma(\varepsilon, T) = A(\varepsilon, T)\lambda^{-n_\varepsilon}$, where $\sigma(\varepsilon, T)$ = deformation stress at strain ε and temp. T, λ = mean free path between $CuAl_2$ particles, $A(\varepsilon, T)$ = parameter depending on ε and T, and n_ε = parameter depending on ε only. The results are in qual. agreement with the correlation between the tensile properties of heat-treated steels and the mean free ferrite path, made by Gensamer *et al.* (*Trans. Amer. Soc. Metals*, 1942, **30**, 983), although differing in some details. None of the dislocation theories for plastic deformation in polyphase systems are in complete agreement with the observations on Al–Cu alloys or with the earlier work on steels. 9 ref.—R. W. R.

***On the Phases of the Ternary System Aluminium–Iron–Silicon.** Marcel Armand (*Compt. rend.*, 1952, **235**, (23), 1506–1508).—Previous investigators of the Al corner of the Al–Fe–Si ternary diagram do not agree on the number of phases, their compn., or the regions of true stability. The liquation method consists essentially in slowly casting the alloys previously brought to the liq. state and drawing off the mother liquor at a convenient temp. in the solidification interval. Chem. analysis of the initial alloys and the mother liquor and structural examination of the solid confirmed the diagram of Gwyer and Phillips (*J. Inst. Metals*, 1927, **38**, 29) as modified by P. and Varley (*ibid.*, 1944, **69**, 317; *M.A.*, **10**, 384), with two important differences: (1) The crystals obtained from the $FeAl_3$ area seemed to correspond more closely to the compn. Fe_2Al_7. (2) The α-region must be sub-divided into three new regions, α_1, α_2, α_3, as follows:

Phase	Lattice	Appearance	Density g./c.c.	Analysis	
				Fe, %	Si, %
α_1	Cubic ($a = 12{\cdot}523$ Å.)	Rhomboidal dodecahedra .	3·50	27·3	7·0
α_2	Hexagonal ($a = 4{\cdot}95$ Å.) ($b = 7{\cdot}07$ Å.)	Regular hexagonal prismatic needles . .	3·58	29·2	11·3
α_3	Cubic ($a = 16{\cdot}00$ Å.)	Octahedral cubes . .	3·65	35·3	12.8

α_1 is the phase that, with Al, gives the characteristic eutectic generally known as "Chinese script"; α_2 has been isolated and analysed by other workers (Pratt and Raynor, *J. Inst. Metals*, 1951, **79**, 211; *M.A.*, **18**, 699; and Fink and Van Horn, *Trans. Amer. Inst. Min. Met. Eng., Inst. Metals Div.*, **1931**, 383; see *J. Inst. Metals* (Abstracts), 1930, **44**, 557); α_3 has not, so far as is known, been previously described. —J. H. W.

***The Liquid Immiscibility Region in the Aluminium–Lead–Tin System at 650°, 730°, and 800° C.** Morgan H. Davies (*J. Inst. Metals*, 1952–53, **81**, (8), 415–416).—The analyses of successive pairs of congruent soln., obtained by adding Sn to liq. Al–Pb or Al–Pb–Sn alloys, have been used to plot the miscibility gaps in the liq. alloys at 650°, 730°, and 800° C. —AUTHOR.

***Plastic Stress/Strain Relations of Alcoa 14S-T6 for Variable Biaxial Stress Ratios.** Joseph Marin, L. W. Hu, and J. F. Hamburg (*Amer. Soc. Metals Preprint*, 1952, (24), 22 pp.).—The results are described of an investigation of the plastic stress/strain relations in 14S-T6 Al alloy (Cu 4·4, Si 0·8, Mn 0·8, and Mg 0·4%); these relations were found for biaxial stresses produced by subjecting a thin-walled tubular specimen to axial tension and internal pressure. Const. axial stress/lateral stress ratio tests and two kinds of variable stress ratio tests were made. The const. stress ratio tests showed that the biaxial P.S. values were in approx. agreement with the distortion energy theory, but that the biaxial fracture strength was in better agreement with the max. stress theory. Comparison of the true biaxial stress/strain relations for various principal stress ratios, with values predicted by the flow and deformation theories, showed approx. agreement in both cases. The ductility was found to increase with increase in the biaxiality of the stresses. The ductility values obtained were in approx. agreement with those predicted by both the significant strain and the deformation theories. The main object of the work was to determine the validity of the flow theory in predicting the plastic stress/strain relations under combined stress. Variable stress ratio tests were conducted for this purpose. In one type of test, a stress was applied in the axial or lateral direction to a selected plastic strain value, and this was followed by appn. of a second stress in a perpendicular direction. In the second kind of variable stress ratio test, a stress in one direction was followed by one in a perpendicular direction and then by a further stress in the first direction. The results obtained in both kinds of variable stress ratio test were in good agreement with the predictions of the flow theory. It was found that the path of loading had little effect on the true fracture strength, which was shown to be approx. given by the max. stress theory. Widely different strains were obtained when different paths of loading were used to arrive at the same final state of stress; this supports the flow theory rather than the deformation or slip theories, which predict that the final strain is independent of the path of loading. 7 ref.—R. W. R.

***Fatigue Strengths of 14S-T4 Aluminium Alloy Subjected to Biaxial Tensile Stresses.** J. Marin and W. P. Hughes ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (2704), 24 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 470).—Tests were made on thin-walled tubular specimens of 14S-T4 Al alloy under a combination of fluctuating axial tension and fluctuating internal pressure. Ratios of circumferential stress : axial stress were 0, $\frac{1}{2}$, 1, and 2, S/N curves being given up to $\sim 5 \times 10^6$ cycles. Results show the pronounced influence of anisotropy of the material, so that it is impossible to verify failure theories. Testing equipment and method are fully described.

***Semi-Conducting Properties of Intermetallic Compounds (CdSb).** Eduard Justi and Günter Lautz (*Abhandl. Braunschweig. wiss. Ges.*, 1952, **4**, 107–116).—In contradistinction to the known characteristic of certain binary alloys, e.g. Au–Cu, which show a *min.* of elect. resistivity, ρ, at the stoichiometric compn., J. and L. now report that other binary alloys, e.g. Cd–Sb and Mg–Sn, attain a *max.* value of ρ, some 4–5 times greater than normal, at approx. the stoichiometric compn. Measurements of the temp. coeff. of ρ at 10°–700° K., the Hall effect, the magnetic increase of ρ, thermoelect. e.m.f., and rectifying properties show that CdSb behaves as a semi-conductor; ρ at 293° K. can attain the value 2·4 ohm-cm.; it becomes semi-conductive by completion of the 7th Brillouin zone. The possibility of using such semi-conducting compounds in place of Ge and Si in diodes and transistors is briefly discussed.—J. S. G. T.

***A Thermodynamic Study of Liquid Metallic Solutions. IV.—Approximate Thermodynamic Data from the Phase Diagrams for the Systems Copper–Bismuth, Copper–Lead, and Copper–Thallium.** O. J. Kleppa (*J. Amer. Chem. Soc.*, 1952, **74**, (23), 6047–6051).—Cf. *ibid.*, 1951, **73**, 385; *M.A.*, **18**, 650. Use of the binary equilibrium phase diagram as a source of thermodynamic data for liq. metallic mixtures is discussed, and a method of calculation is presented which permits sepn. of the free energy of mixing into approx. heat and entropy terms. It is considered that the treatment should be particularly useful in the study of mixtures unsuited to investigation by the e.m.f. and vapour-pressure methods, and its appn. to the systems Cu–Bi, Cu–Pb, and Cu–Tl is described. Some new data on the solubility of Cu in the low-melting metal are obtained.—J. R.

***On the Electrical Properties of Copper–Manganese–Tin Alloys.** Hideo Nishimura, Masao Adachi, and Hiroshi Nakagawa (*Bull. Eng. Research Inst., Kyoto Univ.*, 1952, **1**, (March), 44–48).—[In Japanese, with tables and graphs in English]. An investigation has been made of the elect. properties of Cu–Mn–Sn alloys contg. 2–18 wt.-% Mn and 4–10 wt.-% Sn, balance Cu. The sp. resistance increases with cold drawing, decreases on annealing at lower temp., and increases on annealing at higher temp. (Annealing temp. of

200°, 300°, 400°, and 500° C. were used). The thermo-e.m.f. against annealed Cu changes in a similar manner.
—Authors.

*__Tin Bronzes.__ M. R. Reeve, J. S. Bowden, and J. W. Cuthbertson (*Metal Ind.*, 1953, 82, (2), 23–25; (3), 49–52).—A detailed microscopical investigation was made of the phases present in bronzes contg. 15, 20, 25, and 28% Sn. The alloys were prepared by melting under an oxidizing flux in an induction furnace, deoxidized with phosphor Cu, and slowly chill cast. Specimens were packed in Al$_2$O$_3$ in a mild-steel boat and heat-treated in a purified N atmosphere. They were examined in the as-cast, in the annealed, and in the furnace-cooled condition, and after quenching from various temp., and the photomicrographs form a virtual atlas of bronze structures.—J. H. W.

*__The Indium–Arsenic System.__ T. S. Liu and E. A. Peretti (*Amer. Soc. Metals Preprint*, 1952, (40), 9 pp.).—The phase diagram of the system In–As was determined by means of thermal, X-ray, and metallographic methods. A eutectic reaction was found to occur at 155·2° ± 0·2° C. and 0·2% As between In and InAs; a second eutectic was found at 731° ± 1° C. with 82 ± 2% As at a pressure of 36 atm. between InAs and As. The intermediate phase, InAs, corresponds closely with the stoichiometric compn., and melts at 942° ± 0·5° C. This phase has a f.c.c. structure of the zinc-blende type, with a lattice parameter of 6·058$_4$ Å. The single-phase regions in the solid state are quite small. 4 ref.—R. W. R.

*__The Magneto-Resistance of Ferromagnetic Fe–Al–Si Alloys.__ R. Parker (*Proc. Phys. Soc.*, 1952, [B], 65, (8), 616–620).—Following an earlier paper (*ibid.*, 1951, [A], 64, 447; *M.A.*, 19, 116) P. derives an expression for the magneto-resistance at saturation for mixed ferromagnetic alloys. This is applied to three dilute alloys of Al and Si in Fe, whose magneto-resistance was measured at temp. from −78° to 300° C. Agreement with theory is reasonably good. Below saturation, some unusual variations were found, and these are briefly discussed.—E. O. H.

*__A Study of the Transformation of the Alpha Solid Solution in the Iron–Chromium System by the Transformation-Rate Method.__ I. I. Kornilov and V. S. Mikheev (*Doklady Akad. Nauk S.S.S.R.*, 1949, 68, (3), 527–530).—[In Russian]. Alloys in the compn. range 40·5–51·5 at.-% Cr were quenched from the α-region, and the changes in magnetic saturation observed during transformation at 650° C. The reciprocal of the time for 50% transformation plotted against compn. showed 3 peaks at 42·44, 46·44, and 49·5 at.-% Cr, resp. The transformation peak at 50% is identified with σ formation, and two new phases, β and θ, are suggested. The existence of β and θ was confirmed by a curve of Curie point against compn., which shows breaks at the regions corresponding to non-magnetic β, θ, and σ, the curve being linear over the regions where α is retained. 16 ref.—D. M. P.

*__The Electrolytic Separation and Some Properties of Austenite and Sigma in 18 : 8 : 3 : 1 Chromium–Nickel–Molybdenum–Titanium Steel.__ T. P. Hoar and K. W. J. Bowen (*Amer. Soc. Metals Preprint*, 1952, (4), 19 pp.).—Austenite and σ have been extracted from 18 : 8 : 3 : 1 Cr–Ni–Mo–Ti steel, heat-treated for various times at 1560° F. (850° C.), by selective anodic dissoln. in 25% w/w H$_2$SO$_4$ at 0·4 amp./in.^{2} and in 50% v/v HCl at 30 amp./in.^{2}, resp. The extracted austenite partially transforms to α_2 at room temp., the transformation being the greater, the longer the previous heat-treatment. There is no further transformation at −321° F. (−196° C.). This contrasts with the behaviour of the austenite in the massive material, which does not transform at room temp. but does so at −321° F.; the easier transformation of the extracted austenite is probably due to the release of internal microstresses. The extracted σ is ferromagnetic at low temp. It is much higher in Cr and Mo, higher in Ti, and much lower in Ni, Fe, and Mn than the corresponding austenite. Electrode-potential measurements of compacted austenite and σ residues in NaCl and HCl soln. show that σ has a greater tendency to become passive, and has a more noble passive potential than the corresponding austenite.—Authors.

*__Fundamental Effects of Cold Working on Creep Properties of Low-Carbon N-155 [Fe–Co–Cr–Ni] Alloy.__ D. N. Frey, J. W. Freeman, and A. E. White ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1951, (2472), 45 pp.; *Appl. Mechanics Rev.*, 1952, 5, 215).—The report, which is part of a continuing investigation of the fundamental behaviour at high temp. of austenitic alloys used in aircraft engines, covers the creep behaviour of low-C N-155 (C 0·12, Ni 20, Cr 20, Co 20, W 2, Nb 1, Mo 3, N 0·12%, balance Fe) subjected to varying amounts of cold working. Creep tests were carried out at a stress of 50,000 lb./in.^{2} and temp. of 1200° F. (650° C.) for 200 hr. on samples which had been cold-worked to varying amounts. Fundamental effects of cold rolling on lattice structure were established by microscopic examination and X-ray diffraction. Cold working was found to improve the creep-resistance of this alloy by the introduction of residual elastic stresses into the lattice. Correlation appears to exist between the X-ray-diffraction measurements and the creep rate which may serve as a control for the optimum amount of cold reduction.

*__Fundamental Effects of Cold-Work on Some Iron–Cobalt–Chromium–Nickel-Base Creep-Resistant Alloys.__ D. N. Frey, J. W. Freeman, and A. E. White ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (2586), 12 pp.; *Appl. Mechanics Rev.*, 1952, 5, 430).—Cf. *ibid.*, 1951, (2472); preceding abstract. Describes the influence of cold working on the creep properties of an Fe–Co–Cr–Ni alloy subjected to minor changes in chem. compn. The authors have concluded from earlier work (*loc. cit.*) that there are certain correlations among creep properties, internal stresses, and width of X-ray-diffraction lines. Using these relationships, the effects of cold work on an Fe–Co–Cr–Ni alloy, were evaluated for four different minor chem. changes in the base alloy. The base alloy and its variations all showed the same creep behaviour after cold work, indicating that internal stress-relaxation characteristics of the alloys examined are independent of minor changes in chem. compn.

*__On the Invar Problem.__ Yosiaki Tino (*J. Sci. Research Inst.* (*Tokyo*), 1952, 46, (1283), 141–146).—[In English]. A min. value of the coeff. of expansion of the γ-phase of the Fe–Ni alloys characterizes the alloy contg. 36% Ni, known as Invar. Guillaume attempted to explain the occurrence of this min. value in terms of a hypothetical compound Fe$_2$Ni, which has not yet been shown to exist. The many theories that have been proposed to explain the phenomenon can be summarized in the theory due to Dehlinger (*Z. Metallkunde*, 1936, 28, 194; *M.A.*, 3, 392), which considers the low expansibility as a property characteristic of the ferromagnetic alloy in the γ-phase, and in the theory due to Benedicks, which attributes the phenomenon to a two-phase transformation (*Arkiv. Mat. Astron. Fysik*, 1942, [A], 28, (14); *M.A.*, 10, 318). These theories are briefly reviewed, but neither is considered entirely satisfactory. A new theory, attributing the phenomenon of the low expansibility of Invar to the presence of dispersed α-phase globules in the γ-matrix of the alloy is developed.—J. S. G. T.

*__Heat Transport in Lead–Bismuth Alloys.__ J. L. Olsen (*Proc. Phys. Soc.*, 1952, [A], 65, (7), 518–532).—The thermal conductivity of a series of Pb–Bi alloys (0–30% Bi) has been measured at very low temp., both with and without the presence of a magnetic field. The results are discussed in terms of the processes relating to heat conductivity and the supraconducting state, for it is found that the conductivity is less when the alloy is supraconducting. This is in contrast to the case of pure supraconductors.—E. O. H.

*__The Influence of Thallium on the Creep of Lead.__ R. C. Gifkins (*J. Inst. Metals*, 1952–53, 81, (8), 417–425).—Extruded specimens of high-purity alloys of Pb with addn. of Tl in the solid-soln. range have been tested at room temp., under const. load, after a standard annealing treatment. At a stress of 500 lb./in.^{2}, progressive addn. of Tl produced first a decrease in min. creep rate, then with >0·5% Tl a marked increase, and above ~8% a decrease again. Finally, at 40% Tl, the creep rate increased and corrosion and cracking occurred. Tests at lower stresses revealed similar but less marked changes. The extension on loading decreased regularly, but the complex variations in extension were present after one day's creep. The average creep rate and time to rupture showed variations similar to those of the min. creep rate. The elongation at rupture was a max. at any given stress for alloys cont. $\sim$8% Tl, and this elongation was greater the lower the stress; the largest extension recorded was 372%. The variation in creep characteristics is explained in terms of the proportions of slip and grain-boundary movement, which depend on the grain-size, particularly in alloys cont. $<$8% Tl. This has been confirmed by tests on specimens of one of the alloys prepared in a range of grain-sizes; the creep rate varied in a similar manner to the alloys of comparable grain-size. In alloys cont. $>$8% Tl, although the grain-size continued to diminish, the decreased creep rate was apparently due to strengthening of the grain boundaries, since grain-boundary flow still appeared to be the major mechanism of deformation. It is suggested that this change in creep-resistance at $\sim$8–10% Tl may be connected with the rate of change of lattice spacing, as Tang and Pauling (*Acta Cryst.*, 1952, **5**, 39; *M.A.*, **20**, 13) have recently shown that only a slight decrease in lattice spacing takes place up to 10% Tl, but that it subsequently decreases more rapidly.—AUTHOR.

***Electrical Conductivity in the Compounds PbS, PbSe, and PbTe.** E. H. Putley (*Proc. Phys. Soc.*, 1952, [B], **65**, (5), 388–389).—A letter. Measurements are reported of the elect. conductivity of single crystals of PbS, PbSe, and PbTe. Mobilities are 2–3 times greater than in polycryst. specimens, and the variation with temp. does not follow an exponential law (as was found in polycryst. specimens, and predicted theoretically), but followed the law: $v = v_0T^{-5/2}$, which is not in agreement with any theory. Measurements of mobility in thin films are also discussed, and it is concluded that it is unwise to use values for film as indicative of single-crystal properties.—E. O. H.

***Intrinsic Conduction in PbS, PbSe, and PbTe.** E. H. Putley (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 993).—A letter. Values of the energy-band gap for PbS, PbSe, and PbTe have been measured, using both Hall-effect data and mobility data. Results are of the same order as those obtained from photoconductivity measurements.—E. O. H.

***The Absorption Spectra of Single Crystals of Lead Sulphide, Selenide, and Telluride.** A. F. Gibson (*Proc. Phys. Soc.*, 1952, [B], **65**, (5), 378–388; corrigendum, (7), 555).—Experimental apparatus is described, and curves are drawn for the absorption coeff. of these compounds for wave-lengths from 2 to 12 μ, and at temp. from 20° to 590° K. All compounds show a sharp absorption edge, which moves to longer wave-lengths as the temp. is lowered. Unfortunately a factor of $4\pi^2$ was omitted from the theoretical calculation, and so agreement with measured mobilities of carriers is not as good as G. claims. The long-wave absorption increases with increasing temp., and the results are correlated with photoconductivity. The results are discussed in terms of energy-band transitions.—E. O. H.

***Transistor Action in Lead Telluride.** C. A. Hogarth (*Proc. Phys. Soc.*, 1952, [B], **65**, (12), 958–963).—Experiments are described on a single crystal of *p*-type PbTe. No transistor action was observed at 290° K., but at 90° K. results similar to *p*-type PbS were obtained, and gains of up to 11 recorded.—E. O. H.

***A New Permanent Magnet from Powdered Manganese Bismuthide.** Egmond Adams, William M. Hubbard, and Albert M. Syeles (*J. Appl. Physics*, 1952, **23**, (11), 1207–1211).—The compound MnBi is prepared by melting a Bi-rich mixture in a tilting crucible. When the sluggish peritectic reaction in which MnBi is formed is over, excess Bi is decanted. The mass is powdered in He in a hammer mill, Mn largely removed by magnetic sepn., and the enriched MnBi powder pressed at low pressure at $\sim$300° C. in a pulsating magnetic field. The low pressure allows the particles to align themselves magnetically. Both the residual flux density B_r and the coercive force H_c were very high in well-oriented samples; H_c in particular exceeds that for any known "practical" magnet.—R. W. C.

***Properties of Some Hydrogen-Sintered Binary Molybdenum Alloys.** W. L. Bruckart, M. H. LaChance, C. M. Craighead, and R. I. Jaffee (*Amer. Soc. Metals Preprint*, 1952, (**36**), 25 pp.).—An investigation was made of the room-temp. mech. properties, hot hardness, and workability of binary alloys of Mo with the following elements; Al, B, Be, C, Co, Cr, Fe, In, Mn, Ni, P, S, Si, Sn, Th, and W. The alloys were prepared by sintering mixed powders of the component elements in H at 1650°–2300° C.; the alloys were subsequently hot-rolled and annealed in a H atmosphere. The addn. most effective (% of alloying element) in raising the room-temp. U.T.S. and the hardness at temp. up to 800° C. were Si, Co, Fe, Al, Cr, and W in that order. The highest strengths were observed in Mo–W alloys, since W may be alloyed in large quantities. The most useful combinations of strength and ductility were produced by the following addn. (% added, not % retained): (*a*) 40% W, (*b*) 0·4% Co, (*c*) 30% W, (*d*) 0·3% Co, (*e*) 0·25% Co, (*f*) 0·25% Fe, (*g*) 0·4% Fe, (*h*) 1·0% Cr, (*i*) 20% W, and (*j*) 0·25% Si. The best combinations of strength, ductility, and ease of fabrication were found in alloys contg. the following addn.: (i) 0·25% Co, (ii) 0·25% Fe, (iii) 1·0% Cr, (iv) 20% W, (v) 0·25% Si, and (vi) 1·0% Al; all these alloys have U.T.S. $>$110,000 lb./in.^{2} and elongations $>$10%. The U.T.S. (lb./in.^{2}) of the alloys studied was found to be approx. 400 times the D.P.N. The microstructures of some of the alloys are illustrated. 4 ref.—R. W. R.

***The Spectroscopic Factor, *g*, of the Ferromagnetic Nickel–Copper Alloys.** A. Jean Pierre Meyer (*Compt. rend.*, 1952, **235**, (22), 1382–1385).—The factor g was determined for a number of Ni–Cu alloys from 0% Cu to the compn. Cu_3Ni_2. Resonance was determined by the variation of absorption of energy of a specimen subjected in its plane to an alternating field of fixed frequency and a ⊥ static field of variable intensity. It was confirmed that g is const. at $\sim$2·2 for values of the ratio of the abs. temp. to the Curie point from 0·5 to 1, in agreement with Bloembergen (*Phys. Rev.*, 1950, [ii], **78**, 572; *M.A.*, **18**, 341).—J. H. W.

Developing the Nimonic Alloys. —— (*Aeroplane*, 1953, **84**, (2166), 122–124).—The different types of heat-resisting Nimonic alloys are briefly described. A short account is given of present research on these alloys, and of their prodn. and testing.—S. V. R.

***Binary Alloys of Thallium with Sulphur, Selenium, and Tellurium.** A. P. Obukhov and N. S. Bubyreva (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, **19**, 276–280; *C. Abs.*, 1951, **45**, 4125).—[In Russian]. The 3 systems were subjected to thermal analysis and to microstructural examination. The existence of the following compounds was ascertained: Tl_8S_7, Tl_2S; Tl_2Se_3, TlSe, Tl_2Se; and TlTe, Tl_5Te. The equilibrium diagrams of the systems are given.

***Mechanical Properties, Including Fatigue, of Titanium-Base Alloys RC-130-B and Ti-150-A at Very Low Temperatures.** S. M. Bishop, J. W. Spretnak, and M. G. Fontana (*Amer. Soc. Metals Preprint*, 1952, (**31**), 14 pp.).—Tensile, impact, fatigue, and hardness tests were made on annealed rods of two Ti-base alloys, RC-130-B (Fe 1·3, Cr 2·7, N 0·08 max., and C 0·05% max.) and Ti-150-A (C 0·24, Al 3·8, Mn 3·8%), at temp. in the range 25° to −253° C. The thermal expansivity of the alloys was determined in the range 25° to −196° C. The fatigue strength of RC-130-B rises from 85,000 lb./in.^{2} at 25° C. to 130,000 lb./in.^{2} at −196° C.; corresponding figures for Ti-150-A are 110,000 and 142,000 lb./in.^{2}. Some information is given on the notched fatigue properties of the alloys. The tensile strength of RC-130-B increases from 149,000 lb./in.^{2} at 25° C. to 256,000 lb./in.^{2} at −196° C.; the tensile strength of Ti-150-A rises from 186,000 lb./in.^{2} at 25° C. to 200,000 lb./in.^{2} at −78° C. and thereafter declines to 177,000 lb./in.^{2} at −196° C. The yield strength of both alloys is nearly equal to the U.T.S., and is affected by temp. in the same way. The impact resistance and the ductility of both alloys decreased to low values, and the notch sensitivity increased, with falling temp. The Vickers hardness of both alloys increased by $\sim$107% as the temp. fell from 25° to −253° C. The coeff. of thermal expansion of both alloys was reduced as the temp. fell, in the case of RC-130-B from 16·7 × 10^{-6} in./in./°C. at 20° C. to 5·3 × 10^{-6} in./in./°C. at −200° C.; in the case of Ti-150-A the coeff. fell from 10·6 × 10^{-6} to 4·0 × 10^{-6} in./in./°C. over the same temp. range. 4 ref.—R. W. R.

***The Influence of Insoluble Phases on the Machinability of Titanium.** R. M. Goldhoff, H. L. Shaw, C. M. Craighead, and R. I. Jaffee (*Amer. Soc. Metals Preprint*, 1952, **(30)**, 25 pp.).—The machinability of Ti and Ti containing various alloying addn. was investigated by means of turning and sawing tests. The machinability of Ti is sensitive to the purity of the material; increasing the purity from that of the commercial Mg-reduced grade (99·6–99·9%) to that of iodide Ti (99·96% Ti) doubled the machinability. Machinability was also found to be highly sensitive to machining conditions; for example, an increase in tool thrust load or a decrease in cutting speed increases the machinability of Ti much more than the machinability of 18 : 8 stainless steel under the same conditions. Small amounts (up to 1%) of various elements were added to Ti in an effort to improve the machinability. B, As, Ge, S, Te, and Se produced slight or moderate improvement in machinability; P, Si, and especially C, were harmful. 8 ref. —R. W. R.

***The Titanium–Oxygen System.** E. S. Bumps, H. D. Kessler, and M. Hansen (*Amer. Soc. Metals Preprint*, 1952, **(32)**, 18 pp.).—The phase diagram of the Ti–O system was determined from 0 to 30% O by micrographic and X-ray methods. Two peritectic reactions were found. The first, melt + $\alpha \rightleftharpoons \beta$ occurs between 1 and 5% O at approx. 1740°C.; the second, melt + $\alpha \rightleftharpoons$ TiO occurs between 14·5 and 20·5% O at approx. 1770° C. There is also a peritectoid reaction, α + TiO $\rightleftharpoons \delta$ between 14·5 and 23·5% O at approx. 925° C. The α phase has a max. m.p. of ~1900° C. at approx. 10% O. The solubility limit of O in the α phase is approx. 14·5% between 800° and 1700° C.; the max. solubility of O in β-Ti is ~1·8% at 1740° C. The solubility limit of Ti in the α phase decreases from 100% at 880° C. to ~95·5% at 1740° C. The intermediate phase TiO extends between the limits 20·5 and 29·5% O at 1400° C. and 23·5% and 29·5% at 800° C. The TiO phase has a cubic structure of the NaCl type, the lattice parameter decreasing linearly with increasing O content (at.-%). The existence of the δ phase was confirmed by X-ray studies; this phase appears to possess tetragonal symmetry. The Vickers hardness of Ti is at first markedly increased by addn. of O, but later falls to a min. in the α + TiO phase mixture alloys; it increases again sharply for the TiO solid soln. alloys to values in excess of 1300 D.P.N. 15 ref.—R. W. R.

***The Kinetics of the Eutectoid Transformation in Zinc–Aluminium Alloys.** R. D. Garwood and A. D. Hopkins (*J. Inst. Metals*, 1952–53, **81**, (8), 407–415).—Time/temp./transformation diagrams have been constructed for a Zn alloy contg. 22·5% Al from data obtained in dilatometric, metallographic, and hardness studies. The methods agree as to the times for the start of transformation, yielding a C-shaped curve, with the transformation taking place most rapidly at 150° C. Discrepancies occur, however, in the times for the completion of the transformation; these are discussed and related, on X-ray evidence, to the slow change in compn. of the α constituent in the decompn. product. Age-hardening studies on isothermally transformed specimens are included to explain peculiarities in the hardness results observed at the nose of the C curve. A four-stage mechanism is postulated for the transformation at low temp. The reaction rate becomes controlled by diffusion at a lower temp. than is expected from the theoretical "spinodal" line calculated by Borelius and Larsson (*Arkiv Mat. Astron. Fysik.*, 1948, [A], **35**, (13); *M.A.*, **16**, 197).—AUTHORS.

***A Thermodynamic Study of Liquid Metallic Solutions. V.—The Systems Zinc–Bismuth and Zinc–Lead.** O. J. Kleppa (*J. Amer. Chem. Soc.*, 1952, **74**, (23), 6052–6056).—Cf. *ibid.*, p. 6047; *M.A.*, **20**, 622. A study of the thermodynamic properties of the liq. Zn–Bi and Zn–Pb systems by the e.m.f. method shows relatively large positive ΔH (1·1 and 1·6 kg.cal./g.-atom for equi-atomic Zn–Bi and Zn–Pb, resp.) and positive entropy deviations in both cases. Consideration of the liq. miscibility gaps in the light of equilibrium measurements gives excellent agreement with earlier data for Zn–Pb, but for Zn–Bi a critical temp. of ~605° C. at 85 at.-% Zn is indicated, as compared with the previously accepted value of 750°–800° C.—J. R.

***Some Properties of High-Purity Zirconium and Dilute Alloys with Oxygen.** (Treco). See col. 614.

***The System Zirconium–Silicon.** C. E. Lundin, D. J. McPherson, and M. Hansen (*Amer. Soc. Metals Preprint*, 1952, **(41)**, 14 pp.).—Metallographic, X-ray-diffraction, and thermal-analysis techniques were employed to determine the phase diagram of the Zr–Si system; particular attention was given to the Zr-rich portion. The alloys used were prepared by arc-melting under a He atmosphere. There are 7 intermediate phases Zr_4Si (7·14% Si), Zr_2Si (13·33% Si), Zr_3Si_2 (17·02% Si), Zr_4Si_3 (18·74% Si), Zr_6Si_5 (20·40% Si), ZrSi (23·53% Si), and $ZrSi_2$ (38·09% Si). Only Zr_6Si_5 displays an open max. melting temp. (2250° C.), the remainder being formed peritectically. The peritectic temp. and the compn. of the reacting melts are: (*a*) for Zr_4Si, 1630° C., 4·5% Si; (*b*) for Zr_2Si, 2110° C., 9% Si; (*c*) for Zr_3Si_2, 2210° C., 14·4% Si; (*d*) for Zr_4Si_3, 2225° C., 18% Si; (*e*) for ZrSi, 2095° C., 24·2% Si; and (*f*) for $ZrSi_2$, 1520° C., 46% Si. Two eutectics are present, one between Zr and Zr_4Si at 2·9% Si and 1610° C., and the second between Si and $ZrSi_2$ at 75% Si and 1355° C. The solubility of Si is <0·1% in α-Zr and <0·2% in β-Zr. Numerous photomicrographs illustrate the structures observed. 5 ref.—R. W. R.

***The System Zirconium–Tin.** D. J. McPherson and M. Hansen (*Amer. Soc. Metals Preprint*, 1952, **(34)**, 17 pp.).—The phase diagram of the Zr–Sn system was established by means of metallographic, X-ray-diffraction, and thermal-analysis methods. The alloys used were arc-melted under a He atmosphere. Three compounds exist in this system: (*a*) Zr_4Sn (24·55% Sn) formed by the peritectoid reaction $\beta + Zr_5Sn_3 \rightleftharpoons Zr_4Sn$ at 1325° ± 20° C.; (*b*) Zr_5Sn_3 (43·84% Sn), which melts with an open max. at 1985° ± 25° C.; and (*c*) ZrSn (56·55% Sn) formed by the peritectic reaction Zr_5Sn_3 + melt (83% Sn) $\rightleftharpoons$ ZrSn at 1140° ± 15° C. There is a eutectic at 23·5% Sn and 1590° ± 15° C. The solubility of Sn in β-Zr falls from 21% at the eutectic temp. to 6·5% at 980° C. The $\alpha \rightleftharpoons \beta$ transformation temp. of Zr is raised by Sn, and this results in a peritectoid reaction $\beta + Zr_4Sn \rightleftharpoons \alpha$ at 980° ± 20° C. The solubility of Sn in α-Zr decreases from 9% at 980° C. to 1·5% at 600° C. Zr is not soluble in Sn to any appreciable extent. The (undetermined) eutectic or peritectic between Sn and ZrSn must occur at a Sn content >99·5% and within 1° C. of the m.p. of Sn. The structures observed in the alloys are illustrated by 24 photomicrographs. 6 ref.—R. W. R.

Jets Test Designers' and Metallurgists' Ingenuity. [**Properties of High-Temperature Alloys**]. R. B. Johnson (*Iron Age*, 1950, **166**, (6), 73–78).—The thermal-shock and fatigue properties of some high-temp. alloys used in jet engines and turbo-superchargers are discussed. Examples of some failures are given.—S. R. W.

Approximate Strength of Important Jet-Engine Alloys (Round Test-Bars). G. H. Boss (*Metal Progress*, 1951, **60**, (5), 80-B).—A data sheet giving the properties of some high-temp.-resistant alloys.—S. R. W.

On the Problem of the High-Temperature Strength of Alloys. I. Dekhtyar (*Doklady Akad. Nauk S.S.S.R.*, 1950, **73**, (2), 303–305).—[In Russian]. Consideration of existing data shows that for a number of metals the activation energy of self-diffusion, $E_a = k \times$ (bonding energy, E_b); for co-ordination No. 12 and 8, $k \cong 0{\cdot}65$ and $\cong 0{\cdot}85$, resp. It is also noted that the activation energies for high-temp. creep and for self-diffusion show close similarity; it is claimed that this indicates a similar mechanism, and that E_a may be taken as a criterion of the high-temp. strength of alloys. The connection between the activation energies and the character of the $N(E)$ curves is pointed out; for the same co-ordination number and electron concentration, a smaller bonding energy corresponds to a higher value of $N_{\max.}(E)$. Several examples are given to illustrate this. Since studies of self-diffusion have shown that the most probable mechanism is the displacement of vacancies by diffusing atoms, it is deduced that lattice vacancies are important in creep. Creep rate depends on the rate of formation of vacancies and on the rate at which they are filled. 10 ref.—D. M. P.

An Analytical Expression Relating the Melting Point of Alloys to Their High-Temperature Strength. K. A. Osipov (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (1), 81–82).—[In Russian]. Experiment has verified that equations analogous to that developed by Mott (*Proc. Phys. Soc.*, 1948, **60**, 391; *M.A.*, **16**, 338) for the velocity of slip at grain boundaries of pure metals can be applied to solid soln. The following equation is developed, assuming that the activity coeff. of the solute are the same in the liq. and solid phases:

$$v_c = \frac{\nu_{av} a^2 n \omega \sigma}{kT} \cdot \exp\left\{-n \ln \frac{A_l}{A_s}\left(\frac{T_s}{T} - 1\right)\right\},$$

where v_c = slip velocity, ν_{av} = average frequency of atomic vibration in the solid soln., a = half the lattice const., ω = area covered by one atom, σ = stress, n = number of atoms which must be disordered, A_l and A_s = atomic fractions of solute in liq. and solid phases, T_s = abs. temp. of slip, T = solidus temp., and k = Boltzmann's const. It is deduced that high-temp. strength will increase with solidus temp. and width of solidification range, but other factors must be considered. 5 ref.—D. M. P.

Continuous Solid Solutions of Transition Elements. I. I. Kornilov (*Doklady Akad. Nauk S.S.S.R.*, 1950, **73**, (3), 495–497; *C. Abs.*, 1951, **45**, 4111).—[In Russian]. The rules for continuous-solid soln. formation that have been found to hold for Fe-base alloys are assumed to hold also for all transition elements. Data from the literature are available on 15 out of the 21 possible binary systems of f.c.c. metals and all of these show complete solid solubility when the difference in lattice-const. values is $<10\%$. Data on 5 out of 15 b.c.c. systems give the same result. No data have been found for systems involving two hexagonal metals.

Note on the Theory of Alloys. N. S. Akulov (*Doklady Akad. Nauk S.S.S.R.*, 1949, **66**, (3), 361–364).—[In Russian]. The influence of disorder–order changes in alloys on the phys. properties is discussed math. A general equation for the saturation magnetic moment J_s is evolved for the case of complete disorder:

$$J_s = N\sum_i \mu_{ii} C_i + NZ\sum_{ij} a_{ij} C_i C_j,$$

where N = nodes/c.c., μ_{ii} = mean magnetic moment of an A_i atom surrounded by other A_i atoms, C_i = concentration of A_i, Z = co-ordination number, and a_{ij} = increment in moment resulting from replacement of one neighbouring A_i atom by an A_j atom. Good agreement with the equation is obtained in binary Fe and Ni alloys, and allows μ_{ii} and a_{ij} to be calculated. The degree of order (σ) is introduced, and in the case of (e.g.) a 50 : 50 alloy is defined by $Z_{12} = Z_{21} = \frac{1}{2}Z + \sigma\frac{1}{2}Z$, where $0 \leqslant \sigma \leqslant 1$; thus for any degree of order $J_s = \frac{1}{2}N(\mu_{11} + \mu_{22}) + \frac{1}{4}ZN(a_{21} + a_{12})(1 + \sigma)$. If the curve of J_s/C_2 for complete disorder is obtained and the tangent taken at $C_2 = 0$, it is deduced that the segment ΔJ_s between the curve and the tangent at the concentration C_2 concerned will give the increment in magnetic moment corresponding to the change from disorder to order. The variation of the critical temp., θ, is dealt with in an exactly similar manner on the basis of the existence of parallel and anti-parallel spins.—D. M. P.

The Log–Log Plot of Solubility Data in Ternary Metallic Systems. H. K. Hardy (*J. Inst. Metals*, 1952–53, **81**, (8), 432).—The slope of the solubility curve of an ideal ternary soln. plotted on log–log co-ordinates has been analysed for the case when the phase in equilibrium is a ternary compound. A straight line is not predicted, but the slope will be equal to the ratio of the solute components in the compound when the same ratio exists between the solute components in the soln. in equilibrium. The curvature will be slight for very dilute soln., whose solubility data will give a straight-line plot as though in equilibrium with a binary compound.—Author.

3 — STRUCTURE

(Metallography; Macrography; Crystal Structure.)

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

Micrographic Determination of the Grain-Size of Aluminium and Its Alloys. Jean Hérenguel (*Rev. Aluminium*, 1952, (189), 221–229).—H. considers the techniques of chem. etching, chem. and electrolytic polishing, and anodizing in relation to their use in developing the grain structure of Al and Al–Mg alloys. Compn. and operating conditions are given for the various methods, and the structures developed by these means are illustrated.—A. W. B.

***Recrystallization of Pure Aluminium.** Akiyoshi Yasumori and Seiichi Ishii (*Rep. Osaka Pref. Indust. Research Inst.*, 1951, **3**, (2), 3–9; *C. Abs.*, 1953, **47**, 85).—The electropolished surface of pure Al sheet after immersion in KNO_3–$NaNO_3$ soln. for periods of 10 sec.–2 hr. showed 2 max. in Vickers' hardness before recrystn. was complete. In the 1st max., as calculated from the activation energy, a shift of Polanyi's dislocation occurred and was observed microscopically as the sub-grains acquiring boundary lines. At the 2nd max. the activation energy calculated was not very strong, though it was enough to indicate a structural change accompanied by atomic diffusion. Electron microscopically, a fine structure (perhaps Cahn's polygonization) was observed.

***Formation of Microscopic Cracks in Aluminium During Plastic Deformation.** V. A. Pavlov (*Doklady Akad. Nauk S.S.S.R.*, 1951, **78**, (4), 677–679; *C. Abs.*, 1953, **47**, 84).—[In Russian]. The formation of microcracks as stress was applied to specimens of Al (1 × 5 × 50 mm.) was observed through a microscope. Microcracks were formed along the slip planes after plastic deformation had reduced the tensile strength of the metal; these cracks were oriented ⊥ the direction of applied stress. The orientation was changed by continued deformation, and the presence of the cracks resulted in localized stresses which formed a new set of microcracks oriented ∥ the slip planes.

Galvanic Macro-Etch for High-Purity Aluminium. I. S. Servi (*Metal Progress*, 1950, **58**, (5), 732–733; corrigendum, 1951, **59**, (3), 375).—A letter. A brief account is given of macro-etching recrystallized high-purity Al bars by dipping in 15% HCl and coupling them externally with 18 : 8 stainless steel rod dipped in the same soln.—S. R. W.

***Some Metallographic Observations on Aged Aluminium–Copper Alloys.** I. J. Polmear and H. K. Hardy (*J. Inst. Metals*, 1952–53, **81**, (8), 427–431).—The occurrence of Gayler's "light phenomenon" in Al–Cu alloys has been studied in detail. It has been concluded that it is not formed as an integral part of the ageing process, but that it originates as a grain-boundary recrystn. process to relieve strains induced by cold-water quenching and subsequent pptn. on ageing. The effect of very small addn. of In or Sn on the ageing of Al–Cu alloys has been investigated by opt. and electron microscopy. The particles of θ′ are smaller, more numerous, and occur at an earlier stage in the ageing of the ternary alloy. This supports the previous conclusion that the ternary addn. facilitate the nucleation process in the ageing of Al–Cu alloys.—Authors.

***Thermodynamic Calculations of the Nature of the Guinier–Preston Zones in the Aluminium–Copper Alloys.** J. L. Meijering (*Rev. Mét.*, 1952, **49**, (12), 906–910).—The Al–Cu system has many intermetallic phases, but the structure is the same at both ends, viz. f.c.c. M. investigated by thermodynamic analysis whether, if the formation of all other structures is prevented by difficulties of nucleation, there would be a continuous series of solid soln. or a miscibility gap. The calculations showed that *a priori* a segregation of the homogeneous supersaturated alloys might be expected in two disordered f.c.c. phases in these circumstances. Thermodynamically, the formation of Guinier–Preston zones with essentially the same structure as that of the matrix is then possible and perfectly normal. Hardy (*J. Inst. Metals*, 1948–49, **75**, 707; *M.A.*, **17**, 810) designated G.P. [1] and G.P. [2] the two Guinier–Preston zones indicated by the direction of hardening with the temp. at ~150° C. 15 ref.—J. H. W.

***A New Effect of Cold Work on Powder X-Ray Patterns [of Gold–Copper Alloys].** Shoichi Annaka (*J. Chem. Physics*, 1952, **20**, (10), 1662–1663).—A letter. Annealed $AuCu_3$ and AuCu were filed at room temp., and one-half of the filings were annealed at 350° C. Superlattice lines were faintly present in the cold-worked samples, while in AuCu, diffuse lines of a f.c.c. lattice rather than a tetragonal one were obtained. Possible mechanisms of disordering by filing are discussed.—E. O. H.

***Crosses Observed in the Electron-Diffraction Pattern of an Orientated CuAu Film.** Shiro Ogawa and Denjiro Watanabe (*Acta Cryst.*, 1952, **5**, (6), 848–849).—[In English]. A letter. Some of the superlattice spots in the electron-diffraction pattern of a CuAu film, deposited on NaCl and annealed after removal, were in the form of crosses. The interpretation of this in terms of antiphase domains is given.—R. W. C.

***Study of Order–Disorder Transformations in Copper–Platinum Alloys.** Gérard Fournet (*Compt. rend.*, 1952, **235**, (22), 1377–1379).—The order–disorder transformations of the CuPt alloy have been studied by Johansson and Linde (*Ann. Physik*, 1927, [iv], **82**, 449; *J. Inst. Metals* (Abstracts), 1928, **39**, 539, and by L. (*ibid.*, 1937, [v], **30**, 151; *M.A.*, **5**, 110). The change from the ordered to the disordered state occurs without any notable change in the distribution of near neighbours. Since the difference in energy between pairs of like and unlike atoms at a distance r is given by: $W(r) = W_{Cu-Cu}(r) + W_{Pt-Pt}(r) - 2W_{Cu-Pt}(r)$, F. assumed that the difference, $W(r_2)$, ($1{\cdot}665 \times 10^{-13}$ erg) relative to the second neighbours is greater than the term, $W(r_1)$ ($0{\cdot}996 \times 10^{-13}$ erg) corresponding to the near neighbours. On this assumption he found that the experimental results agreed well with the calculated ones. A more detailed note will appear later.—J. H. W.

***Mechanical Twinning of Indium.** J. H. Becker, B. Chalmers, and E. C. Garbow (*Acta Cryst.*, 1952, **5**, (6), 853).—[In English]. A note. In has a f.c.c. structure distorted to a tetragonality of $c/a = 1{\cdot}078$. This can be twinned by deformation. Twins form on {011}, but not on {110} planes. —R. W. C.

***The Crystallographic Relationship Between the Phases γ and ε in the System Iron–Manganese.** J. Gordon Parr (*Acta Cryst.*, 1952, **5**, (6), 842–843).—[In English]. A letter. The orientation relationship between γ and ε phases in powders quenched from the γ field is consistent with the Nishiyama martensite shear mechanism, for a f.c.c. → h.c.p. transformation. $(111)_\gamma \parallel (0001)_\epsilon$ and $(110)_\gamma \parallel (11\bar{2}0)_\epsilon$. This relationship was confirmed from the correspondence of individual spots on certain pairs of adjacent diffraction rings of the two phases. A photomicrograph shows a very fine-scale Widmanstätten-type structure which reinforces the belief that the transformation is martensitic.—R. W. C.

Sigma Phase in Stainless [Steels]: What, When, and Why.—I.–II. George V. Smith (*Iron Age*, 1950, **166**, (22), 63–68; (23), 127–132).—A review of the nature of the σ phase, its occurrence in stainless steels, and its effect on their properties. 17 ref.—N. B. V.

***Crystal Structure of Lanthanum Nitride.** R. A. Young and W. T. Ziegler (*J. Amer. Chem. Soc.*, 1952, **74**, (21), 5251–5253).—Specimens of the approx. compn. LaN were prepared by direct combination of La metal and N, giving loose black powders, the properties of which were investigated. The NaCl-type crystal structure previously reported for LaN was confirmed, and the dissociation pressure was found to be very small at 900° C. No evidence of supraconductivity was observed down to 1·8° K. Powder X-ray-diffraction patterns for LaN and f.c.c. La metal were demonstrably different.—J. R.

Electrolytic Polishing of Nickel. Glen W. Wensch (*Metal Progress*, 1950, **58**, (5), 735–736).—A letter giving a modification of the method used by Hothersall and Hammond (*J. Electrodepositors' Tech. Soc.*, 1940, **16**, 83; *M.A.*, **8**, 74). —S. R. W.

***X-Ray Study of the Structure of Silver Amalgams.** S. Rama Swamy and T. G. Shammana (*Current Sci.* (*India*), 1952, **21**, (1), 7–8).—The crystal structure of Ag amalgam prepared by six different methods has been studied by the X-ray powder technique and, in one case, single crystals of amalgam were examined by the rotation method. The results show that the γ phase appears at 16% Ag, that a β phase does not exist at any stage, and that γ persists even above 35% Ag, along with α. It is suggested that disagreement of the results of earlier workers with these is due to their different methods of prepn. The chem. formula corresponding to γ is being studied.—S. V. R.

Microstructure of Titanium. H. P. Roth (*Metal Progress*, 1951, **59**, (6), 816-B).—A data sheet. The microstructures of extruded and annealed, powder, sponge, and arc-melted Ti under polarized light after etching in Jacquet's electrolyte (100 parts glacial acetic acid and 6 parts perchloric acid; sp. gr. = 1·54), are illustrated.—S. R. W.

***An X-Ray Powder Study of Beta-Uranium.** J. Thewlis (*Acta Cryst.*, 1952, **5**, (6), 790–794).—[In English]. Small differences between the powder diffraction patterns of β-U and the β-phase in the U–1·4 at.-% Cr alloy are presented in detail. These make it highly probable that the two structures differ significantly. Experimental and interpretative difficulties experienced in the study of β-U are recounted. Ito's (*Nature*, 1949, **164**, 755) general method of indexing powder patterns failed completely when applied to the β-U pattern. Values for the parameters, thermal expansion, and Debye temp. are given.—R. W. C.

Metallography. [—I.–III.]. Th. Zürrer (*Pro-Metal*, 1952, **5**, (28), 48–50; (29), 78–79; (30), 96–98).—[In French and German]. [I.—] After a brief review of the history of metallography, from Sorby to the publication, by 1952, of >50,000 papers, the structure of the atom is very briefly discussed. [II.—] The crystallographic systems are described, and the existence of 32 classes and 230 groups is mentioned. The morphology of cubic, tetragonal, and hexagonal crystals is illustrated briefly, and the concept of the space lattice is defined. [III.—] F.c.c. and b.c.c. lattices are described, and the $\alpha \rightarrow \gamma$ transformation in Fe is illustrated by an expansion curve. The lattice parameters of the most important metals are quoted.—K. S.

Metal Interfaces. (——). See col. 668.

Grain-Size Key to Mechanical Properties. H. A. Unckel (*Iron Age*, 1952, **170**, (25), 148–152).—U. discusses how the mech. properties of Al, Cu, and brass vary to different degrees with variation of grain-size. These differences are due to the combined effect of orientation and porosity concentrated along the crystal boundaries. In extruded Al the front ends show cast structure; the grains in the trailing ends are highly deformed owing to special flow in extrusion. Extruded rods of small cross-section show preferred orientation. 10 ref.—J. H. W.

***Electrolytic Production of Extremely Thin [Metallic] Films and Their Application.** I. Epelboin (*Rev. Mét.*, 1952, **49**, (12), 863–874; discussion, 874–875).—To study the properties of metals in very thin section it is necessary to remove relatively large quantities of metal from specimens without altering the macrostructure, and electrolytic removal after electropolishing lends itself admirably to this purpose. E. describes the best conditions for a good polish on specimens that have not been previously mech. or chem. polished. A number of considerations led to the hypothesis that water plays an important part in electropolishing: a very resistant film is formed round the anode, and this film can dissolve in the aq. medium and transport the metallic ions. Since it is difficult to control the water content of the usual perchloric–

acetic anhydride baths, E. used salts, chiefly Mg(ClO$_4$)$_2$ and NaClO$_4$, which are also less explosive. Owing to the extreme final thinness of the specimens, special precautions are required to ensure even removal of metal, but most of the obvious methods are not easily practicable. Various arrangements are described for thinning down wires and strips uniformly. Since electropolishing does not appreciably alter the structure of the surface layers—as opposed to the work-hardening caused by rolling, &c.—E. investigated to what extent the properties, especially the magnetic properties, of very thin films of Ni and an Ni–Fe–Mo alloy of the Mumetal type are those of deep layers and discusses the appn. to which they can be put.—J. H. W.

***Analysis of Polarized Light Reflected from Absorbing Materials at Normal Incidence.** J. Woodrow, B. W. Mott, and H. R. Haines (*Proc. Phys. Soc.*, 1952, [B], **65**, (8), 603–616).—The complete soln. for the intensity of a beam of plane polarized light and normal incidence is given for low-symmetry crystals and for uniaxial crystals. Specimens of Sn, U, Bi, and stibnite (orthorhombic) were examined. With crossed polarizing units, four max./complete rotation were observed, but with the analyser slightly rotated from the crossed position (up to 8°), the number of max. could often be reduced to two, as predicted by W., M., and H.—E. O. H.

***Crystal Growth and Orientation in Deposits Condensed from the Vapour.** D. M. Evans and H. Wilman (*Acta Cryst.*, 1952, **5**, (6), 731–738).—[In English]. When a vapour stream is directed obliquely at a cool flat surface, which need not be cryst., the condensed solid often shows a "one-degree" orientation, i.e. it has a particular lattice plane ‖ substrate surface. Sometimes, however, the orientation is "oblique," i.e. the lattice plane tends to be ⊥ the vapour stream; the tendency is only partial. It is shown that the most important variable affecting the choice is the surface mobility of the condensed atoms on the substrate and deposit. Where this is high, a one-degree orientation results. Lower mobility leads to a fine-scale roughness of the deposit (even if the substrate is perfectly smooth), and surface-tension considerations then cause certain grains to grow and produce the oblique orientation. Metal condensates examined include Zn, Cd, Sb, Bi, and Fe. 51 ref.—R. W. C.

***Filamentary Growths On Metal Surfaces: "Whiskers".** K. G. Compton, A. Mendizza, and S. M. Arnold (*Corrosion*, 1951, **7**, (10), 327–334).—An account of an investigation into the cause and nature of filamentary growths. The effects of the following variables are discussed: (1) humidity, (2) organic materials, (3) film thickness and methods of appn., (4) surface prepn., and (5) chem. and phys. properties. 7 ref. —W. A. M. P.

Some Recent Ideas on Recovery and Recrystallization. J. A. Ramsey (*Australasian Eng.*, 1952, (Nov.), 58–62).—The main features of recovery and recrystn. are discussed, with particular ref. to structural changes occurring before recrystn. 29 ref.—T. A. H.

***Microstructure and Crystal Structure After Cold Working and Fatigue Loading.** Max Hempel, Hans-Rolf Sander, and H. M. Möller (*Stahl u. Eisen*, 1952, **72**, (18), 1076–1086; discussion, 1086–1087).—Ferritic and austenitic steels were deformed in tension at moderate and very rapid rates and then tested under alternating tension and compression with zero mean stress. The fatigue limit was higher the greater the amount of cold work; in ferritic steels very rapid deformation is more effective than moderately slow stretching; in the austenitic steel deformation at moderate rates led to the formation of more martensite than did very rapid extension, and thus yielded the better fatigue properties. In mild steel, cold working followed by fatigue loading at stresses above the fatigue limit for $>2 \times 10^5$ cycles led to the pptn. of cementite in slip planes; ageing tests indicated that the particle size of the precipitated cementite was equivalent to that formed at temp. below 150° C. X-ray examination revealed that diffuse patterns resulting from the preliminary cold working become gradually sharper during fatigue testing above the fatigue limit, i.e. when cementite is precipitated, and continue to become sharper as the stress rises, even to values which result in fracture at $<2 \times 10^5$ cycles, where no cementite pptn. is observed. In the discussion, *E. Houdremont* interprets the results in terms of dislocation theory and "Cottrell clouds" (i.e. C atoms near dislocations). C atoms accumulate near dislocations arranged in slip planes as a result of the cold work; fatigue permits C to diffuse locally, and at a stress near the fatigue limit cementite is "shaken out of the lattice". At higher stresses, self-diffusion and local flow of Fe atoms became possible, and internal stresses are relieved, sharpening the Debye–Scherrer rings. This very localized flow then leads to incipient "brittle" (deformationless) fracture, i.e. the beginning of the fatigue crack. 16 ref. —K. S.

***Role of Deformation Bands in the Plastic Deformation of Metallic Single Crystals.** Hideo Nishimura and Jin-Ichi Takamura (*Tech. Rep. Eng. Research Inst. Kyoto Univ.*, 1952, **2**, (7), 139–169).—[In English]. The tensile deformation of single crystals of Al and Cu was studied experimentally, and the results are used to discuss: (1) the part played by crystallographic orientation in the development of deformation bands, (2) the origin of X-ray asterism, (3) the precise detn. of inhomogeneous crystal rotation within a deformed single crystal, (4) the correlation between the development of deformation bands and crystal twisting during plastic extension, and (5) the stress/strain curves of extended single crystals in various stages of deformation. 27 ref. —J. S. G. T.

***The Mean Shear Stress in an Array of Dislocations and Latent Hardening.** A. N. Stroh (*Proc. Phys. Soc.*, 1953, [B], **66**, (1), 2–6).—Following work of Taylor and Mott, the R.M.S. shear stress on a slip plane is taken as a measure of the work-hardening on that plane. S. calculates the possible work-hardening of a latent glide system in terms of an array of screw or of edge dislocations on an operative system. The results are shown to be independent of the distribution of dislocations. Since the rate of work-hardening in duplex glide is as high as in simple glide, he concludes from his calculation that edge dislocations account for the greater part of the hardening process.—E. O. H.

***Epitaxy in [Thin] Polycrystalline Deposits [of Metals].** O. S. Heavens and L. E. Collins (*J. Phys. Radium*, 1952, **13**, (12), 658–660).—In a previous paper (*ibid.*, 1950, **11**, 355; *M.A.*, **18**, 207) H. found that a considerable increase of adhesion of thin deposits of Al and Ag to a Cr substrate deposited on glass occurs when the thickness of the Cr deposit $> \sim$300–400 Å. It is now shown that the adhesion of Cu, Ag, Au, and Al deposits to a Cr substrate deposited on glass increases, the less the difference between the interatomic distances in the superficial metal and those of Cr. The adhesion of Cr to glass attains a high value when the rate of its deposition $\ngtr$5–10 Å./sec. Electron-diffraction study of the thin deposits suggests an epitaxial mode of growth of the films upon Cr. Interatomic distances in Ag, Au, and Al do not differ by $>1\%$, and their adhesion to Cr is distinctly superior to that of Cu, in which the interatomic distance is less by $\sim 10\%$.—J. S. G. T.

Crystal Growth Ideas and Crystallographers. H. E. Buckley (*Proc. Phys. Soc.*, 1952, [B], **65**, (8), 578–579).—B. criticizes pictures of "growth spirals" on crystal faces on the grounds that these faces often do not grow at all. CdI$_2$, artificially restricted to grow between cover-slips, is taken as an example. Spirals in SiC are considered to be due to "some vortex of vapour" and to arise near the end of crystn.—E. O. H.

***Application of Fluorescence X-Rays to Metallurgical Microradiography.** H. R. Splettstosser and H. E. Seemann (*J. Appl. Physics*, 1952, **23**, (11), 1217–1222).—Homogeneous X-rays may be obtained at little expense by allowing polychromatic X-rays from a target of fairly high atomic number (e.g. Mo) to excite fluorescent radiation from a range of secondary targets. A convenient set-up is described. Microradiographs, in each case for several different radiations, are reproduced of alloys, e.g. 18 : 8 stainless steel, 12% Cr steel, 93·5 : 5·5 : 0·55 : 0·5 Al–Zn–Mg–Cr, and a bronze. With appropriate radiations the contrast is good, even between elements of similar atomic number (e.g. Fe and Cr). Where atomic numbers of constituent elements differ widely, the primary polychromatic X-rays give good results.—R. W. C.

***Effect of Internal Strains on Linear Expansion, X-Ray Lattice Constant, and Density of Crystals.** P. H. Miller, Jr., and B. R. Russell (*J. Appl. Physics*, 1952, **23**, (10), 1163–1169).—Math. The influence of the presence of lattice vacancies or interstitial atoms on the position of diffraction line peaks is assessed, on the hypothesis that all atoms surrounding such a singularity are displaced radially. The result is $\Delta d/d = 2{\cdot}0\ (4\pi A/3N)$, where N is the number of atoms/defect and the radial distortion is given by $\Delta r = Ar(a/r)^3$. The macroscopic dilatation $\Delta l/l$ is just half $\Delta d/d$. Comparison with experimental data for several ionic salts shows that $\Delta d/d$ and $\Delta l/l$ are more nearly equal, and it is shown how this can be explained by an effect of the presence of Schottky defects on $\Delta l/l$ (the simple theory applies only to Frenkel defects). The above expression for $\Delta d/d$ presupposes that *both* vacancies and interstitial atoms give rise to lattice dilatation, which is stated to be true of ionic crystals.—R. W. C.

A Graphical Method for the Determination of the Unit-Cell Dimensions of Non-Cubic Materials from X-Ray Powder Photographs. J. Thewlis (*Acta Cryst.*, 1952, **5**, (6), 849–850).—[In English]. A letter. Often a powder pattern shows few lines with suitable indices or Bragg angles for a direct precision value for the parameters to be obtained. A simple two-stage approximation method applicable to patterns already indexed is here described, and illustrated by an example. The method is applicable to tetragonal, hexagonal, and (sometimes) orthorhombic substances.—R. W. C.

***The Interpretation of Electron-Diffraction Patterns from One-Degree-Orientated Polycrystalline Deposits and Rotated Crystals.** H. Wilman (*Acta Cryst.*, 1952, **5**, (6), 782–789).—[In English]. Several methods, trial-and-error or analytical, applicable to any symmetry, for identifying a fibre axis ⊥ the face of a flat specimen when the electron beam is almost ∥ the face, are presented in detail. Some of the methods are not subject to this restriction.—R. W. C.

***Diffuse Reflection of Neutrons from a Single Crystal.** R. D. Lowde (*Proc. Phys. Soc.*, 1952, [A], **65**, (10), 857–858).—A letter. Neutron-diffraction studies of an Fe single crystal were found to yield diffuse reflections analogous to the observations with X-rays. The intensity of the diffuse reflections is closely proportional to the abs. temp., but is a little more intense than current theory predicts. The difference corresponds to an inelastic scattering cross-section of ~10 millibarns.—E. O. H.

A New Device for Metallic Shadow Casting. Tadatosi Hibi (*Rev. Sci. Instruments*, 1952, **23**, (7), 383–385).—A device for Fe, Cr, Sb, Cu, and Mo shadow-casting is described for which certain advantages are claimed.—E. J.

5 — POWDER METALLURGY

***Aluminium Powder Products Compared.** E. Gregory and N. J. Grant (*Iron Age*, 1952, **170**, (26), 69–73).—Al products made from three grades of sintered Al powder were tested in creep-rupture at temp. from 400° to 900° F. (205°–480° C.) for up to 1000 hr. Materials made from the flake products showed good high-temp. stability. Compared with the usual forged and cast Al alloys, powdered Al products showed extreme gains in rupture life and creep-resistance. In some longitudinal sections, large elongated inclusions were found; these were not frequent, but appeared to be characteristic of powder products. 6 ref.—J. H. W.

***Fabrication of Aluminium Bronze by Powder-Metallurgy Method.** A. Gupta (*Current Sci.* (*India*), 1952, **21**, (2), 39–40).—A method is described of preparing homogeneous compacts of an alloy by heating above the m.p. of one of its constituents. Freshly reduced Cu powder was ball-milled with different proportions of pure Al powder, the mixtures compacted under a pressure of 25 tons/in.^{2}, and the compacts sintered in pure H at 900° C. for 32 hr. The resultant alloys were homogeneous, but showed some porosity because of the low compacting pressure.—S. V. R.

***Properties of Some Hydrogen-Sintered Binary Molybdenum Alloys.** (Bruckart *et al.*). See col. 626.

Improvement of Sintered-Metal Bearings. Theodor Hövel (*Stahl u. Eisen*, 1952, **72**, (19), 1157–1159).—The lubrication properties of porous bearings made of sintered Fe can be improved by leaving a hollow ring in the green pressings; this " oil chamber " is loosely filled with powder before sintering. As a result the porosity in the oil chamber is greatly increased and 200% more oil is absorbed. Self-lubricating bearings may seize up at very high speeds and at low ones where oil is not drawn to the surface. Successful bearings have been made by extending the established practice of adding 2% graphite, which improves porous bearings, to the addn. of considerably more graphite and dispensing with oil absorption altogether. Supporting bushes made of sintered Fe powder can absorb the molten white metal thrown on the surface and thus ensure a perfect bond between bearing surface and bush.—K. S.

A Method of Presentation Facilitating the Comparison of Properties of Cemented Carbides. Carl Ballhausen (*Stahl u. Eisen*, 1952, **72**, (9), 489–492).—The properties of cemented carbides are plotted as ordinates over a rectangular base plane, one side of which represents the ratio of the vol. of WC : TiC, the other the content of Co in % cross-sectional area. It is found that most German commercial products lie on a straight line across the low-Co end of the rectangle. The variation of properties with Co content tends to lie on a straight line if the scale for Co is chosen as the power ⅔ of the vol.-%; this follows from the geometry of close-packed spheres set in a binder; deviations from the straight line are due to the irregular shape of the carbide particles. Experimental values of the compressive strength, modulus of elasticity, Vickers hardness, relative bending strength (absolute values cannot be calculated because the neutral fibre does not lie in the centre of the specimen), coeff. of expansion, and cutting properties (in terms of the cutting speed which wears a point down to a width of 0·15 mm. in 10 min.), are plotted in this three-dimensional model. The cutting properties improve steeply towards the TiC corner.—K. S.

***Joining of Powder Compacts to Solid Metals by Pressing and Sintering.** Wilhelm Hofmann and Ernst Schmalenbach (*Abhandl. Braunschweig. wiss. Ges.*, 1951, **3**, 163–176).—The characteristics of joints between sintered Pb-bronze and solid steel, sintered W–Cu alloy and solid Cu, sintered Ag-graphite and solid Cu, sintered Fe and solid steel, and sintered Fe–Cu alloy and solid steel (and more especially the first three of these combinations), were studied. After being compacted, the various combinations were sintered in H. The influences of grain-size, applied pressure, and sintering temp. upon the shear- and tensile strengths of the joints were studied. Joint strengths of the order of a few kg./mm.^{2} were found, and these could be increased by forging. Long storage of the powdered materials introduced troubles due to superficial oxidation, which caused vol. changes during sintering in H owing to formation of H_2O vapour. In the case of the W–Cu alloy joint with Cu, compression of the sintered joint produced a marked increase of hardness and shear strength. Considerable deposition of graphite occurred at the joint between Ag-graphite and Cu and led to much difficulty in the jointing process, but this was overcome by the use of a layer of Ag powder between the sintered Ag-graphite and the Cu.—J. S. G. T.

Methods for Testing Metal Powders. Rolf Schwalbe (*Stahl u. Eisen*, 1952, **72**, (15), 898–903; discussion, 903–904).—Methods for testing metal powders are reviewed, based on the " Tentative Methods of Testing " issued by the Metal Powder Association. The importance of accurate sampling and of chem. analysis is emphasized. The packing and compression characteristics should be determined according to the American methods, including min. pressure required to give a coherent compact, the corresponding *d*, and the ratio of that *d* to the packing *d*, which serves as an index of the cohesion of the compact and the stability of its shape. Bend tests and abrasion tests are also described. The layout of a suitable data sheet is given, and the characteristics of a number of Fe powders are quoted. Swedish electrolytic Fe powder has the best properties, but it appears that in the U.S.A. powders with low packing *d* are preferred because they give more uniform *d* distributions in simple moulds. 10 ref.—K. S.

Metal Powders in Industry. H. W. Greenwood (*Machinery Lloyd* (*Eur. Edn.*), 1952, 24, (23A), 61–63).—A survey of the chief uses of metal powders, including jointing, protecting, and cutting.—C. P. F.

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6 — CORROSION AND RELATED PHENOMENA

The Chemical Resistance of Aluminium. J. M. Bryan (*Aluminium Courier*, 1951, (17), 9).—B. states that Al is much less liable to corrosion than its position in the electrochem. scale would indicate by reason of the natural oxide film on it. The addn. of Cu and Ni lowers the corrosion-resistance; Zn has little effect, provided Cu is absent, and Mn, Si, Mg, or Mg_2Si produce the most resistant alloys. Up to 0·25% Cr may improve these alloys. Alloys contg. Mg are preferable for alkaline conditions, and those with Mn for acid conditions. Contact with the more noble metals should always be avoided. Porosity in anodic films used as protection can be reduced or eliminated by subsequent sealing treatments.—C. P. F.

The Durability of Aluminium in Horticulture. —— (*Aluminium Courier*, 1952, (18), 9).—Describes the performance of Al in resisting the attack of soils in agricultural districts, manures, fertilizers, animal urine, &c. Al is said to be superior to most common metals.—C. P. F.

***Study of Pyrophoric Metals: The Oxidation of Cerium and Uranium.** Jean Loriers (*Rev. Mét.*, 1952, 49, (11), 801–810; (12), 882–905).—Ce is the base of all industrially pyrophoric alloys, and since U, although not used indust. for this purpose, behaves similarly, L. experimentally investigated the phenomenon of pyrophoricity by studying the oxidation of these metals by means of mass/time curves and by determining the nature and structure of the oxides found, which should not be of a protective nature. He used differential thermal analysis, a sensitive thermobalance, and electron or X-ray diffraction. Pyrophoricity can be conferred on any oxidizable metal. Ce is the most pyrophoric metal, with an ignition temp. of 320° C. in the massive state and a much lower one as powder or filings; it is little affected by the air supply, compn. of the atmosphere, or rate of heating. The transformation $\beta \rightarrow \gamma$ Ce is accompanied by a short-lived but important increase. The rate of oxidation is low at room temp., but increases abruptly between 290° and 300° C. Electron diffraction disclosed a new oxide, Ce_2O_3, of the cubic system with $a = 11{\cdot}26$ Å. The pyrophoricity of U is comparable to that of Ce, in accordance with a newly propounded law of oxidation involving a progressive alteration in the compn. of the oxide layer. In O the rate of oxidation starts to become appreciable at ~150° C. and has two regions of rapid increase at 240° and 350° C. Ignition occurs above one or other of these two temp. according to the form of the specimen. In air, oxidation is less rapid and combustion less violent. Below 150° C. is the region of f.c.c. UO_2; between 150° and 250° C. that of U_3O_7, and above 250° C. that of orthorhombic U_3O_8. 97 ref.—J. H. W.

[Corrosion of] **Tin Bronze Condenser Tubes.** —— (*Tin and Its Uses*, 1952, (26), 6–7).—Results of sedimentation and jet-impingement tests on 10% Sn bronze and conventional condenser-tube alloys indicate that the corrosion-resistance of the bronze is superior. This is confirmed by service tests in a number of installations. Sn bronze is especially suitable for condensers in land power-stations using estuarine cooling water. It is suggested that the use of Sn bronze in this appn. would represent a considerable economy in Ni.
—K. S.

[Corrosion-Resistance of] **Hastelloy Alloy C.** Edward D. Weisert (*Chem. Eng.*, 1952, 59, (6), 297–298, 302, 304–306, 308–310, 312)—The corrosion-resistance of Hastelloy alloy C to over 150 reagents is given, with data on its appn. in chem. plant.—K. C. R.

Uniform Test Procedures Evaluate Corrosion-Resistance of Titanium. G. C. Kiefer (*Iron Age*, 1952, 170, (23), 170–173).—Corrosion tests in various mineral and organic acids, boiling and at room temp., showed that unalloyed Ti 75A is greatly superior in resistance to chem. pitting to most common metals, and in many cases has better corrosion-resistance than stainless steel.—J. H. W.

***The Corrosion of Titanium in Acids: The Rate of Dissolution in Sulphuric, Hydrochloric, Hydrobromic, and Hydroiodic Acids.** M. E. Straumanis and P. C. Chen (*Corrosion*, 1951, 7, (7), 229–237).—The action of H_2SO_4, HCl, HBr, and HI on Ti is discussed, and some experimental work to determine the rates of dissoln. of Ti in these acids is described in detail. The experimental results are given. 15 ref.—W. A. M. P.

***The Dissolution of Metals Over the Temperature Range 25° to −60° C. III.—The Dissolution of Zinc in Hydrochloric Acid–Methyl Alcohol–Water Solutions.** J. Y. Welsh and A. B. Garrett (*J. Phys. Colloid. Chem.*, 1952, 56, (6), 727–730). —Cf. *ibid.*, (4) 449; *M.A.*, 20, 351. Data have been obtained for the rates of depolarized dissoln. of polished and pre-etched Zn in HCl–methanol–water soln. under dynamic conditions. The depolarized dissoln. process appears to be first-order with respect to the acid concentration, with an activation energy of the order of 4000 cal./mole.—K. C. R.

***Effect of Packaging on Corrosion of Zinc-Plated Equipment.** K. G. Compton, S. M. Arnold, and A. Mendizza (*Corrosion*, 1951, 7, (11), 365–372).—Some experimental work to determine the effect of various packing materials and methods on the rate of corrosion of Zn-plated equipment is described, and remedial measures are discussed.—W. A. M. P.

***Dissolving Velocity of Metals in Deuterio-Electrolytes.** J. Brun (*Acta Chem. Scand.*, 1951, 5, 1402–1404; *C. Abs.*, 1953, 47, 34).—[In English]. The rate of soln. of metals in acids and bases in which the H content was partly or wholly replaced by D was measured. The dissolving rate ρ of Ni in 99% DCl was 1·63 ml. H/min./cm.², as compared with 2·59 for HCl. After 2 hr., ρ for Al in D_2SO_4 at 50° C. was 0·7 ml. H/hr., compared with 1·25 for H_2SO_4; and 7·8 for KOD at 25° C., compared with 12·6 for KOH. In 4·5*N*-DCl at 25° C., however, ρ was 22·7 after 0·35 hr., compared with 4·7 after 0·25 hr. in HCl.

Resistance to Attack by Liquid Metals. LeRoy R. Kelman, Walter D. Wilkinson, and Frank L. Yaggee (*Metal Progress*, 1950, 58, (6), 868-B).—A data sheet compiled from "Liquid Metals Handbook," p. 733, 1950: Washington (U.S. Government Printing Office); *M.A.*, 18, 684.—S. R. W.

Stress-Corrosion Cracking in Alkaline Solutions. H. W. Schmidt, P. J. Gegner, G. Heinemann, C. F. Pogacar, and E. H. Wyche (*Corrosion*, 1951, 7, (9), 295–302).—A discussion of the results of an indust. survey on the failure of materials (mainly Fe and alloy steels) in alkaline soln. (mainly NaOH). Corrective measures are also discussed. 5 ref.
—W. A. M. P.

Stress-Corrosion: Its Relation to Other Types of Corrosion. U. R. Evans (*Corrosion*, 1951, 7, (7), 238–244).—Corrosion damage is classified as follows: (1) General (unlocalized) corrosion, (2) pitting, (3) general intergranular corrosion, (4) stress-corrosion, and (5) corrosion-fatigue. Stress-corrosion is discussed in relation to the other types of corrosion, and preventative measures are briefly reviewed. 34 ref.—W. A. M. P.

Effect of Specimen Shape on Corrosion in the Atmosphere. H. R. Copson (*Corrosion*, 1951, 7, (10), 335–338).—C. discusses the effect of specimen shape on corrosion in the atmosphere. 10 ref.—W. A. M. P.

Prevention of Corrosion and Metal Attack in the Steam-Water Cycle of the Steam Power Plant. Frederick G. Straub and Harry D. Ongman (*Corrosion*, 1951, **7**, (9), 312–315).—The presence of dissolved Cu and Fe in the water used in steam power-plant is discussed, and it is considered that the forms in which these metals enter the boiler and their distribution within the boiler may be the explanation of boiler corrosion.—W. A. M. P.

Corrosion Problems in Rural Power Distribution. Orville W. Zastrow (*Corrosion*, 1951, **7**, (9), 308–311).—A brief general review.—W. A. M. P.

Corrosion Problems of Motor Coaches. K. L. Raymond (*Corrosion*, 1951, **7**, (9), 303–307).—W. A. M. P.

†Prevention of Corrosion in Naval Aircraft.—I.–II. N. E. Promisel and G. S. Mustin (*Corrosion*, 1951, **7**, (10), 339–352; (11), 377–388; discussion, 388–389).—A comprehensive review. 18 ref.—W. A. M. P.

Corrosion-Proofing Grumman War Planes. Frank L. Bonem (*Products Finishing*, 1952, **17**, (3), 24–32, 34, 36, 38).—The use of anodizing, conversion coatings, and electrodeposition is described.—G. T. C.

A Modified Method for the Estimation of Corrosion Due to the Free Sulphur and Sulphur Compounds in Oil. Richard A. Patton and Joseph H. Lieblich (*Amer. Soc. Test. Mat. Bull.*, 1952, (186), 59–66).—The A.S.T.M. method of test for free and corrosive S in petroleum products (D130-50T) involves immersing a Cu strip under standard conditions and estimating the thickness of the film formed by visual observation of the interference colours produced. The opt. phenomena involved are examined in detail by P. and L., and a modified procedure for carrying out the test is described. It is suggested that a cathodically-prepared microcryst. matt surface be used for the corrosion test specimens, and that a simple filter type of monochromatic reflectance meter be considered as a substitute for visual examination of the corroded specimens. 7 ref.—P. T. G.

***Adsorption Balance for the Study of the Corrosion of Metals and Alloys by Gases at High Temperatures.** I. A. Makolkin (*Zavod. Lab.*, 1949, **15**, (10), 1209–1212).—[In Russian]. The design and use of an adsorption balance for the investigation of the corrosion of alloys in various gases at temp. in excess of 1000° C. is described. The balance is based on the design of Leontis and Rhines (*Trans. Amer. Inst. Min. Met. Eng.*, 1946, **166**, 265; see *M.A.*, **13**, 350), but is modified so that the sensitive spring operates at room temp. in a separate chamber, and the effect of convection currents from the furnace on the balance is eliminated. Despite this precaution, the spring oscillated when gas was flowing through the furnace, and the flow was stopped whilst readings were taken. The sensitive spring was made of Mo wire, 0·2 mm. in dia., wound into 200 turns of 10-mm.-dia. and, when used in conjunction with a micrometer eyepiece cathetometer, enabled abs. precision and sensitivity 25 and 200 times greater, resp., than those of the L. and R. balance to be obtained.—D. M. P.

7 — PROTECTION

(Other than by Electrodeposition)

Anodized Aluminium: Choice of Alloy and Preparation of the Surface. A. Reuter (*Aluminium*, 1952, **28**, (3), 78–80).—R. discusses briefly the effect of foreign elements on the quality of the oxide film, the choice of alloy for decorative anodizing, and prepn. of the work.—M. A. H.

Anodizing Aluminium with Oxalic Acid. Sakae Tajima (*Products Finishing*, 1952, **17**, (3), 42–56, 58, 60).—The development of the oxalic acid method of anodizing Al is traced, and full details are presented of the soln. required and of the operating characteristics of the bath. Detailed information is presented on the characteristics of the film produced, and comparative data are included for films produced by other methods. 14 ref.—G. T. C.

***Anodic Oxidation Phenomena of Aluminium.** Giampaolo Bolognesi (*Ann. Chim. appl.*, 1949, **39**, 535–541; *C. Abs.*, 1951, **45**, 4151).—Semi-quant. comparisons of protective coatings of Al were carried out by comparing the vol. of H developed electrolytically in 0·1*N*-HCl at 25° ± 0·2° C. with 9 cm.^{2} metal surface which had been anodically oxidized for 20 min. in a 15% H_2SO_4 bath at a c.d. of 0·8 amp./dm.^{2}. The following substances, deposited on the Al, caused an increase of vol. of H formed in the following order: $BaSO_4$ (precipitated from a $Ba(NO_3)_2$ soln.) < urea and formaldehyde $\approx BaCrO_4 < SrCrO_4 < PbCrO_4 \approx$ mixture of Co acetate and Mn acetate $< K_2CrO_4 <$ distilled $H_2O < ZnCrO_4 < BaSO_4$ (precipitated from a $BaCl_2$ soln.) < phenol and formaldehyde.

The Testing of Anodic Coatings. —— (*Metal Ind.*, 1953, **82**, (1), 11–13).—Describes the method of determining the thickness of oxide coatings on Al and its alloys by measuring the breakdown voltage.—J. H. W.

***Some Properties of Aluminium Flake Powder: The Degree of Flakiness.** G. W. Wendon (*Paint Manuf.*, 1952, **22**, (12), 455–459, 472).—Experiments are described which were aimed at determining the degree of flakiness of Al powders produced by different methods. The investigation entailed determining the resp. size distribution of the different pigments and the mean flake thickness of all the size fractions obtained. Certain general conclusions are reached from the data presented. 8 ref.—G. T. C.

Galvanized Sheet and Strip [American Practice]. Wilhelm Brachmann (*Stahl u. Eisen*, 1952, **72**, (14), 820–832; discussion, 832–833).—A report on B.'s visit to U.S.A. galvanizing plants. Continuous strip galvanizing produces a more uniform Zn coat, because thickening of the layer at the edges does not occur. A similar trend towards continuous galvanizing is predicted for Germany, although problems may arise from the need for softer grades for manual working of the sheet, which still plays an important part in Europe. In the prodn. of galvanized sheet and strip the following differences between U.S.A. and German practice are emphasized: the use of a skin pass in preparing the surface for galvanizing; the advantages of a neutralizing wash after pickling; galvanized sheets are treated by galvannealing, which extends the Fe–Zn layer throughout the coating and improves adhesion, phosphatizing, or chromatizing, all of which prepare the surface for painting; the importance of painting in conjunction with galvanizing as a protection against corrosion is more widely understood in the U.S. than Germany; greater care is taken with the surface condition and the grooves of the rolls in the Zn baths. When galvanizing without a flux, Al can be added to the bath, and this greatly improves adhesion. Samples of American galvanized sheet and strip were examined in the Max-Planck Institut für Eisenforschung; the quality of the coat was fairly high, the deposit was uniform; it was thinnest in samples produced in Pb–Zn baths; a few defects were found, such as flux or dross residues. The uses of galvanized sheet and possible competitive materials are briefly discussed.—K. S.

The Hot-Dip Galvanizing of Tubes [in U.S.A.]. Rolf Haarmann (*Stahl u. Eisen*, 1952, **72**, (14), 805–811; discussion, 811–812).—H. compares U.S. with German galvanizing practice. Methods of degreasing, pickling, automatic dipping, removal from the bath, and blowing off of surplus zinc are described, and various ways of heating the zinc baths are discussed. Most galvanized tubing in Germany is treated in plant which must be capable of galvanizing other shapes as well. The few special tube-galvanizing installations in Germany are as efficient as the American ones, from the point of view of output, consumption and recovery of Zn, fuel economy, &c. American plant for galvanizing conduit tubing, 3 m. long, for which there is no German equivalent, is also described.—K. S.

American Methods of Galvanizing Steel Wire and Steel-Wire Products. Peter Hillesheim (*Stahl u. Eisen*, 1952, **72**, (14), 812–819; discussion, 819–820).—American practice is described on the basis of a visit to ten works. In contrast to German practice, Pb bath annealing predominates. After galvanizing excess Zn is always removed with asbestos pads; the Continental practice of introducing Pb into the bath is avoided, because it is believed that even 0·5% Pb interferes with good adhesion of the Zn coating. The wire is usually cooled or quenched as soon as possible after it leaves the Zn bath, in order to suppress the formation of an Fe–Zn alloy layer, which is believed to make crack-free bending and coiling of the wire more difficult. In most other respects, general practice in the two countries is similar. Two electrolytic methods are also discussed. Hexagonal wire netting is usually made from galvanized wire, whereas square mesh netting has to meet such severe specifications that very heavy coatings are applied after weaving. Nails are treated by the flake process, using heated rotary drums. The quality of galvanized steel wire is no better in the U.S.A. than in Germany, but the output is considerably higher.—K. S.

Operations and Maintenance of Pipe-Line Cathodic-Protection Systems. Marshall E. Parker (*Petroleum Eng.*, 1952, **24**, (6), D36, D38–D40, D42, D44).—The need for const. checking and supervision of cathodic protection systems is emphasized. Ref. is made to the use of non-ferrous anodes in such systems.—G. T. C.

Combating Corrosion in a Chemical Plant With Magnesium Anodes. Oliver Osborn (*Corrosion*, 1951, **7**, (1), 2–9).—O. discusses the use of Mg anodes for the cathodic protection of steel and cast Fe structures in a chem. plant.—W. A. M. P.

Dow Conquers Sea-Water Corrosion [by Cathodic Protection]. D. P. Thornton, Jr. (*Petroleum Processing*, 1952, **7**, (11), 1640–1643).—The use of Mg anodes for cathodic protection and also the use of various non-ferrous metals are mentioned.—G. T. C.

Performance of Pretreatment Primers as Revealed by Accelerated Weathering Tests. —— (*Products Finishing*, 1952, **17**, (2), 46, 48, 50, 52, 54, 56).—Brief information is included on the performance of pretreatment primers when used on Al, brass, hot-tinned steel, Zn-plated steel, and Cd-plated steel.—G. T. C.

Review of Pigment Progress. W. M. Morgans (*Paint Manuf.*, 1952, **22**, (12), 447–451).—Brief ref. are made to advances made in the past year with metallic pigments. —G. T. C.

What's Ahead in Corrosion Prevention. Herbert H. Uhlig (*Chem. Eng.*, 1952, **59**, (6), 314).—U. outlines improvements that can be expected in various methods of protection against corrosion.—K. C. R.

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11 — ANALYSIS

Chemical Analysis of Light Metals and Alloys. S. T. Payne (*Light Metals*, 1952, **15**, (168), 105–106; (169), 127–128; (170), 151–152; (171), 189–190).—P. gives a collection of methods recommended for the detn. of Cu, Mg, Si, Fe, Mn, Ni, Zn, Pb, Sn, Ti, and Cr in Al and Al alloys.—A. W. B.

***On the Influence of Silicon on the Intensity of the Aluminium Lines During the Analysis of an Aluminium–Silicon Alloy in an Alternating-Current Arc.** N. K. Rudnevsky (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 642–647).—[In Russian]. R. measured the intensities of Al lines in arc and spark spectra of Al–Si alloys contg. 1·15–34·24% Si. The "abs." intensities of the arc lines Al I 2660·39, 2652·49, and 2567·99 Å. increased with Si content over the whole range, but that of the spark line Al II 2816·18 Å. was const. for Si contents up to 18% and then fell slightly. The relative intensities of homogeneous Al line-pairs were independent of Si content, but for non-homogeneous pairs, such as Al I 2652·49–Al II 2816·18 Å., the relative intensity of the Al I line increased with increasing Si content. Tests of transport of material in the discharge by Sventitsky and Taganov's method (*ibid.*, 1947, [Fiz.], **11**, 3) showed more transport in alloys of higher Si content, while the intensification of this effect on introducing NaCl into the arc suggests that in addn. the temp. of the arc is lowered with higher Si content. With spark spectra the relative intensities of non-homogeneous pairs of Al lines was independent of Si content.—G. B. H.

***Direct Determination of Small Quantities of Arsenic in Copper and Its Alloys: Rationalized Turbidimetric Method.** Charles M. Dozinel and J. P. Vande Briel (*Ing. Chim.*, 1952, **34**, (188), 73–77; *C. Abs.*, 1953, **47**, 67).—For determining $<0·1\%$ As, the colorimetric method of Case was simplified, without loss of accuracy. Dissolve the sample in HCl and H_2O_2, remove excess H_2O_2 by gentle heating, and add a soln. of $NaH_2PO_2.H_2O$. Care must be taken not to precipitate the As, but to keep it in a form that can be determined by the turbidimeter. Details of the procedure are given, with a description of the apparatus, reagents, and the calibration curve.

***Colorimetric Determination of Calcium, Using Murexide.** (Mlle) H. Ostertag, and E. Rinck (*Chim. analyt.*, 1952, **34**, (5), 108–109).—The soln. contg. the Ca in the form of nitrate is adjusted to *p*H 6, and a few drops of an aq. soln. of murexide (the NH_4 salt of purpuric acid) are added; the extinction value is then determined in a Lange photometer, using a yellow filter 500–550 mμ. The effects of the presence of the anions ClO_4^-, CH_3COO^-, Cl^-, SO_4^{--}, and $HCOO^-$, and of the cations Li^+, Na^+, K^+, NH_4^+, Mg^{++}, and Ba^{++}, which diminish the sensitivity of the reaction, are discussed. Sr^{++} must be removed completely, together with Al^{+++} and any heavy-metal ions.—E. N.

[Discussion on a Paper by L. Bertiaux *et al*:] The Oxidation of Chromium by Perchloric Acid. B. Emile Jaboulay (*Chim. analyt.*, 1952, **34**, (6), 131–132).—Cf. *ibid.*, 1951, **33**, 129; *M.A.*, **19**, 732. Experience shows that the oxidation of large amounts ($>$25 mg.) of Cr by $HClO_4$ in the presence of H_3PO_4 is always incomplete, and some Cr is lost as CrO_2Cl_2; the presence of Fe tends to lessen volatilization losses.—E. N.

***Rapid Determination of Copper in Aluminothermic Ferro-Titanium.** Louis Bonnafous (*Chim. analyt.*, 1952, **34**, (8), 176).—The finely ground sample is treated with dil. H_2SO_4 (1 : 2) for $\not>$30 min.; this dissolves the main constituents but leaves any Cu as a spongy mass which is filtered off, washed, dissolved in HNO_3, and determined electrolytically. The method is applicable to material contg. up to 2% Cu, and is more rapid and more accurate than the classical method of pptn. with H_2S.—E. N.

Complete Analysis of Copper Alloys by the Spectrography of Their Solutions. E. V. Rouir and A. M. Vanbokestal. (*Spectrochim. Acta*, 1951, **4**, (5), 330–337).—[In French]. R. and V. first deal with the advantages of using synthetic standard soln. made from spectrographically pure materials. A short survey of methods involving soln. is then given, including the method employed in the test. The upper electrode is graphite, the lower a graphite disc, which dips into the soln., carrying it round to the sparking position; the velocity is altered according to the soln. under test. In order to prevent the suppression of Pb and Sn lines by Cu, the Cu is complexed with NH_4Cl and NH_4NO_3. Working details and details of the soln. are given—nitrates and chlorides are used. Cu, Pb, Zn, Ni, Fe, Al, and Mn are determined, and a mean accuracy of 1·5% is claimed. 12 ref.—L. D. H.

***The Low-Voltage Spark: A Universal Light Source for Emission Spectrum Analysis [of Bronzes and Aluminium Alloys].** G. I. Kibisov (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 623–629).—[In Russian]. K. points out the advantages of N. S. Sventitsky's low-voltage spark source, viz. simplicity, ability to vary excitation conditions over a wide range, and ability to excite the ultra-violet spectra of halogens and other non-metals. By measurement of abs. line intensities in spectra of Cu, he showed that, with a large capacitance (20 μF or more) and a small inductance (~100 microhenries) in the main discharge circuit, the intensities of the lines varied greatly according to the width of the discharge gap and the spark-gap width and capacitance in the H.F. exciting circuit; therefore this type of discharge can easily be adjusted to the optimum conditions for any particular problem. Details are given of the optimum discharge conditions and line-pairs for analysis of bronzes contg. Sn 1·9–6·0, Zn 5–12·4, and Pb 2–5·9%, with probable errors of ±3·9 to ±2·0%, according to the line-pair used. With a discharge current of 6 amp., the Sn content affected the calibration curves for other elements, but with a current of 9 amp. this effect disappeared. Similar data are given for the analysis of Al alloys contg. Mn 0·13–1·5, Mg 0·31–2·55, Cu 10·4–6·99, Fe 0·64–2·17, and Si 0·23–2·04%. The probable errors ranged from ±2·2 to ±4·3%. For these alloys all the usual line-pairs were used, except that K. used Cu 2294·36–Al 2321·57 Å. for Cu and Mg 2779·83–Al 2652·49 Å. for Mg. Data are also given for the detn. of Cr, Mn, Si, and Mo in steels.—G. B. H.

***A New Method for the Separation of Hafnium from Zirconium.** Shizuo Fujiwara (*J. Chem. Soc. Japan* (*Pure Chem. Sect.*), 1949, **70**, 132–133; *C. Abs.*, 1951, **45**, 4163).—A mixture of pyrophosphates of Zr and Hf is dissolved in 1 : 1 HF, and to this soln. 1% NaOH is added. The hydroxides of both elements, contg. a small amount of phosphate, are filtered and treated with conc. H_2SO_4 immediately after washing. A large amount of water is then added, and the soln. thus formed is gently heated in a Pt dish on the water-bath after the addn. of a small amount of HF. By this treatment the greater part of the Hf is precipitated, while Zr remains in soln.

***Spectral Analysis of Small Amounts of Lead and Antimony in Nickel Alloys.** N. V. Buyanov (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 649–652).—[In Russian]. B. gives details of the procedure for the spectrochem. estn. of Pb (0·0005–0·02%) and Sb (0·0015–0·15%) in alloys Kh. N. 80, Kh. N. 60, 25–20, and Zh. –27. A 6·5-amp. A.C. arc source was used, with a medium-dispersion quartz spectrograph, and line intensities were measured by photometric interpolation. The line-pairs used were: Pb 2833·07–Cr 2840·02, Pb 2833·07–Fe 2831·56, and Sb 2598·06–Cr 2603·57 Å. The more intense line Sb 2528·53 Å. was unsuitable, as it coincided with a Si line. The line Fe 2598·37 Å. caused difficulty in estimating Sb in alloys contg. >4–6% Fe. The choice of upper electrode affected the sensitivity of the lines; the Sb lines were stronger with a Cu upper electrode than with one of Al, but the opposite was true of the Pb lines. In view of the difficulty of chem. analysis, the standard specimens used to construct calibration curves for Pb in Kh. N. 80, Kh. N. 60, and 25–20 were analysed by spectrochem. measurements on their soln. For Sb in Kh. N. 80 and Zh. –27 the standards were chem. analysed. Estn. of Sb in 25–20 was impossible because of the high Fe content. —G. B. H.

***The Determination of Small Amounts of Magnesia [Magnesium].** Daniel Bourdon (*Chim. analyt.*, 1952, **34**, (10), 221–222).—The Mg is precipitated as $MgNH_4PO_4$; this is dissolved in a known quantity of standard HCl, the excess of which is back-titrated with standard NaOH, using bromocresol green as indicator for the colour change at *p*H 4·4 the ref. standard for which is a buffered Na citrate–HCl or K hydrogen phthalate soln. contg. a few drops of the indicator. Up to 6 mg. Mg can be determined with an accuracy of +2 to –1%. 12 ref.—E. N.

***The Determination of Magnesium and Aluminium in Titanium Metal.** J. A. Corbett (*Analyst*, 1953, **78**, (922), 20–24).—0·5-g. samples of Ti metal are dissolved in boiling 5*N*-HCl and the soln. is diluted and cooled. Ti is precipitated with cupferron and the precipitate extracted into chloroform. The aq. phase is separated, the *p*H adjusted to 3·5, and the Al cupferrate then extracted into chloroform and finally determined colorimetrically with aluminon. Mg is determined in the normal manner, by pptn. with NH_4 phosphate, in the aq. phase from the Al sepn.—F. M. L.

***The Amperometric Titration of Nickel with Dioximes.** V. M. Peshkova and Z. A. Gallai (*Zhur. Anal. Khim.*, 1952, **7**, (3), 152–157).—[In Russian]. Details are given of the amperometric titration of Ni with three different reagents: (1) *Dimethylglyoxime.*—A background of 0·1*N*-Na acetate is satisfactory in the presence of small quantities of Fe (Fe : Ni < 3 : 1). With Fe : Ni > 3 : 1 2*N*-Na acetate is used, and when the ratio exceeds 10 : 1 the excess Fe must first be removed. (2) *Na dimethylglyoxime.*—Conditions are similar to those in (1), but equilibrium is attained more rapidly. (3) *Dioximecyclohexanedione.*—The titration can be carried out on a background of 0·1*N*-Na acetate + 1 ml. excess NH_4OH in the presence of Al up to Al : Ni = 70 : 1, and of Fe up to Fe : Ni = 20 : 1, beyond which the background should be Na citrate + 1 ml. NH_4OH. The half-wave potential in all cases was $E_{1/2} = -1·40$ V., and the experiments were carried out with a dropping Hg electrode, at a potential of –1·70 V. and in an open vessel. 15 ref. (Abstracted from Fulmer Research Institute Translation No. 38.)—D. M. P.

***A Quantitative Colour Reaction for Metallic Palladium.** R. A. McAllister (*Analyst*, 1953, **78**, (922), 65–67).—A note. To a 0·01*N*-HCl soln. contg. Pd are added 0·5 c.c. 1% gum ghatti soln. and a trace of capryl alcohol. A stream of water-washed coal-gas is passed through a fine capillary into the soln.; CO in the coal-gas precipitates the Pd as metal. 1 c.c. 10% phosphomolybdic acid and 0·5 c.c. conc. HCl are then added and the soln. heated at just below 100° C. for 5 min. The green colour developed is measured absorptiometrically.—F. M. L.

***The Spectrochemistry of Technetium and Promethium.** William F. Meggers (*Spectrochim. Acta*, 1951, **4**, (5), 317–326).—[In English]. The isolation of ^{43}Tc and ^{61}Pm by chain-reacting U piles is described. Both elements were examined spectrographically. A list of 407 Tc lines is given, with arc and spark intensities; the wave-lengths range from 8829·80 to 2298·10 Å. In the case of Pm, 272 lines are given, with wave-lengths from 6811·28 to 2632·00 Å.—L. D. H.

***The Determination of Tin in Bronzes, Particularly Silicon Bronzes.** Jean Besson with R. Budenz (*Chim. analyt.*, 1952, **34**, (8), 163–167).—Various methods for the detn. of Sn—by pptn. as metastannic acid—in commercial bronzes (contg. Cu 85–90, Sn 15–10%, the usual impurities, and with or without Si, Fe + Al 0·2, and Ni 0·4%) are reviewed, and it is shown that the precipitate usually contains traces of adsorbed impurities. Consequently, the Sn in the precipitate can be determined accurately only after sepn. of such impurities. The following scheme is recommended, therefore, for all types of alloys. Treat 1 g. fine turnings with 20 ml. $HClO_4$ (*d*, 1·615), warm, and add 1 ml. HNO_3 (*d*, 1·42); heat to 80°–100° C. to remove nitrous fumes, and any HCl formed during soln., and evaporate to fumes. Cool, add 100 ml. cold H_2O, boil for 2–3 min., cool, and filter off the precipitated SiO_2 and SnO_2, and wash several times with dil. $HClO_4$ (2%) followed by H_2O. Calcine the precipitate and filter-paper in a Pt crucible at 1000° C., cool, and weigh. Grind the precipitate with 2–3 g. NH_4I (freshly prepared), place the mixture in the Pt crucible, and heat for 15 min. at 450°–500° C. to volatilize the Sn as SnI_4. Cool, treat the residue with 2–3 ml. conc. HNO_3, and calcine at 900°–1000° C. to reconvert the residual impurities and any Si—now partly in the form of iodides—to oxides; cool, and reweigh, the loss in wt. corresponding to SnO_2. (If Sb is present, it is volatilized also and must, therefore, be determined separately). Treat the residue with 2 ml. H_2SO_4 (1 : 2) and 2–3 ml. HF (*d*, 1·14), evaporate, and calcine at 1000° C.; cool, and reweigh, the loss in wt. corresponding to SiO_2. The final residue (mostly CuO) is fused with $NaHSO_4$, dissolved in H_2O, and added to the original filtrate; this soln. now contains the remaining constituents and impurities—Cu, Pb, Fe, Al, Ni, and Zn—which are determined by classical methods. 12 ref.—E. N.

***The Opening of White Metals for Analysis by Dry Chlorine.** E. Bishop (*Analyst*, 1953, **78**, (922), 61–62).—A note. Weighed samples of, e.g. white metal or solder are burned in Cl, and $SbCl_5$ and $SnCl_4$ distilled into aq. NaOH and then determined, resp., by reduction and titration with bromate and I. The residue of chlorides of Cu, Pb, Zn, &c., is dissolved in water and analysed by normal methods.—F. M. L.

***Influence of the Presence of Tin in Lead, and Vice Versa, on the Spectrographic Analysis of the System Zinc-Lead-Tin, Rich in Zinc.** Juan Manuel López de Azcona and Luis Jimeno Martin (*Spectrochim. Acta*, 1951, **4**, (4), 265–279).—[In French]. L. de A. and M. describe experiments to investigate the influence of a third element in Zn-base alloys contg. Pb and Sn. In the first series of experiments, soln. were sparked in graphite cups with graphite counter electrodes. In the detn. of Pb, various Zn/Pb line pairs were evaluated, and those chosen in which the influence of Sn was most marked. The results were treated statistically, and after allowing for variations inherent in the method, it was shown that, up to approx. 6 at.-% Sn, an increase of the apparent Pb content, increasing with the Sn content, occurred. Similarly, when Sn was determined in the presence of Pb, very high figures were obtained, up to approx. 10 at.-% Pb. Parallel results were obtained when Zn–Pb–Sn alloys of varying compn. were used as electrodes. It was found possible to apply an empirical correction, which, as was shown statistically, reduced the error to negligible amounts. 26 ref.—L. D. H.

***The Spectral Method of Determining Titanium and Niobium in Metallic Tantalum Powder.** L. S. Lomonosova (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 693–695).—[In Russian]. L. gives details of the procedure for detn. from arc spectra with C electrodes, the Ta powder being contained in a hemispherical crater in the lower electrode. The line-pairs used were: Ti I 3186·45–Ta I 3184·55 and Nb I 3215·95–Ta II 3223·83 Å. A Hilger auto-collimating spectrograph was used, with a slit width of 0·02 mm. Standards for the calibration curves, contg. Ti 0·69–2·06% and Nb 1·05–4·60%, had been analysed chem. The spectrochem. results were reproducible to ±6·5% for Ti and ±6% for Nb. In discussion, L. states that 200–300 mg. of powder were required for spectrochem. analysis.—G. B. H.

Micro-Analysis of the Alkaline Metals. Clément Duval (*Chim. analyt.*, 1952, **34**, (10), 209–221).—A review of methods for the sepn. of small amounts of Li, Na, K, Rb, Cs (and Fr) one from the other, and for their detn. by microscopic, thermo-micrograv., photometric, volumetric, electrolytic, and polarographic means. 32 ref.—E. N.

Electrochromatographic Separations of the Rare Earths. Takuya R. Sato, Herbert Diamond, William P. Norris, and Harold H. Strain (*J. Amer. Chem. Soc.*, 1952, **74**, (23), 6154–6155).—A letter. One-way discontinuous migrations of rare-earth ions obtained on filter-paper after electrolysis are discussed and diagrammatically illustrated. Sepn. of many binary and ternary mixtures, including parents and daughters as Nd–Pm, Ce–Pr, Ba–La, and Sr–Y, was complete, and in conjunction with a transverse flow of solvent (*Anal. Chem.*, 1952, **24**, 776) the method has given continuous sepn. of alkalies and alkaline earths from the rare earths.—J. R.

Sorting of Cemented Carbides with the Use of the Spectroscope. A. A. Dobrinskaya and E. P. Selyaninova (*Zavod. Lab.*, 1949, **15**, (12), 1480–1482).—[In Russian]. The spectral lines used for identification of Ti, Co, Ni, and Mo in cemented carbides are given, and intensity variations corresponding to various compn. are outlined. The procedure is explained and has been found satisfactory for the sorting of cemented carbides.—D. M. P.

***A Modified Method of Sparking Thin Metal Sheets for Spectrographic Analysis, Using the Alternating-Current Condensed-Spark Unit.** R. St. J. Emery (*Spectrochim. Acta*, 1951, **4**, (5), 327–329).—[In English]. The usual methods of holding a sheet sample in contact with a metal block or bar are mentioned; it is pointed out that variation in temp. and elect. conductivity much greater than those found in a more massive sample may still occur. The modified method described is to hold the sample freely between graphite electrodes, and spark on both sides. Two Tufnol platforms are inserted between the arms of the de Gramont stand, the lower of which is attached to the main column, while the upper can slide freely, and by means of a remotely controlled cam can be moved vertically ~5 mm. After aligning the electrodes and placing the sample in the upper platform, the sheet is aligned centrally during pre-sparking. A table of results is given, showing a reduction in S.D. from 6 to 3% when using the modified method.—L. D. H.

***Synthetic Standards for Spectrographic Analysis.** E. S. Kudelya (*Zavod. Lab.*, 1949, **15**, (6), 691–695).—[In Russian]. The prepn. of standard samples of Mn and Al in Fe and the spectrographic calibration curves obtained are described. The samples were prepared by mixing predetermined weights of powder for 4–5 hr. in a flask mounted with its axis at 5° to that of a motor revolving at 50 r.p.m. The mixtures were loaded without agitation into a die lubricated with paraffin wax and compacted under 60 tons/in.^{2}. The compacts were sintered in H at 1000° C. for 2 hr., re-pressed under 70 tons/in.^{2} and resintered at 1000° C. for 3½ hr. (relative $d = 93$–95%). No loss of Mn or Al was detected, the distribution of Mn was uniform along the length of the samples, and the calibration curves obtained were entirely satisfactory. Allowance must be made for the loss of O from the Fe powder during the sintering in H.—D. M. P.

***Diagram of a Universal Generator for Spectral Analysis.** I. E. Klimenko (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 618–622).—[In Russian]. K. describes a circuit for exciting arc or spark spectra, which works on the following basic principle: a 9-kV. L.F. transformer charges a 0·002-μF condenser until it is discharged through a spark-gap and the primary of an H.F. transformer. The H.F. current in the secondary of this transformer produces discharges through a regulating spark-gap in series with the spark-gap used for exciting the spectra. By means of switches which insert or remove condensers and inductances in the L.F. and H.F. circuits, the following modes of excitation can be obtained: (i) A.C. arc; (ii) spark imposed on arc; (iii) controlled spark; (iv) "hot" arc; and (v) "flame type" arc. A circuit diagram is included, and results obtained are briefly described. The change of line intensity with time on sparking for periods up to 12 min. is shown by curves of ΔS against time for 4 line-pairs in the spectrum of a controlled spark between steel and Cu electrodes. For the same electrodes spectrograms, reproduced in the paper, were taken with increasingly drastic excitation, from arc (1) to condensed spark (21). The intensity of each line rose to a max. and then decreased as the violence of the excitation was increased. The higher the excitation energy of a line, the more violent was the excitation needed for max. intensity.—G. B. H.

***Some Investigations on Transport of Material in the Discharge During Spectral Analysis.** K. I. Taganov (*Izvest. Akad. Nauk S.S.S.R.*, 1950, [Fiz.], **14**, (5), 634–640).—[In Russian]. T. describes a method of investigating this process, in which an arc or spark discharge is maintained between the electrode under investigation and another electrode which resists erosion by the discharge (Cu, W, or Mo). The erosion-resistant electrode is then subjected to spectrochem. analysis to determine the amount of material transported. Study of the relative rates of transport to the anode and cathode in a D.C. arc (4 amp. for 10 sec.) and D.C. low-voltage spark (0·4 amp. for 30 sec.) showed that erosion-resistant metals (Al, W, Cu, Mo) are preferentially transported to the anode, while others (Cd, Ni, Pb, Sb, Zn) are preferentially transported to the cathode. These relations do not hold for transport of the constituents in alloy electrodes. The transport of Zn to a Cu electrode in an A.C. arc was studied with an arc between a Zn block and a Cu disc rotated at 1500 r.p.m. by a synchronous motor. The amount of Zn transported to the disc was found to vary with the phase of the supply voltage in the same way as the arc current. T. considers theoretically the effects of diffusion, migration of positive ions, and directed emission from the cathode, stating that diffusion conditions should become const. in less than one half-cycle of an A.C. arc, that migration of positive ions plays a relatively small part, and that emission from the cathode is usually a controlling factor. Comparison of " ageing " curves for Si line intensities in the triggered arc spectrum of a white cast Fe with measurements of Si transport to the auxiliary electrode, shows that the " ageing " process is connected with the saturation of the auxiliary electrode with transported material. T. describes an apparatus for sampling specimens outside the laboratory by transporting material from them to an electrode in a spark discharge.—G. B. H.

***Spectrophotometry in Chemical Analysis.** E. Geffroy (*Chim. analyt.*, 1952, **34**, (6), 119–125).—The importance of using monochromatic light in photometry is discussed. The design, construction, and appn. (including the detn. of Ti and V) of two high-precision photometers—one for the visible and the other for the infra-red part of the spectrum—are described and illustrated.—E. N.

Some Aspects of Spectrophotometry: Colorimetry Applied to the Chemical Analysis of Technical Products. M. Jean (*Chim. analyt.*, 1952, **34**, (10), 226–230; (11), 250–256).—A review, dealing with: (1) the basic principles of abs. colorimetry, and (2) the appn. of photometric methods, in conjunction with organic reagents, to the detn. of impurities—and, where appropriate, constituents—in electroplating baths, steels, and Al, Mg, Ni, Pb, Sn, and Zn, and their alloys; the analysis of refractories is also noted. 48 ref. —E. N.

An Analytical Comparison Scheme. E. Scheuer and F. H. Smith (*Light Metals*, 1952, **15**, (168), 80–82).—Describes a system developed by the ALAR Analytical Committee for checking the accuracy of routine analytical methods in use by member firms. Each laboratory supplies the others with samples from ten different melts and from a standard in duplicate. The twelve samples are analysed together and with other samples under routine conditions. Variations in results are assessed statistically to check whether investigation of the analytical method is needed.—A. W. B.

12—LABORATORY APPARATUS, INSTRUMENTS, &c.

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

***A Light-Profile Microscope for Surface Studies.** S. Tolansky (*Z. Elektrochem.*, 1952, **56**, (4), 263–267).—[In English]. Paper presented at a Symposium held by the Deutsche Bunsengesellschaft at Berlin, Jan. 1952. A new microscope technique, employing a development of the Schmaltz light-cut method (" Techn. Oberflächenkunde ", Berlin, **1936**; see also Räntsch, *Zeiss Nachr.*, 1945, **5**, 189), in which an image of a slit is projected on the surface being studied, at an angle of incidence of 45° by a microscope objective, and the specularly-reflected beam is viewed by a separate microscope; thereby a profile in depth is converted into a line pattern *in extenso*; the limit of magnification is $\ngtr 400$. Difficulties of the method are discussed, and it is shown how the problems of obtaining high magnification and high resoln. can be overcome by using any good met. microscope, and (1) placing a slit close to the iris diaphragm, and (2) utilizing the full effective aperture of the objective, by the use of a small metal-tongue sector reflector in place of the 45° glass reflector. Monochromatic illumination must be used. Profile magnification up to 2000 and high resolution of the order of 0·25 λ are obtained. The technique gives topography over a complete area with a single photograph, and is easy to use. Calibration is effected interferometrically. Illustrations are given of results obtained by the study of a diamond-turned concave cylindrical soft-metal bearing, and of the improvement effected in the photographs by using an opaque fine line on a bright field in place of a slit before the iris, or by replacing the single opaque line by a coarse graticule of lines. Illustrations are also given of the profiles of crystal faces, e.g. (1) a " trigon " growth on diamond; (2) the onset of spiral structure on SiC out of an irregular mass; and (3) a tetragonal Sn crystallite on a single Sn crystal produced by casting. 8 ref.—J. S. G. T.

***A New Photographic Method of Measuring the Dispersion of the Optical Constants of Metals.** J. Bor (*Proc. Phys. Soc.*, 1952, [B], **65**, (10), 753–758).—Plane polarized light at azimuth 45° is reflected from a plane metallic surface, and converted into elliptically polarized light. In this apparatus, the obliquity of the ellipse is converted into intensity differences, which are recorded on a photographic plate. At the same time, different wave-lengths may be dispersed along the photographic plate, giving a quick method of studying the opt. const. of metals over a wave-length range of 0·45–0·6 μ.—E. O. H.

A Torsion Pendulum of Improved Design for Measuring Damping Capacity. James W. Jensen (*Rev. Sci. Instruments*, 1952, **23**, (8), 397–401).—The construction of an improved torsion pendulum and an opt. recording system are described in detail. It was possible to obtain 6–10 results/hr. on high-damping Mn–Cu specimens.—E. J.

Instrumentation [Bonded Wire-Type Strain Gauges]. Ralph H. Munch (*Indust. and Eng. Chem.*, 1952, **44**, (7), 65A–66A, 68A). A brief survey of the operation and appn. of bonded wire-type strain gauges to the measurement of pressure, load, and torque.—D. P. M.

A Metals Comparator for the Inspection and Classification of Metals. B. M. Smith (*Non-Destructive Testing*, 1952, **11**, (2), 41–46).—The instrument is basically an impedance comparator, consisting of a balancing circuit, an oscillator (50, 250, 500, 2500, 4000 and 10,000 c./s.), and an indicating instrument, with a test coil for small parts and a test head for large parts. It is applicable to both magnetic and non-magnetic materials. For the latter, the higher frequencies are used, the impedance of the test coil being changed by the eddy current produced in the specimen so that the higher the resistivity of the material, the more difficult it is to separate. For each job, the unknown is checked against a specimen of known characteristics. The instrument is recommended for separating mixed metals, control of hardness, indicating case depth and plating thickness, control of quality and uniformity, and surface fault detection. Examples of its use and limitations are given. S. stresses: (1) misleading results may be obtained if the specimen differs from the ref. in more than one characteristic, (2) two non-magnetic materials of the same resistivity cannot be separated, and (3) the specimens compared must be in the same phys. form and size.—L. M.

Various Penetrameter Types and Their Limitation. Norman C. Miller and Gerold H. Tenney (*Non-Destructive Testing*, 1952, **11**, (2), 28–32).—The theoretical aspects of penetrameters for assessing radiographic sensitivity are reviewed and experiments on steel are described, involving the use of: (1) the standard flat strip penetrameter with drilled holes; (2) wires; (3) steel balls; (4) cylindrical holes drilled in the specimen with their axes normal to the beam; (5) similar holes with their axes co-incident with the beam direction; (6) ball-type cavities, at various object–film distances, using 1-, 1·1-, and 22-MeV. X-rays and 1·3-MeV. radiation (Co^{60}). Generally the sensitivity *figure* for the flat strip penetrameter is greater than that for wires and horizontal cylindrical cavities, and less than that for balls and spherical cavities. It is concluded that no one penetrameter can represent all kinds of cavities and that the strip type is the best generally. —L. M.

A Sample Cutter for Non-Ferrous Metals. I. S. Solet (*Rev. Sci. Instruments*, 1952, **23**, (10), 566).—The prepn. of small metal samples for analysis using a conventional pencil sharpener driven by an elect. motor is described.—E. J.

Metallic Vacuum-Tight Gasket [Made of Indium]. Arthur W. Knudsen (*Rev. Sci. Instruments*, 1952, **23**, (10), 566–567).—

When non-metallic materials cannot be used in vacuum work, the use of In metal as sheet and wire is proposed, and various techniques are described.—E. J.

Electromagnetic Pump for Liquid Metals. A. H. Barnes, F. A. Smith, and G. K. Whitham (*U.S. Atomic Energy Commission Publ.*, 1952, (**AECD-3431**), 20 pp.).—The construction and performance up to 325° C. of an A.C. linear induction electromagnetic pump for liq. metals are described, and considerations are given to the problems involved in operation using liq. Na or Na–K alloys.—B. W. M.

***Aluminium Monochromator with Double Curvature for High-Intensity X-Ray Powder Photographs.** G. Hägg and N. Karlsson (*Acta Cryst.*, 1952, **5**, (6), 728–730).—[In English]. A point-focusing Al monochromator was made by a variant of Tiedema's (*ibid.*, 1949, **2**, 261; *M.A.*, **17**, 344) technique of growth into a preselected orientation; full details are given. The crystal is grown as part of a cylinder of radius $2R$, and afterwards is bent into a double curvature (of major radius R). Exposures with a Guinier-type focusing camera are ∼16 shorter than when a singly curved quartz monochromator is used, but the resoln. is slightly inferior.
—R. W. C.

A New Evaluating Instrument for Spectrochemical Analysis. H. Bückert (*Spectrochim. Acta*, 1952, **4**, (6), 476–481).—[In German]. This multi-purpose instrument is a spectrum comparator. For use as a microphotometer an electron multiplier cell is incorporated. The instrument can also be used as a measuring micrometer; distances of up to 50 mm. can be measured with an accuracy of ±0·01 mm. Full constructional details are given, with details of each type of operation.—L. D. H.

A New Arcing Apparatus [for Spectrochemical Analysis]. H. Bückert (*Spectrochim. Acta*, 1952, **4**, (6), 525–532).—[In German]. The advantages and necessary conditions for accurate work with an arc are outlined, and an apparatus is described and illustrated, for the generation of an A.C. intermittent arc, triggered by a spark. The apparatus can be changed, by operation of a switch, to A.C. or D.C. continuous arc. For the operation of the intermittent arc, 20 different settings are possible. Constructional and operating details are given.—L. D. H.

An Apparatus for Quantitative Spectrographic Analysis in Controlled Atmospheres. R. St. J. Emery (*Spectrochim. Acta*, 1952, **4**, (6), 513–515).—[In English]. A gas-tight glass chamber is described, which can be used for sparking or arcing in controlled atmospheres at normal or reduced pressures. Examples of its appn. in ferrous metallurgy are quoted.—L. D. H.

Simple Apparatus for Flame Spectrography. J. Monvoisin and R. Mavrodineanu (*Spectrochim. Acta*, 1951, **4**, (5), 396–399).—[In French]. The appn. described for this apparatus is the analysis of Fe alloys, but it can equally well be used for non-ferrous materials. Particles from the sparking of the sample are introduced into the flame of an air–acetylene burner, designed to give a reproducible flame. The spectra produced are simple, and a medium-dispersion Hilger spectrograph gives sufficient sepn. for ferrous as well as non-ferrous materials. Working details are given, but the degree of accuracy of the method is not stated.—L. D. H.

***On a Method of Obtaining Alloys of Varying Concentration.** V. S. Kogan and B. Ya. Pines (*Zhur. Tekhn. Fiziki*, 1948, **18**, (3), 377–382).—[In Russian]. The prepn. of a thin film of alloy of varying compn. for X-ray examination, &c., is described. The components are evaporated from conical Al_2O_3 crucibles and deposited on a flat plate which is of mica and can be heated to promote adequate grain-size in the deposit. The compn. of a multi-component film can be calculated from the quantity of components evaporated and the distribution of thickness, y, about the intersection of the axis of the cone and the plate $y = 1\Big/\left(1 + \frac{x^2}{R^2}\right)^{3/2}$, where R is the distance from the apex of the cone to the plate placed ⊥ the axis of the cone. Perforated screens may be placed in front of the condenser plate to isolate islands of various compn. (Abstracted from Fulmer Research Institute Translation No. 35.)—D. M. P.

13—PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

***Fatigue and Static Tests of Flush-Riveted Joints [in Aluminium Alloys].** D. M. Howard and F. C. Smith ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1952, (2709), 38 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 463).—In regard to the design of flush-riveted Al-alloy aircraft-joint details under repeated load, extensive tests showed superiority in both fatigue and static strength for the dimpling procedure as compared with machine countersinking. Considering joints using machine countersunk holes, comparison involving fatigue strength showed Alclad sheets to be better than bare sheets and lap joints better than butt joints. In both of the foregoing comparisons, no difference was noted with regard to static strength. Materials studied included 24S-T3 and 75S-T6, both bare and Alclad, 0·032 in. thick, and 76S-T6 Alclad 0·064 in. thick. No satisfactory single relation between static properties and fatigue life covering the four materials could be found.

***Fatigue Tests on Typical Two-Spar Light-Alloy Structures (Meteor IV Tailplanes) Under Reversed Loading.** K. D. Raithby (*Aeronaut. Research Council Curr. Paper*, 1952, (88), 13 pp.; *Appl. Mechanics Rev.*, 1952, **5**, 513).—Results are given for fatigue tests on six Meteor IV tailplanes, vibrated in flexure under simple reversed loading, ranging from ±10 to ±30% of the static failing load. Corresponding endurances varied from $\sim 5 \times 10^6$ to $0{\cdot}06 \times 10^6$ cycles for complete failure of a spar boom. Skin-cracking and rivet failures occurred at an appreciably lower number of cycles. In each test the alternating load was kept const., although progressive skin cracking caused a considerable change in alternating stress at the section where eventual failure occurred. The test results are discussed, and the endurance curves obtained are compared with data from reversed-axial-load fatigue tests of 14S-T alloy.

Testing the Hardness and Strength of Metals. —— (*Pract. Eng.*, 1952, **26**, (665), 630–632).—A general account.
—E. J.

***The Influence of the Coefficient of Work-Hardening of Metals Upon the Profile of the Mound Surrounding the Impression Obtained in Hardness Testing.** (Yakutovich, Vandyshev, and Surikova). See col. 616.

***The Determination of the Elastic Constants of Metals by the Ultrasonic Pulse Technique.** Myron B. Reynolds (*Amer. Soc. Metals Preprint*, 1952, (28), 15 pp.).—The theory of the ultrasonic pulse method of determining elastic const. is described, and an account is given of the apparatus and technique used in the present investigation. The results of detn. of Young's modulus, the shear modulus, and Poisson's ratio for a number of metals are tabulated below. The

Metal	Young's Modulus, dynes/cm.^{2} × 10^{11}	Poisson's Ratio	Shear Modulus, dynes/cm.^{2} × 10^{11}
α-Brass . .	9·35	0·36	3·45
Be . . .	30·9	0·05	14·7
Zr . . .	8·84–9·57	0·33–0·35	3·28–3·61
Nb . . .	10·4	0·38	3·75
Ti . . .	10·6	0·34	3·98
U . . .	17·6	0·25	7·03
Th . . .	7·28	0·30	2·80
Stainless steel .	19·6	0·30	7·57
V . . .	12·7	0·36	4·66

above elastic const. are estimated to be correct to within ± 1%, except in the case of Nb, Ti, and V, where the error may be as high as ±4%. Comparison with the results of conventional detn. shows that the ultrasonic method tends to give results which are slightly lower. The detn. carried out on Nb, Ti, Th, and V are the first complete detn. to have been made on these metals. 9 ref.—R. W. R.

***The Determination of the Elastic Constants of Solids by the [Ultrasonic] Pulse Method.** — Krishnaji (*Proc. Nat. Inst. Sci. India*, 1950, **16**, 227–234; *C. Abs.*, 1951, **45**, 4110).—A thin tabular sample of the solid was placed in water, and the energies of the reflected longitudinal sound wave in the water, the refracted longitudinal wave in the solid, and the refracted shear wave in the solid were measured with an oscillograph as a function of the angle of incidence. Materials investigated included Al, Fe, brass, and stone.

The Applicability of Technological Methods of Measuring Internal Stresses in Metals. Hans Bühler (*Stahl u. Eisen*, 1952, **72**, (16), 947–952).—The accurate detn. of internal stresses involves the destruction of the specimen; with shapes that can be turned, a length equal to twice the dia. is suitable as the specimen. Various suggestions for cutting rings from the surface or the end of the component, or for trepanning cores and measuring their dimensions, as well as the detn. of surface stresses by X-ray back-reflection methods, are rejected because the stress distribution is so variable that it is impossible to estimate the stress in one area from detn. elsewhere. Cutting a series of rings throughout the section is an approximation to the complete detn. of the stress distribution by gradual turning or boring. The use of strain gauges greatly facilitates continuous detn. of stresses and it may be possible to cut the specimen into two equal lengths and turn one, bore the other to the same point in the section; in this way the continuous detn. is complete and there is no need for interpolation. 29 ref.—K. S.

Measurement of Residual Stresses. S. Sjöström and R. Gunnert (*Industritidn. Norden*, 1951, **79**, (23), 303–305; *Appl. Mechanics Rev.*, 1952, **5**, 460).—[In Swedish]. S. and G. discuss different methods for measuring residual stresses in heat-treated or welded parts. Thin parallel layers of an initially flat test-specimen can be successively polished away, and the ensuing change of curvature measured. On the assumption that the residual stresses are const. in surfaces ‖ the polished one, those stresses may be computed. This method, due to Stäblein and others, is modified by Sjöström, who measures the strain with a strain gauge on the surface ‖ the polished one instead of measuring the curvature. To find residual stresses in a welded plate at some given point, G. measures the distance between eight small conical cavities at the surface of the plate on a circle of ~9 mm. dia. Then, a concentric annular groove is drilled outside the cavities to a depth of at least 6 mm. The material between the cavities is thus released of stresses, and by measuring the new distances, the initial residual stresses can be found.

Are You Overlooking Ferromagnetic Analysis? J. K. Stanley (*Steel*, 1950, **127**, (14), 88–92).—The use of magnetic analysis in non-destructive testing, stress analysis, and structural studies of ferromagnetic materials is discussed. Induction methods, identity or comparison tests, coating-thickness measurements, stress gauges and methods of determining stress, and studies involving equilibrium diagrams, transformations, &c., are considered briefly.—S. R. W.

Inspection and Electronics. James R. Cornelius (*Coventry Eng. Soc. J.*, 1949, **30**, 3–14).—The requirements of the inspector are indicated, and the use of the electronic comparator in checking dimensions, materials (compn., finish, and defects), and stress and strain is described.—S. R. W.

Detection of Defects by the Luminescent Method in the Metal Industry. M. M. Laushkina and F. I. Rogov (*Izvest. Akad. Nauk S.S.S.R.*, 1949, [Fiz.], **13**, (2), 251–253).—[In Russian]. Appn. of the fluorescent crack-detection method are discussed, and it is noted that the intensity of fluorescence is related to the depth of the defect. The method will not detect large defects normally visible to the naked eye, nor will it work in the presence of slag or corrosion products, which prevent the penetration of the fluorescent liquid into the defects.—D. M. P.

Fluorescent Ink as an Inspection Tool. —— (*Light Metals*, 1952, **15**, (172), 234–235).—Describes the use of this ink in crack detection on Al and Mg alloy castings.—A. W. B.

The Application of Ultrasonics to Casting Inspection. P. Bastien ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 419–422).—[In French]. B. describes briefly the appn. of ultrasonics to the inspection of castings, for the detection of defects. The principal sources of error which may influence the technique are discussed. The general problem of making the defects visible is surveyed briefly.—G. J. S.

The Application of X-Rays Inspection Method to Castings. J. Oosting and B. Schuil ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 475–490).—[In French]. O. and S. give some results obtained in the fabrication of pressure-tight castings with the use of the " atmospheric " riser, the effect of which was controlled by X-ray examination, and then review the possibilities of using the X-ray technique in the foundry.
—G. J. S.

The Present Position of X- and γ-Radiography for the Testing of Materials. Rudolf Berthold, Otto Vaupel, and Norbert von Wetterneck (*Stahl u. Eisen*, 1952, **72**, (9), 492–499; discussion, 499–500).—Recent modifications in the circuits of X-ray equipment have increased by ~15% the max. thickness of steel that can be tested economically. The use of fine-grained film and Pb foil intensifiers has greatly improved the resoln. of images obtained by X-rays or by γ-radiation, but this involves some increase in exposure time or corresponding reduction in economically permissible thickness. Exposure graphs for Co^{60} and Ir^{192} γ-sources are plotted. Thicknesses below 10 mm. are best studied by X-rays, for thicknesses between 10 and 50 mm. Ir^{192} radiation gives the most sensitive detection of defects, and Co^{60} should be used for steel thicknesses above ~45 mm. Prolonged exposure can be compensated by suitable organization of the routine. Safety precautions for the handling of γ-ray sources are discussed in detail.—K. S.

Modern Techniques in High-Voltage Radiography. E. Alfred Burrill (*Non-Destructive Testing*, 1952, **11**, (2), 23–27).—The relative merits of 250-kV., 2- and 20-MeV. X-ray units, and of Co^{60} sources, are reviewed in terms of cost for: (*a*) operating, (*b*) each exposure for 2, 6, 10, and 12 in. of steel, and (*c*) per ft.^{2} of film exposed. The intensity distribution in each beam is described. The assessment of the relative sensitivity obtainable depends on the product of: (1) film and observer factors, (2) radiation factors, and (3) unsharpness factors; these functions are detailed in a formula given for calculating sensitivity which involves both contrast and definition. Radiographic sensitivities obtained with 2- and 20-MeV. X-rays cannot usually be compared favourably with those obtained with 250-kV. X-rays. Considerations of the absorption coeff. for steel, the scattered radiation produced, the cost/ft.^{2} of film exposed, and penetration show that the 20-MeV. set has, in general, no marked advantage over the 2-MeV. set in practice for indust. radiography.—L. M.

Iridium 192 in Industrial Radiography. J. V. Rigbey and C. F. Baxter (*Non-Destructive Testing*, 1952, **11**, (2), 34–40).—The radiographic appn. and usage of Ir^{192} are described and compared with those of 220-kV. X-rays. Most of the γ radiographs were made with Ca tungstate intensifying screens, and the films were developed for 9 min. at 68° F. (20° C.). For an Al casting Pb intensifying screens were used. It is concluded that an Ir^{192} source: (1) is generally less cumbersome in use than an X-ray unit, (2) gives a lower contrast and therefore records a greater specimen thickness than do 220-kV. X-rays, making it very suitable for ferrous and non-ferrous heavy metal specimens, (3) gives inferior results to 220-kV. X-rays for Al castings, and (4) gives less scatter in the image, thus eliminating the need for masking the specimen.—L. M.

Gamma Radiography in the Foundry. R. J. Hart (*Inst. Brit. Found. Preprint*, 1952, (1040), 6 pp.; and *Found. Trade J.*, 1952, **93**, (1890), 583–588).—The advantages of radiography involving the use of radioactive isotopes Co^{60},

Ta^{182}, and Ir^{192} for the routine inspection of castings are discussed.—G. J. S.

Measuring Thicknesses up to 10 mm. by Means of X-Rays and Counting Tubes. Hans Werner Fritze (*Stahl u. Eisen*, 1952, 72, (16), 943–945).—X-rays are used for measuring the thicknesses in the centre of hot and moving sheet up to 10 mm. thick, whereas micrometer readings are confined to edges. American instruments using a fluorescent screen and a photo-cell are too elaborate and expensive, but satisfactory results can be achieved by the use of counting tubes. X-rays are filtered out in two directions at right angles; one beam passes through the sheet to be measured, the other through a ref. sheet, each on to a separate counting tube. Hot metal is less dense than cold, and gives thinner readings (in contrast to micrometer readings which are thicker hot than cold); this is compensated by using a slightly thinner ref. sheet. Counters tend to age and must be regenerated after 1000 hours' use.—K. S.

Measuring Thicknesses of 10–100 mm. by Means of Gamma-Rays and Counting Tubes. Wolfgang Kolb (*Stahl u. Eisen*, 1952, 72, (16), 945).—Stationary plates are measured by placing the γ-source on one surface and moving the counter over the opposite surface until the max. reading is reached. Continuous measurement of moving plate in the rolling mill is being developed. In order to eliminate the influence of temp. and mains voltage variations, a differential method is employed, using a measuring counter and a compensating counter which has a separate but weaker γ source. —K. S.

Measurement of Thickness by Means of Radiation Methods: Measuring Thicknesses of Less Than 1 mm. by Means of Beta Rays. Adolf Trost (*Stahl u. Eisen*, 1952, 72, (16), 941–943).—Radiation methods avoid direct contact with the sheet, and can therefore be used on hot sheets during rolling. The intensity of radiation which is absorbed by the sheet is a measure of its thickness; the energy actually penetrating can be measured quite accurately ($\pm 1\%$); the hardness of the radiation to be used depends on the thickness of the sheet. For thin foil, β rays are suitable. Possible sources of β rays are Tl^{204}, Sr^{90}, Ru^{106}, and Ra; the last is more expensive but more convenient because the source is smaller and its half-life longer, so that it need not be renewed and the equipment need not be re-calibrated so frequently. The intensity of radiation is measured by means of an ionization chamber. A differential method, using two instruments and a standard ref. foil, gives more accurate readings because various deviations originating from variations in air *d* or the circuit resistance can be compensated.—K. S.

Measuring the Thickness of Non-Ferromagnetic Layers on Non-Ferromagnetic Bases by Means of Beta-Rays and Counting Tubes. Rudolf Berthold (*Stahl u. Eisen*, 1952, 72, (16), 945–947).—The hardness and intensity of the X-rays generated at an anode depend on the at. no. of the metal employed. Electrons from a beta-isotope are directed on to a metal surface from which they are reflected into a counting tube which is screened against direct beta-radiation; if the metal surface carries a layer thin enough not to absorb all the electrons received, the intensity and hardness of the X-rays measured by the counter will depend on the difference in at. no. between the coating and the base, and on the thickness of the coating. Using Tl^{204} it is possible to measure thicknesses of Cr electrodeposits on Al up to 10 μ with an accuracy of 10%.—K. S.

Portable X-Ray Unit Has 250,000-Volt Capacity. —— (*Foundry*, 1952, 80, (11), 280–281).—A description is given of a portable X-ray unit which is less than half the size of a conventional unit of the same power. The appn. of the unit to the inspection of castings is described.—R. W. R.

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15 — FOUNDRY PRACTICE AND APPLIANCES

***Study of Microporosity in Aluminium Alloy Casting.** Marcel Bardot (*Fonderie*, 1952, (83), 3207–3226).—B. briefly describes the various types of porosity occurring in Al alloy castings, for the reduction of which various workers have proposed various remedies. He carried out a series of experiments on Al alloys A-S4G, A-S10G, and A-U5GT to investigate this defect. He studied the effect on the formation of micropiping of the pouring temp., the temp. of the mould, the grain-size, the extent of the degassing of the metal, and the reheating and the height of the feeder-heads; he determined the effect of micropiping on the mech. properties and defined the best conditions for confining this defect to a min. Oxidation favours the concentration of pinholes in thick sections with coarse grains. In perfectly degassed metal, the micropores are very small and localized. Grain-size is important; the length and number of the pores increases with grain-size. The pouring of well-gassed metal into wet sand moulds seems to reintroduce gas into the casting, and this tends to increase the vol. of the pores. For gassy metal the effect of the mould is negligible. Neither the height nor reheating of the feeder-head has any effect on microporosity. This defect greatly lowers the U.T.S. of Al–Si alloys, but has little effect on their Y.P. or elongation; the Y.P. of the other two alloys closely follows the degree of porosity. The effect on the fatigue limit was not conclusive. The recommendations for reducing microporosity follow directly from the results of these experiments. 12 ref.—J. H. W.

Gas Removal from Molten Aluminium Alloys. A. W. Brace (*Inst. Brit. Found. Preprint*, 1952, (1034), 9 pp.; and *Found. Trade J.*, 1952, 93, (1870), 3–11; discussion, (1882), 359–361).—Past work on the problem of the removal of dissolved gas from molten Al alloys is reviewed, and a survey made of the mechanism of gas absorption and removal, including: (1) sources of H, (2) mechanism of absorption, (3) factors affecting the distribution and amount of porosity, and (4) conditions for the removal of gas from the melt. The results of an investigation into the effects of Cl and volatile chlorides on the soundness of castings under prodn. conditions are given. 23 ref.—G. J. S.

Absorption of Gases in Aluminium Castings. Shigeo Ōya (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1951, (2), 33–37).—[In English]. O. describes the effect on the gas absorption in Al castings of the following variables: (i) surface film during melting, (ii) humidity, (iii) reaction between molten metal and water vapour, (iv) melting temp. and holding time, (v) mould material and mould temp., (vi) holding the metal at the solidification point, and (vii) remelting. —G. J. S.

Producing a Large [Aluminium–12% Silicon Alloy] Gravity Casting. —— (*Light Metals*, 1952, 15, (169), 116–117).—Describes the prodn. of a motor lorry gear-box unit weighing 60 lb. in Al–12% Si alloy LM6M, using a gravity die with a sand core.—A. W. B.

The Production of High-Quality Aluminium Die-Castings. Henry H. Ryffel (*Machinery* (*Lond.*), 1952, 81, (2076), 383–387).—R. describes the practice of the Hoover Co., Ohio, using 13 Al-base alloys. Flash and runners are segregated from pot supplies, and are re-alloyed. Reflectoscopes are employed to detect voids. Die-steels are reviewed, those not contg. W being selected for casting Al.—C. P. F.

Aluminium Matchplate Patterns by the Pressure-Cast Plaster Process. I.—American Practice. II.—British Development. (I) E. C. Mantle. (II) D. H. Potts (*Found. Trade J.*, 1952, 93, (1874), 117–120, 120–124; joint discussion, (1875), 157–160).—[I.—] M. deals with the economy, labour saving, and high rate of productivity in the U.S.A. [II.—] British developments are described and illustrated by P.—G. J. S.

The Control of Light-Alloy Castings for Jet-Engines. J. Ophoven ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 119–126).—[In French]. O. describes how control of quality of jet-engine castings is maintained before and after pouring by various operations.—G. J. S.

Bells and the Casting of Bells. —— (*Pro-Metal*, 1952, **5**, (28), 38–42).—[In French and German]. The oldest records of the use of bells in Switzerland go back to the 6th century, the oldest dated bells were made in the 12th century, and a bell cast in 1367 is still in use in Freiburg Cathedral. Attempts have been made to use steel or modified Cu alloys with low Sn content, but the best tone can only be obtained with the classical bell metal (80 : 20 Cu–Sn). The process of moulding and casting of bells is illustrated.—K. S.

Insulating Feeders for Use with Copper-Alloy Sand Castings. R. A. Skinner and R. W. Ruddle (*Found. Trade J.*, 1952, **93**, (1876), 181–184; discussion, 184–185).—The use of exothermic materials and insulating feeder sleeves in increasing the yield of good metal and reducing the size of feeder heads is described, and the relative merits of both types of material discussed.—G. J. S.

Sand-Cast Beryllium Bronze. L. Grand (*Inst. Brit. Found. Preprint*, 1952, (**1033**), 8 pp.; and *Found. Trade J.*, 1952, **93**, (1881), 317–324; discussion, 324–325; (1887), 511–513).—After listing suitable compn. of Be bronze alloys, G. describes and illustrates micro- and macro-structures obtained. Several casting variables and defects—fluidity, grain refinement, pouring temp., shrinkage, piping, cracks, and gassing—are described and compared with those in other Cu-base alloys. The oxidation in air during melting and casting is discussed and the melting losses tabulated. The influence of impurities, metal structure, heat-treatment, and casting conditions are shown, and the relations of U.T.S. and hardness to tempering temp. are illustrated graphically. Details of casting procedure, moulding sands, runners, and feeders are also discussed.—G. J. S.

***Gases in Bronze.** W. T. Pell-Walpole ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 55–69).—[In English]. P.-W. has studied the reaction between gases and the metal constituents in the liq. state, and examined the behaviour of the reaction products during solidification. He discusses the effect of H, O, S, SO_2, and water vapour, and the possible defects they may cause. He enumerates the precautions required and indicates the heat-treatment necessary to minimize gas contamination. 64 ref.—G. J. S.

Nitrogen Degassing and Metal/Mould Reaction in the Production of Gun-Metal Castings. E. C. Mantle (*Found. Trade J.*, 1952, **93**, (1873), 95–98; discussion, 99–100).—M. describes the advantages of a slight gas content in a Sn bronze which is subjected to pressure tests. This gas content is the result of a carefully controlled mould/metal reaction involving the addn. of phosphor Cu to the melt and the thorough degassing of the melt by N.—G. J. S.

Porous Gun-Metal Bush. —— (*Metal Ind.*, 1952, **82**, (2), 31–32).—From the examination of a very porous gun-metal bush casting it is concluded that the moulding and core sands, as well as the alloying and melting of the metal, are of considerable importance.—J. H. W.

***Foundry Crucible Service Life [in Brass Melting].** R. A. Heindl (*Amer. Foundryman*, 1952, **22**, (3), 40–43).—The life of crucibles made in Alabama, Pennsylvania, or Madagascar graphites and bonded with clay or C, was tested for brass melting. The number of heats obtained with C-bonded crucibles was double that obtained from crucibles bonded with clay. The origin of the graphite or the size of flakes did not have any marked effect on crucible life.—V. K.

Manual Handling Minimized in Sixth-Floor Brass Foundry. —— (*Foundry*, 1952, **80**, (11), 110–113).—A short account is given of a small highly mechanized brass foundry.—R. W. R.

Simple Mechanical Layout for the Jobbing Brass Foundry. F. C. Evans (*Found. Trade J.*, 1952, **93**, (1877), 201).—E. describes and discusses a simple plant layout that is widely used in America, and suggests a similar layout using standard British equipment.—G. J. S.

The Non-Ferrous Cleaning Room [in the Brass Foundry]. Martin G. Dietl (*Amer. Foundryman*, 1952, **22**, (3), 44–49).—The layout and equipment of a brass foundry, with particular ref. to the fettling operations, is described.—V. K.

***Segregation [in Lead–Antimony Alloys] During Casting Shown by Radioactive Antimony.** P. J. Killaby, E. J. Taylor, and W. C. Winegard (*Amer. Foundryman*, 1952, **22**, (5), 59).—Segregation in Pb–Sb alloys was studied with radioactive Sb. It is shown that slow cooling in the mould results in the gravity segregation of Sb.—V. K.

Seeking for Grain Fineness in Magnesium Alloys. J. Gris ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 97–104).—[In French]. The results obtained using three different techniques for grain refining Mg alloys are discussed: (1) superheating, as applied to Mg–Al alloys; (2) phys.-chem. reactions, including the effect of the following addn. agents, Cl, $FeCl_3$, $Mg(CO_3)_2$, hexachlorethane, and hexachlorocyclohexane on G-A9 alloy (Al 8·3, Zn 0·5, Mn 0·35%); (3) metallic addn.—addn. of Ce, Ca, and Zr.—G. J. S.

Magnesium Die-Casting. Gilbert C. Close (*Machinery* (*Lond.*), 1952, **81**, (2076), 387–389).—C. describes the work of the Mg die-casting department of McCulloch Motors Corp., Los Angeles. Trials determine optimum temp. of molten metal, temp. of die (controlled by water cooling), die lubricant, length of time die must remain closed after injection to ensure solidification before ejection, injection pressure, and injection speed. Prevention of oxidation of the molten metal is essential to successful Mg die-casting. Taper on contact surfaces of dies and cores ranges from ⅛° to 2°. It has been found practicable to core holes of 0·060 in. dia. —C. P. F.

***A Study of the Direct Bonding of a Tin-Base Babbitt Metal to Grey Cast Iron.** F. T. Smith (*Australasian Eng.*, **1952**, (Oct.), 59–65).—The prepn. of cast Fe surfaces for tinning was investigated. With grey Fe, better bonding was obtained with treatment in fused chloride and fused nitrate baths than in the conventional acid pickle. More research is required to develop a suitable method for spheroidal graphite Fe. 10 ref.—T. A. H.

***Flow of Metal.** —— (*Inst. Brit. Found. Preprint*, 1952, (**1026**), 8 pp.; and *Found. Trade J.*, 1952, **92**, (1869), 673–680; discussion, (1878), 251–252).—Report of Sub-Committee T.S. 35 of the Technical Council. Cinematography has been used to study the flow of metal into moulds. Alterations in the gating and running systems have been studied and the results are illustrated and discussed. The results of experiments studying the effect of metal temp. and the effect of various metals on the flow-patterns are also given. —G. J. S.

Production of Sound Castings. H. W. Griffiths (*Found. Trade J.*, 1952, **93**, (1875), 153–155).—G. describes and illustrates four methods which are used to ensure sound castings. They are: (1) runner system, (2) chilling, (3) pressure feeding, and (4) feeder heads (with or without Thermit materials). —G. J. S.

Why Doesn't Somebody Make a Long Thin Ingot? L. M. Long (*Amer. Foundryman*, 1952, **21**, (6), 53).—It is pointed out that considerable saving in melting time and fuel economy could be obtained if the charge consisted of long thin ingots rather than customarily short fat ones.—V. K.

Applied Hydraulic Theory Improves Casting Technique. James A. Clark (*Iron Age*, 1950, **166**, (3), 90–91).—Proper rate of feed in centrifugal castings is important. Too rapid a rate leads to hot tears and low phys. properties, whereas too slow a rate leads to cold shuts. By using special pouring boxes and maintaining a head for an initial period and then allowing gravity feed to assume control, the lowered head provides the desired reduced metal flow. By experiments with steel C. has shown that hydraulic formulæ are applicable, and they should also apply to the prodn. of ingots and sand castings.—S. R. W.

Knock-Off Feeding Heads.—J. R. Robinson (*Found. Trade J.*, 1952, **93**, (1870), 13–19).—The use of exothermic core compounds and of the Connor block for feeding castings, both ferrous and non-ferrous, are discussed and illustrated. —G. J. S.

Dimensioning Risers and Feeding Heads. J. S. Abcouwer ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 1–11).—[In English]. Official exchange paper of the Nederlandsche Vereeniging van Gieterij–Technici. A. finds that the graphical results of Chvorinov are in agreement with the values obtained using the heat and mass flow analyser. He derives a theoretical equation (linear in form) which is in agreement with the experimentally obtained results of Caine. Two simple examples of the appn. of this formula are given. —G. J. S.

Strain Theory of Hot Tearing. William S. Pellini (*Foundry*, 1952, **80**, (11), 124–133, 192, 194, 196, 199).—The causes of hot tearing in castings were examined by heating lightly loaded rods of a low-m.p. alloy in an oil bath; it was found that ductility vanished and the bar ruptured as soon as the solidus was reached, thus confirming the work of Singer and Cottrell (*J. Inst. Metals*, 1947, **73**, 33; *M.A.*, **15**, 264) on Al alloys. In further experiments, the contraction during solidification of Al–4% Cu and steel castings was restrained and the formation of hot tears observed by radiography; thermocouples embedded in the casting showed that tearing occurred before the casting was fully solid. It was concluded that hot tearing results from the rupture of thin interdendritic films of liq. metal present in the later stages of solidification. The tendency towards hot tearing is a function of the strains imposed during hindered contraction of the casting; these strains are governed by the constitution of the alloy, in particular its freezing range. The theory advanced is essentially similar to those proposed earlier by Lees (*ibid.*, 1946, **72**, 343; *M.A.*, **14**, 274) and Singer *et al.* (*ibid.*, 1947, **73**, 197, 273; *M.A.*, **15**, 264). Hot tearing during welding is also considered. 5 ref.—R. W. R.

Grain-Refinement and Its Effects in Non-Ferrous Casting Alloys. A. Cibula (*Found. Trade J.*, 1952, **93**, (1894), 695–703).—The influence of grain-size on the casting and mech. properties of non-ferrous alloys is described; it is shown that by grain refinement, feeding and resistance to hot-tearing during solidification may be improved, and the mech. properties greatly increased. The mechanism of grain-refining processes is reviewed, and reasons for some of the anomalous effects observed are suggested. Recent developments in the refinement of Al- and Cu-base alloys are indicated. 45 ref.—G. J. S.

Grain Refinement of Non-Ferrous Castings. G. Swinyard (*Inst. Brit. Found. Preprint*, 1952, (**1035**), 6 pp.; and *Found. Trade J.*, 1952, **92**, (1868), 647–652).—After listing the advantages and disadvantages of grain refinement, S. discusses various theories on this subject. The grain refinement of Al by Ti or B, and of Mg by superheating, inoculation, agitation, and the Elfinal process are described, and a possible mechanism suggested. The grain refinement of Cu alloys is also discussed briefly.—G. J. S.

***On Metal Penetration in Casting.** [—I.] Jiro Kasima (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1950, (1), 7–12).—[In English]. K. discusses briefly the effect of two phys. variables, sand grain-size and mould binder, on metal penetration. Penetration by a chem. reaction, oxidation of the metal surface, is described, and the results of several experiments using various alloys of Cu–Zn, Cu–Sn, and Cu–Sn–P are given. It is suggested that CuO is formed at the metal/mould interface and that high moisture contents induce metal penetration.—G. J. S.

Quality Control in the Small Foundry. W. J. Sommer (*Foundry*, 1952, 80, (12), 92–95, 265–268).—The advantages of quality control in the foundry are discussed, and schemes suitable for the small foundry are described.—R. W. R.

Standardizing Casting Practice. A. S. Grot and L. H. Carr (*Amer. Foundryman*, 1952, 21, (6), 42–48).—By maintaining records of examinations of valve castings it was possible to standardize valve design as well as the methods of casting, with large improvements in casting yield.—V. K.

A Classification of Foundry Defects. G. Henon ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 221–248).—[In French]. Official exchange paper of the Association Technique de Fonderie de France. H. presents the foundry defect classification as adopted by the International Commission on Foundry Defects at the Buxton meeting in 1950. The proposed classification is based on the aspect, form, position, dimension, and, generally speaking, on criteria which can be directly observed or measured.—G. J. S.

Utilization of the Microscope in the Non-Ferrous Foundry. P. Mathy ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 89–95).—[In French]. M. gives a few appn. of the use of the microscope in the Cu foundry. He suggests that the microscopic examination of a sample may be more rapid than a chem. analysis, and can be used to control the mix from one melt to another. Macrography is also useful, for the detection of certain defects.—G. J. S.

Mechanical Metal Charging [in Die-Casting]. H. K. Barton (*Metal Ind.*, 1953, 82, (1), 7–8).—Although mech., but operator-controlled, ladling methods have frequently been tried and then abandoned by American die-casters, this technique can contribute substantially to increased productivity, and a simple form of mech. ladling is likely to be adopted in the near future. K. describes the various modifications of the method available, including the more recent Ajax–Tama electromagnetic pump for molten Al.—J. H. W.

New Precision Investment Foundry. Ted Operhall (*Foundry*, 1953, 81, (1), 86–93, 172, 174, 176).—A detailed description, including illustrations, is given of the main features of the investment casting process. Problems arising in process control and equipment are then discussed, and ways of overcoming these described.—G. J. S.

Shell-Moulding Developments [Croning Process]. —— (*Found. Trade J.*, 1952, 93, (1892), 647–650).—Details of a suitable technique for shell-moulding are described and illustrated. Appn. and the economics of the process are also given.—G. J. S.

Current Status of Shell Moulding. Richard Herold (*Amer. Foundryman*, 1952, 22, (2), 42–46).—Prodn. stages: pattern-making, moulding, gating, and risering and pouring of shell moulds are described.—V. K.

Mechanized Shell Moulding. —— (*Metal Ind.*, 1953, 82, (3), 52–53).—Describes mechanized shell moulding in investment and curing, using the Polygram automatic shell-moulding machine, Mark IV.—J. H. W.

Shell Moulding and Core Making Adapted to the Small Shop. Wilbur S. Walters (*Amer. Foundryman*, 1952, 22, (6), 42–45).—A small shell moulding foundry at LaGrange Shell Molders, Inc., LaGrange, Ind., is described.—V. K.

Shell Moulding at International Harvester Co.—I.–II. Garnet P. Phillips (*Foundry*, 1952, **80**, (11), 102–107, 282–292; (12), 96–101, 177, 182).—[I.—] A detailed description is given of the " Contour " shell-moulding process, a development of the Croning process. [II.—] P. describes the layout and techniques used in the shell-moulding foundry of the International Harvester Co., Chicago. Some general information is given on the machines available for the mechanized prodn. of shell-moulds.—R. W. R.

New Method Simplifies Shell-Mould Assembly. Richard Herold (*Foundry*, 1952, **80**, (12), 142, 144).—In the method described the two halves of the mould are joined by means of a number of male bosses moulded in the cope, which engage with female recesses in the drag. Glue is applied to the bosses before they are inserted into the recesses.—R. W. R.

Some Recent Danish Experiments with the C-Process [Shell Moulding]. O. Hoff ([*Proc.*] *Congr. Internat. Fonderie, Bruxelles*, **1951**, 399–408).—[In English]. H. first enumerates the advantages of the C-process and then describes various sands and moulding techniques that have been studied. The use of this process in the manufacture of cylinders for motor cycles is described and illustrated.—G. J. S.

Casting in Cement-Bonded Sand. J. B. McIntyre (*Foundry*, 1952, **80**, (12), 90–91, 269).—A short account is given of the Randupson process of making cement sand moulds, with particular ref. to the prodn. of marine propellers and other large castings.—R. W. R.

On Moulding-Sand Binders. Toshisada Makiguchi (*Rep. Casting Research Lab., Waseda Univ., Tokyo*, 1951, (2), 38–41).—[In English]. The relationships between binders, baking strength, and the high-temp. strength of binders are examined. The connection between binders and the stability of grains on the sand surface is also discussed.—G. J. S.

Investment Casting for Engineers. (Wood and Ludwig). See col. 671.

19 — WORKING

Arvida [Aluminium Rod] Mill Is Fully Automatic and Repeating. H. Jones (*Iron Age*, 1950, **166**, (22), 72–75).—A description is given of a Belgian rod mill, installed in Canada, which is chiefly used for producing $\frac{3}{8}$-in.-dia. rod in soft Al for elect. conductors.—N. B. V.

New Rolling Mills for the Production of Aluminium Foil. M. Coquand (*Rev. Aluminium*, 1952, (189), 235–238).—A description is given of a new plant installed by Ets. Charles Coquillard, Isère, for the manufacture of Al foil ~0·0035 in. thick and 3·33 ft. wide. It consists of a 4-high roughing mill and 2-high finishing mill, with a tunnel-type continuous annealing furnace and auxiliary plant. Rolling speeds up to 3280 ft./min. can be obtained with this equipment. —A. W. B.

The Production of Aluminium Foil. —— (*Aluminium Courier*, 1951, (17), 10).—Describes the manufacture of foil stock by hot and cold rolling from cast billets to $\frac{1}{2}$ mm. thickness in widths up to 30 in. This is reduced to the thinnest foil—~0·005 mm.—in five or six passes without intermediate annealing. Two thicknesses are spooled together for the final pass to the thinnest gauges. Annealing at ~500° C. over long periods is essential to volatilize the rolling oil.—C. P. F.

Stretch-Forming Machine Shapes Extruded Aluminium Parts. —— (*Iron Age*, 1952, **170**, (25), 137–139).—A 60-ton capacity Model A 12 Hufford stretch-forming machine is used to form the contour in Al-alloy aircraft components, the material being usually stretched in the range of 1·5–3·0%.—J. H. W.

Producing Aircraft Parts by Hydroforming. G. B. Lewis and J. S. Corrall (*Machinery* (*Lond.*), 1952, **80**, (2045), 160–164).—Describes the Hydroform hydraulic press in which the ram is designed to act as a forming member. Pressures up to 8000 lb./in.^{2} are used, and the hydraulic fluid, contained in a cavity in the ram, exerts pressure on a rubber diaphragm. The use of the rubber diaphragm causes the forming forces to be everywhere normal to the surface, thus minimizing local stresses.—C. P. F.

[Aluminium] Sheet-Metal Forming Without Vertical Press Movements. —— (*Machinery* (*Lond.*), 1952, **81**, (2079), 631–633).—Describes the Wheeldon direct-acting hydraulic press, which is only about one-tenth the size of a comparable press required for the Guerin process, yet is capable of delivering up to three times the pressures currently employed for this type of work. 0·25 in. Al-alloy sheet is workable. —C. P. F.

Techniques and Problems in Large Extrusion Production [of Aluminium and Aluminium Alloys]. T. F. McCormick (*Iron Age*, 1952, **170**, (24), 158–161).—The operation of an extrusion press for Al and Al alloys with a rated capacity of 13,200 tons is described. This press is of 2·5 times greater capacity than any other now in operation, and requires auxiliary equipment on a comparable basis.—J. H. W.

How To Taper Aluminium Plate by Abrasive Belt Grinding? P. J. Queyrel (*Iron Age*, 1952, **170**, (23), 174–176).—Abrasive-belt grinding provides a faster and cheaper method of tapering aircraft wing structures by removing large amounts of metal over wide flat surfaces at high speeds without overheating the metal to a tolerance of down to 0·01 in.—J. H. W.

Modernization of the Rolling Mill at Dornach [for Copper and Copper Alloys]. —— (*Pro-Metal*, 1952, **5**, (26), 970–985).—[In French and German]. The cold-rolling department of Metallwerke A.G., Dornach, has been completely modernized. Hot rolling continues on the old installation. but is not taken as far as formerly; the hot-rolled Cu and Cu alloy plates are milled on both faces. The Bliss 4-high cold-rolling mill is capable of very heavy reductions and very rapid rolling (100–300 m./min.). Annealed strip or sheet moves through a continuous pickling installation on a roller table with stainless-steel rollers; the pickling soln. and wash water are sprayed on to the stock. Mech. transport and facilities for coiling and uncoiling the brass strip at various stages ensure rapid prodn.—K. S.

Brass and Bronze Forgings. Carl H. Pihl (*Machinery Lloyd* (*Eur. Edn.*), 1952, **24**, (1A), 53–57).—P. defines forging and surveys briefly the history of this method of working. He lists the properties of forged Cu and Cu-base alloys and stresses that the grain structure of brass and bronze forgings is uniform and dense. A second table lists the commercial dimensional tolerances for Cu and Cu-base alloys. Economic reasons for chosing forging as a manufacturing technique are expounded, with examples.—C. P. F.

Production of Brass and Copper Wire. Benjamin H. McGar (*Wire and Wire Products*, 1952, **27**, (11), 1178–1179).—A brief discussion of the properties of yellow (65 : 35) brass and cartridge (70 : 30) brass, used for making screws and rivets. Extrusion details are quoted, together with a short description of the prodn. of wire from Cu-rich alloys. —C. P. F.

Transfer Moulding of [Copper] Commutators. —— (*Brit. Plastics*, 1952, **25**, (277), 198–200).—Describes the mass-production of commutators for fractional-h.p. motors, including sizes down to $\frac{3}{16}$-in. dia. commutators. The commutator made by Nelco Ltd., of Shalford, consists of Cu segments and mica or Micanite separators which are assembled by hand. The whole is compressed, pressed again in a mould filled with a phenolic moulding powder, and heat-treated at 330° F. (165° C.). The moulded commutators can be dip-soldered and skimmed-up as finished items.—C. P. F.

Manufacture of a [Brass Musical] Instrument. Fr. Hirsbrunner (*Pro-Metal*, 1952, **5**, (25), 945–948).—[In French and German]. Tough sheets of brass or nickel silver, 0·35 to 0·60 mm. thick, have to withstand a considerable amount of plastic deformation if the brass instrument produced is to be free from imperfections which affect its tone. The sheet is cut and bent to the desired shape, and all seams are brazed. The joint is hammered on a mandrel until its thickness is uniform with the sheet. The enlarged end of the tube is spun to shape and filled with Pb before bending. The highly polished components are assembled and soft soldered; correct adjustment of the overall length determines the pitch of the instrument. The piston mechanism is then mounted, and stabilizing supports are soldered on the tube. Finally the instrument receives a high polish.—K. S.

Tools for the Impact Extrusion of Magnesium. —— (*Machinery* (*Lond.*), 1952, **80**, (2064), 982–986).—The impact extrusion of Mg is carried out by basically the same methods as those commonly adopted for other metals, save that the operating temp. may vary from 350° to 700° F. (212°–372° C.) according to the alloy used, the speed of operation, and the size of part to be produced. Details of the punch steels and clearances are quoted. Colloidal graphite is preferred as a lubricant, but this must be applied just before the operation and removed immediately afterwards to eliminate corrosive action.—C. P. F.

Forming and Fabricating Seamless Nickel Tubing. —— (*Machinery* (*Lond.*), 1952, **80**, (2044), 109–113).—Describes modern machining practices suitable for seamless tubes of Ni, Monel, and Inconel, including expanding into supporting plates or tube-sheets, flanging to form Van Stone or rolled joints, threading, bending, coiling, and bending with the aid of fillers. Various proprietary fillers are reviewed, sand not being recommended. Press-bending, roll-bending, and hot-bending are described, with recommended size-ranges for each appn., and temp. ranges for hot-bending of the alloys are indicated.—C. P. F.

Forming Titanium Parts for Aircraft. O. A. Wheelon (*Machinery* (*Lond.*), 1952, **80**, (2047), 223–228).—W. instances an aeroplane in which some 849 Ti parts effected a weight-saving of 395 lb. over stainless or other alloy steels. After detailing the machining of Ti and the speeds, cuts, and lubricants suitable, W. mentions a tendency to warp as a result of machining, and remarks that almost all parts require subsequent straightening.—C. P. F.

***The Influence of Insoluble Phases on the Machinability of Titanium.** (Goldhoff *et al.*). See col. 627.

The Machining of the Rarer Metals. —— (*Machinery* (*Lond.*), 1953, 82, (2095), 72–73).—Brief notes on the optimum cutting speeds, cut depths, feeds, and tool-angles applicable to W, Mo, Ta, Ti, and Zr. Prolite 1C tool steel is recommended throughout, except for the low-speed turning of Ta, for which grade 15A is specified.—C. P. F.

Die Cost Cut by Tool Shop Foundry. Walter G. Patton (*Iron Age*, 1950, **166**, (18), 81–83).—Describes an American plant in which a foundry has been installed for casting Zn alloy dies. Their low cost makes such dies suitable for making die-design trials on difficult pressings and also for short runs where the expense of permanent steel dies is not justified.—N. B. V.

Low-Cost Methods Speed Die Construction. W. Curtis Miller (*Machinist* (*Eur. Edn.*), 1953, **79**, (2), 49–52).—M. describes the appn. of Zn alloys with expanding plaster patterns and of non-shrinking resin to the manufacture of large dies for stamping. The use of these materials—sometimes combined—cuts costs and accelerates accurate prodn.
—J. H. W.

***On the Temperature Rise of the Surfaces in Rolling Contact.** Toshio Nishihara and Kichirō Endō (*Bull. Eng. Research Inst., Kyoto Univ.*, 1952, **1**, (March), 52–56).—[In Japanese, with figures and graphs in English]. As temp. rise seems to influence the pitting strength of a material, N. and E. have determined the max. temp. rise on surfaces in rolling contact by Cu/Constantan and steel/Constantan thermocouples. Results confirmed the calculated values of the effects of loads, sliding speeds, heat conductivities of the metals, and coeff. of friction of the lubricants.—AUTHORS.

A Fully Automatic Extrusion Press. —— (*Light Metals*, 1952, **15**, (169), 109–110).—A description of the main operating features of a 2,500-ton automatic extrusion press suitable for non-ferrous and light metals.—A. W. B.

[Mathematics of] Extrusion and Punching Processes. A. Geleji (*Acta Tech. Acad. Sci. Hungar.*, 1952, **4**, (1/4), 273–292).—[In German]. A math. theory of extrusion and punching processes, enabling the pressures required for working different materials at various speeds to be calculated with sufficient accuracy for practical purposes is developed. 13 ref.
—J. S. G. T.

Metal Forming by the Spinning Process. R. H. Warring (*Machinery Lloyd* (*Eur. Edn.*), 1952, **24**, (19A), 53–57).—W. describes the technique and scope of metal spinning, the use of lubricants in spinning, and the tools used. The % suitability of 22 ferrous and non-ferrous metals and alloys for shallow spinning and for deep spinning is listed.—C. P. F.

The Graphical Determination of Drawing Sequences for Tubes. A. Geleji (*Acta Tech. Acad. Sci. Hungar.*, 1952, **4**, (1/4), 347–364).—[In German]. A graphical method for ascertaining quickly the essential information required in any tube-drawing process—the necessary traction, number of draws, &c.—is developed, and the use of the proposed method is exemplified by appn. to the drawing of brass tubes. 6 ref.
—J. S. G. T.

Machining and Drawing Fluids. E. L. H. Bastian (*Steel*, 1950, **126**, (22), 58–62, 80, 83).—The types and appn. of cutting and grinding fluids and drawing oils and compounds are discussed and advice given on the selection of the correct fluid to use for various ferrous and non-ferrous metals.
—S. R. W.

***New Electro-Mechanical Methods for Cutting and Grinding of Metals in Laboratory Practice.** L. Ya. Popilov (*Zavod. Lab.*, 1948, **14**, (3), 358–361).—[In Russian]. Two machines for electro-mech. cutting and grinding of metals and alloys are described. The process depends on the disintegration which occurs at the anode when two metal surfaces carrying a current slide against each other in the presence of a liq. medium. The cutting machine consists of a sheet Fe cutting disk, 0·3–0·8 mm. thick and 250–300 mm. dia., revolving at an optimum peripheral speed of 12–18 m./sec., which is pressed against the specimen held in an insulated clamp. D.C. at 22–25 V. and 20–100 amp., as supplied by a welding generator, may be used (the specimen being the anode), but increased cutting speed is obtained by superimposition of 6–9 V. A.C. on 15–18 V. D.C. The liquid, which must be supplied copiously, may vary widely, but that most used is based on water-glass ($Na_2SiO_3.nSiO_2.mH_2O$, $SiO_2 : Na_2O =$ 2–3, sp. gr. = 1·30–1·35); satisfactory results can be obtained from suspensions of clays with 0·1–0·2% NaOH or 2–3% borax. The finish depends on the conditions, and roughness may vary from 1–2 to 200–300 μ. The cutting speed is 5–12 cm.^{2}/min. The grinding machine is similar, but employs a grooved brass grinding disk, 12–22 V. D.C. and 5–60 amp., depending on area and finish, and water-glass soln. of sp. gr. 1·28–1·32. By decreasing the voltage, the surface roughness may be reduced to a fraction of 1 μ. The advantages of the methods are increased prodn., saving in abrasives, improved surface finish, and absence of surface disturbance. 8 ref.
—D. M. P.

20 — CLEANING AND FINISHING

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

Polishing Aluminium. Frank Spicer (*Pract. Eng.*, 1952, **26**, (660), 485–486).—A general account is given of the finish polishing of components made of Al castings.—E. J.

Tinplate Cleaning and Annealing Cycle May Be Cut to Less than Two Days. Alfred E. Kadell (*Steel*, 1950, **127**, (4), 72).—A continuous cleaning and annealing line for tinplate strip is diagrammatically shown and briefly described.
—S. R. W.

Barrel Finishing: An Economical Means of Processing Small Die-Castings. H. K. Barton (*Machinery* (*Lond.*), 1952, **80**, (2054), 555–560).—Wet and dry tumbling with both carriers and agents are reviewed, together with the optimum shape and lining of the barrel. The proportions of loading to vol., and of work to carrier or agent are discussed. Practical remedies to overcome such faults as the hammering up of the flash on a die-casting and the incomplete removal of fins are detailed.—C. P. F.

How Aircraft Engine Parts Are Barrel Finished. Dudley J. Kaharl (*Iron Age*, 1950, **166**, (2), 81–83).—The improved quality and lowered costs obtained at Pratt and Whitney Aircraft plant at East Hartford, Conn., by barrel finishing a wide variety of Al, brass, and steel parts are described.
—S. R. W.

Wet Tumbling Lowers Deburring Costs. Robert H. Anderson (*Iron Age*, 1950, **166**, (22), 69–71).—A. describes the deburring and burnishing of Al, Mg, Zn, and steel parts, varying in weight from 1 oz. to 15 lb., in Roto-Finish barrels at an American plant making aircraft components. Granite chips of various sizes are used as abrasive.—N. B. V.

Cleaning Work with the Aid of Ultrasonic Vibrations. Charles R. Fay (*Machinery* (*Lond.*), 1952, **80**, (2061), 853–855).—F. describes a method of cleaning small precision metal parts by immersion in an ultrasonically activated solution. A power oscillator supplies voltage at 750 kc./s. to a quartz crystal transducer which produces ultrasonic vibrations within a solution of C_2HCl_3 via a diaphragm. A slightly saline soln. couples the crystal to the diaphragm.
—C. P. F.

Monorail Conveyor System Reduces Blast-Cleaning Costs. W. I. Gladfelter (*Foundry*, 1952, **80**, (12), 104–105, 263–265).—A highly mechanized system for cleaning cylinder-block castings is described.—R. W. R.

How to Select Abrasives for Blast Cleaning Metals. N. F. Stine (*Amer. Foundryman*, 1952, **22**, (2), 39–41).—Properties and appn. of abrasives for surface cleaning are reviewed.—V. K.

21 — JOINING

Sealing Metal to Glass. R. H. Warring (*Elect. Rev.*, 1952, **151**, (3898), 301–304).—The methods of sealing metals to glass used in the lamp and valve industry are reviewed. —S. R. W.

Synthetic Resin Bonding of Aluminium and Aluminium Alloys. —— (*Aluminium Courier*, 1952, (19), 10–11).—Describes the use of synthetic resin adhesives for bonding sheets of Al and Al alloys to each other or to other media, including wood, rubber, ceramics, &c.—C. P. F.

***Joining of Powder Compacts to Solid Metals by Pressing and Sintering.** (Hofmann and Schmalenbach). See col. 636.

The Properties and Driving of Large Aluminium Alloy Rivets. J. C. Bailey (*Symposium on Welding and Riveting Larger Aluminium Structures* (*Aluminium Development Assoc.*), **1951**, 133–154; discussion, 176–219).—Progress made since 1946 in the development of Al alloy rivets $>\frac{3}{8}$ in. in dia. is discussed. Investigations into the manufacture and characteristics of such rivets, and the technique of driving them, are reported. The alloys N 6, H 10, and H 13 are most suitable for rivets. Rivet points of reduced size are described which enable rivets of $\frac{7}{8}$ in. dia. to be closed. The effect on rivet shear strength of hole clearance, method of driving, point shape, and alloy compn. are considered. Experience with riveting is reported in the discussion. —K. B.

The Design Characteristics of Aluminium Riveted Joints. S. C. Redshaw (*Symposium on Welding and Riveting Larger Aluminium Structures* (*Aluminium Development Assoc.*), **1951**, 155–175; discussion, 176–219).—Factors influencing the design of Al alloy riveted joints are discussed and, from the results of an extensive programme of tests on double shear joints, the effect of variations in joint materials and proportions are considered. Recommendations are made for the design of satisfactory joints. The fatigue-resistance of joints is also discussed. 9 ref.—K. B.

***Studies in the Closing of Large-Diameter [Aluminium] Rivets.** Charles Guinard (*Rev. Aluminium*, 1952, (189), 249–255).—After giving performance data for a range of pneumatic hammers closing half-round and countersunk heads G. reports the results of tests made on factors influencing driving behaviour. A smaller size, and therefore lesser weight, of snap and good backing up were found to assist driving, whilst a reduction in height and dia. of half-round head enables larger rivets to be driven. The max. sizes of countersunk, small half-round, annular, and recessed heads which can be closed with various hammers were established for rivets from 14 to 22 mm. dia.—A. W. B.

Induction Heating Doubles Cylinder Soldering Production. John Nelson (*Steel*, 1950, **127**, (11), 95–98).—The appn. of Ag soldering using induction heating is described. Increased prodn., lower costs, and improved quality are among the advantages indicated.—S. R. W.

Induction Brazing Can Be Used for Copper and Copper-Alloy Parts. E. M. Laughner (*Machinist* (*Eur. Edn.*), 1953, **97**, (3), 106–108).—In spite of the high elect. and thermal conductivity of Cu and its alloys, brazing by induction heating often results in considerable saving in cost, better quality, and smaller losses owing to automatic control. A number of examples are given.—J. H. W.

Copper Paste Lowers Brazing Costs. Dan Reebel (*Steel*, 1950, **127**, (3), 82–83, 98).—The use of a Cu paste consisting of the red Cu oxide in a jellied synthetic petroleum material, as a brazing medium, is illustrated. After appn. of the paste to the work to be brazed it is passed slowly through reducing gases in a furnace held at 2070° F. (1130° C.).—S. R. W.

Brazing with Metal Powder. H. W. Greenwood (*Machinery* (*Lond.*), 1952, **81**, (2070), 98–99).—G. reviews the brazing of Fe and steel with Cu, brass and bronze, and Sif-bronze for working temp. up to 500° C., and Ag solders for higher temp. Reference is made to the braze-jointing of Al and Al alloys with an Al–Si eutectic, and salt-bath brazing is described as being both efficient and economical.—C. P. F.

Recent Researches on the Arc Welding of Thick Aluminium-Alloy Plate. P. T. Houldcroft, W. G. Hull, and H. G. Taylor (*Symposium on Welding and Riveting Larger Aluminium Structures* (*Aluminium Development Assoc.*), **1951**, 12–43; discussion, 78–130; also (abridged) *Welding*, 1952, **20**, (1), 27–31).—Problems involved in joining thick plate are reviewed. High heat input is necessary. The weldability of Al alloys, the cracking tendencies of various alloys, and the cause and prevention of gas porosity are discussed. Metal-arc welding techniques are described. Continuous welding is effected by using two electrodes alternately or by employing two operators welding alternately. A V-type of edge prepn. is preferred for butt joints in plates from $\frac{1}{4}$ to 1 in. thickness, a double V for those over 1 in. Requirements of electrodes are discussed. A-arc (W-arc) welding technique is described. A.C. is almost always used for Al alloys. Elect. characteristics are discussed, and oscillographic studies are reviewed. Torches suitable for currents up to 600 amp. with water-cooled nozzles are described. Thoriated-W electrodes increase the max. rating of a torch. Details are given of the "self-adjusting arc" process (known in America as "Aircomatic"), in which an arc is struck between the work-piece and a thin wire electrode fed at const. speed; and of the "controlled-arc" process, in which the wire electrode is fed at a speed which is controlled by some external device so that the arc length remains const.; both processes utilize an inert shielding gas. Techniques and appn. of the processes are considered. Experiences with the arc-welding processes are reported in the discussion.—K. B.

Practical Aspects of the Argon-Arc Welding of Aluminium Alloys. J. R. Handforth (*Symposium on Welding and Riveting Larger Aluminium Structures* (*Aluminium Development Assoc.*), **1951**, 44–77; discussion, 78–130; also *Welding*, 1952, **20**, (1), 31–34).—Practical aspects reviewed are equipment (elect. apparatus, torches, electrodes, filler-rods), techniques for hand and machine welding and for circumferential welding, and safety precautions. An extensive *bibliography* is appended.—K. B.

The Welding of Aluminium by Argon-Arc Methods. —— (*Aluminium Courier*, 1952, (20), 11).—Describes the consumable-electrode process and A-arc welding with a W electrode; the former is considered preferable for marine appn.—C. P. F.

Repair of Aluminium Castings, &c. [by Welding]. —— (*Aluminium Courier*, 1952, (18), 10–11).—Describes the fusion welding processes applicable to Al, and stresses that the compn. of an alloy to be welded should be known so that the correct filler-rod or electrode may be selected.—C. P. F.

***Surface Preparation in the Spot Welding of Light Metal Alloys.** Isamu Ukita and Tatsuya Hashimoto (*Bull. Eng. Research Inst., Kyoto Univ.*, 1952, **1**, (March), 24–28).—[In Japanese, with figures and graphs in English]. In order to spot-weld the light alloys successfully, it is necessary to clean the surfaces and to minimize the contact resistance between the contacting surfaces. U. and H. report their experiments with Al, SDH, and SDCH to determine the effect of various surface prepn. and also the correct conditions.—AUTHORS.

Welding and Brazing Seamless Nickel Tubing. —— (*Machinery* (*Lond.*), 1952, **80**, (2051), 426–429).—Reviews the jointing of pipes and tubes of Ni, Monel, and Inconel to similar parts or to headers, flanges, and fittings by arc, oxy-acetylene, or inert-gas metal-arc welding; Ag or Cu brazing; and soft soldering.—C. P. F.

The Argon-Arc Welding Process.—— (*Light Metals*, 1952, **15**, (172), 232–234).—The A-arc process is discussed, and comparison of the characteristics of an A-shielded arc with D.C. current and the electrode either positive or negative, with those of an arc with A.C. current shows that the latter is more effective in enabling both the oxide film to be broken and the electrode to be kept cool than the former. Since the arc system acts to some extent as a rectifier, a D.C. suppressor and H.F. unit are incorporated in the circuit.—A. W. B.

M.L. "Queen Elizabeth": Field Trial of Aircomatic Welding. —— (*Light Metals*, 1952, **15**, (171), 187).—A brief description of the A-shielded self-adjusting arc-welding process applied to the reconstruction of a river launch. No details of the actual technique used in this field trial are given.—A. W. B.

***A.C. Argon Arc Welding at Less than 50V. R.M.S. Open Circuit.** J. C. Needham and L. H. Orton (*Trans. Inst. Weld.*, 1952, **15**, (6), 161–165).—The stability of the A.C. arc depends on the re-ignition voltage available at zero current. An electronic surge injector unit has been designed which provides adequate re-ignition voltage for A arc welding at a power-circuit voltage of <50 V. r.m.s. This represents a considerable improvement on H.F. spark injection, which is not always satisfactory and which generates severe radio interference. The arc maintenance surge is inadequate for starting from cold, but suitable sparks can be generated from the surge for spark-starting across a cold gap. This device can be used with any type of low-voltage welding transformer. —K. S.

Resistance Welding. C. E. Slade (*Trans. Liverpool Eng. Soc.*, 1950, **71**, 11–25).—The development of the resistance welding process is reviewed, and the process itself is examined in detail. The types of equipment suitable for particular appn. are considered, and the welding of both steel and light alloys is described.—S. V. R.

Hard Facing with Stellite. —— (*Machinery* (*Lond.*), 1952, **80**, (2046), 179–188).—Reviews the Co–Cr–W alloys known collectively as Stellite and details several appn. of these as hard-facing and cutting media. Classifies the group as between high-speed steel and the cemented carbides. —C. P. F.

24 — BIBLIOGRAPHY

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25 — BOOK REVIEWS

Fatigue and Fracture of Metals. Edited by William M. Murray. A Symposium held at the Massachusetts Institute of Technology, 19–22 June 1950. 9 × 6 in. Pp. viii + 313, with 190 illustrations. 1952. Cambridge (Mass.): Technology Press of the Massachusetts Institute of Technology; New York: John Wiley and Sons, Inc. ($6.00); London: Chapman and Hall, Ltd. (48*s.*)

It is over a hundred years since the meeting at Birmingham of the Institution of Mechanical Engineers at which J. E. McConnell declared that metals undergo mechanical deterioration as a result of fluctuating stresses. It is only five years short of a century since Wöhler's pronouncement:

> "Wrought iron and steel will rupture at a stress not only less than the ultimate static strength of the material but even less than the elastic limit if the stress is repeated a sufficient number of times."

Since these fine pioneering efforts, fatigue of metals has, at ever-increasing tempo, been under consideration by the engineers, metallurgists, and physicists, not only of this country but of the whole scientific world, yet can it be said that agreement has been reached as to its mechanism?

In 1950, a symposium on fatigue and fracture was held at the Massachusetts Institute of Technology, and now the fourteen papers then presented are available in book form. While one cannot say they constitute any considerable contribution to fundamental knowledge—the writer doubts whether any author has actually made a truly novel contribution—each in his way deals interestingly with some particular aspect of the problem. There are three papers relating to fatigue in particular fields, namely aeroplane structures, the hulls of sea-going vessels, and machinery. In these, the authors' approach is an engineering one. On the other hand, there are a number of other papers dealing with the question from the metallurgical, physical, or statistical point of view.

To metallurgists probably the two most interesting papers are those of Professors Newmark and MacGregor. The former gives a workmanlike review of present evidence as to the "Cumulative Damage Rule" and, in the opinion of the writer, rightly finds the case for the acceptance of Miner's attractive concept "Not proven". MacGregor's paper contains much to excite attention, but, in particular, his experiments showing that the transition temperature of ferritic steel is markedly raised by cyclic stresses *below* its fatigue limit, has a particular importance for all those concerned with the use of such material at low temperatures. Somewhat surprisingly, both these authors consider that this change, arising from stress reversals, might be used as a means of determining the measure of fatigue which has taken place. The writer would have thought that the making of a stress/strain diagram would have been equally informative and far less troublesome.

The papers, as is almost always the case with American technical publications, are excellently reproduced, but the book, as is the regrettable custom with symposia, is without an index, and is therefore destined to be less useful than might otherwise have been the case.

P. L. Teed.

Metal Interfaces. A Seminar held during the Thirty-Third National Metal Congress and Exposition, Detroit, 13–19 October 1951. 9 × 6 in. Pp. [vii] + 348, with numerous illustrations. 1952. Cleveland (3), O.: American Society for Metals, 7301 Euclid Avenue. ($5.00.)

This volume contains twelve papers and records some discussion. As is to be expected in a symposium, there is a certain amount of duplication, but the papers are all good, some are excellent, and the volume makes a valuable addition to metallurgical literature.

The last few years have seen very rapid advances in our knowledge of the structure and behaviour of grain boundaries in metals. While the importance of crystal boundaries has never been lost sight of since the early work of Bengough, Rosenhain, and others, there was until about 1947 a sort of stalemate in that the amorphous-cement theory suggested by the early workers, although theoretically unattractive, accorded well with the observed behaviour of grain boundaries, whereas the theoretically more attractive transition lattice hypothesis had no direct experimental support. The difficulty lay in obtaining quantitative information concerning intrinsic properties of grain boundaries for comparison with theoretical prediction. In 1948 the full importance of the boundary as an interface characterized by a surface energy was realized, the significance of surface-energy considerations in dictating details of microstructure was shown by Smith, and means of measuring boundary energies were discovered. At about the same time, theory reached the stage where boundary energies of transition lattices treated as assemblages of dislocations could be calculated. Numerous workers have now investigated grain-boundary energies and have found very encouraging agreement with theoretical predictions. Another boundary property to which quantitative measurement has been applied, starting with the work of Kê in 1947, is that of resistance to relative shear between crystal grains. Much work has been done in this field, and the advances in both these directions have stimulated considerable interest in the study of crystal boundaries. There was therefore much new and interesting work to record at the symposium.

There are two papers concerned with free surfaces; the first by C. Herring outlines the atomistic theory of metallic surfaces, and in the other, on "Metal Surface Phenomena",

H. H. Uhlig deals briefly with adsorption, passivation, and corrosion-resistance.

In an excellent paper on the "Theory of Internal Boundaries", H. Brooks deals in detail with the treatment of boundaries as assemblages of dislocations and compares the predictions with experimental results. Methods of measuring the interfacial energies of solid/gas and solid/liquid interfaces are critically described by H. Udin and those of solid/solid boundaries by J. B. Hess. In a paper on "Energies and Structure of Grain Boundaries", K. T. Aust and B. Chalmers review the information available on boundary structure and consider in detail measurements on the variation of boundary energy with crystal orientation.

Fundamental considerations relating to microstructure are discussed in C. S. Smith's excellent paper on "Grain Shapes and Other Metallurgical Applications of Topology". In this, the mathematical requirements of space filling—a field relatively unfamiliar to the metallurgist—are lucidly described and the details of microstructure considered as resulting from the interaction between topological demands and geometrical requirements of surface-tension equilibrium. R. J. Fullman discusses "Boundary Migration during Grain Growth" and P. A. Beck, "Interface Migration in Recrystallization". As might be expected, there is much that is common to these two papers, but the approaches are somewhat different, and together they make a useful study of the whole field of boundary migration.

Mechanical properties of grain boundaries are described in two papers. A. S. Nowick deals very efficiently with anelastic phenomena associated with grain boundaries in a paper on "Relaxation Across Interfaces", and B. Chalmers in a short paper on "Mechanical Effects of Interfaces" discusses critically the problem of relating the differences in behaviour between single crystals and polycrystalline aggregates to the existence of internal interfaces and deals briefly with some effects of the free surface.

The significance of the boundary as a region where phase changes may be initiated more readily than in the grain interior is discussed by A. H. Geisler in a paper on "Phase Transformations at Interfaces".

The discussion of the papers is surprisingly brief, but there are some valuable contributions.

Although it is to be expected that in a symposium there may be some omissions, it is a matter of surprise that no attention is given to grain-boundary diffusion and very little to general chemical properties. Nevertheless, this is the first time that most of the known facts relating to boundaries have been brought together in one volume, and it is altogether a valuable book.

RONALD KING.

Practical Metallurgy for Engineers. By the Research Staff of E. F. Houghton and Co. Fifth edition. $8\frac{1}{2} \times 5\frac{1}{2}$ in. Pp. [vi] + 599, with 193 illustrations. 1952. Philadelphia (33), Pa.: The Company, 303 W. Lehigh Avenue. ($3.50.)

The fifth edition of this book begins with a 16-page account of the physics and chemistry of metallurgy. In the next 94 pages metals and their ores, fuels, refractories, non-ferrous metals, iron ores, the manufacture of iron and steel, mechanical treatment, and temperature measurement are dealt with. Although the metallography and heat-treatment of non-ferrous alloys are included, the rest of the book is mainly concerned with steel and its heat-treatment: quenching, hardenability, testing and inspection, metallography, furnaces, salt baths, surface-hardening, and cleaning. The subject matter is so much out of balance as to belie the title of the book. An example of the authors' bias is the fact that an exploded view of a "Carbo Cleaner" for cleaning and screening used carburizing compounds is allotted four times the space occupied by a photograph of an electron microscope. The photographs both of equipment and of microsections are a mixed bag; some are very good, some useless.

The authors often show an unusual choice of words. One is intrigued by "The high cost of the tools themselves has caused metallurgists to devote much time to this (hardening) treatment"; suspicious of "A gas is thus obtained which is essentially composed of N_2, CO and H_2 which poses as a suitable atmosphere for a wide variety of purposes"; irritated by "Sorbite, the most ideal structure for nitriding"; and enchanted by "Merry-go-round with work pots in a circle and work conveyed by center jig".

For those who need to acquire a smattering of metallurgy or a sound up-to-date knowledge of steel heat-treatment practice, this book may be recommended. The reading matter is refreshingly lucid, and is written for and by practical men. References are rare. Engineers will find the book easy to read, and will be left with a pleasant but false impression that never again will they be fuddled by the metallurgist.

D. M. DOVEY.

Modern Magnetism. By L. F. Bates. Third edition. $8\frac{1}{2} \times 5\frac{1}{2}$in. Pp. xii + 506, with 134 illustrations. 1951. Cambridge: Cambridge University Press. (30*s.*)

This book has been written by one who has received international recognition for his outstanding contributions to magnetic research, and who is, at the same time, an experienced university teacher. This happy combination of teaching and research is displayed in the book under review. The first edition enhanced the author's reputation, and its deserved popularity among students and research workers has led to the publication of a third revised and much expanded edition. The book covers almost every aspect of the subject, starting with early fundamentals and ending with an account of the latest trends in magnetic research. The great appeal of the book lies in the excellent balance which is maintained between the theoretical and experimental treatment of the subject and its continued emphasis on fundamental principles. The mathematics employed throughout the book is well within the compass of the average honours student in physics.

Among the many excellent chapters, the reviewer was especially interested in the one dealing with the domain concept and the hysteresis cycle. The description of the experimental study of domains by the Bitter figure powder pattern method, to which Professor Bates has made important contributions, will have a special appeal to the metallurgical student, for it employs a micrographic technique reminiscent of "magnetic etching" which is used to distinguish magnetic from non-magnetic phases. The discussion of magnetic domains leads naturally to the idea of "critical domain size" and the practical realization of the idea in the form of a permanent magnet made by compressing pure, finely-divided iron particles. The method of preparing such powders, which would have been of great interest to a metallurgist, is not disclosed. A metallurgist would have also welcomed a description of the preparation and properties of somewhat coarser iron powders used in telecommunication and radio applications, and some comparison of these properties with those of the ferrites, of which brief mention is made.

The authoritative text is supported by a large number of tables and graphs. There are over 300 references to original publications, making the book a useful guide to the literature. The book does not seek to give data on ferromagnetic materials for specific engineering applications and, therefore, cannot be used as a data handbook for the design of electrical apparatus. Since it is written primarily from the standpoint of a physicist, metallurgical details are naturally somewhat meagre, so that the metallurgist will derive little assistance for the production of consistent and reproducible batches of magnetic alloys either on a laboratory or on a commercial scale. Nevertheless, the book can be confidently recommended to all metallurgists who wish to gain a real insight into the nature of magnetic phenomena as a background to their practical approach to the subject.

A. TAYLOR.

Investment Casting for Engineers. By Rawson L. Wood and Davidlee Von Ludwig. 9 × 6 in. Pp. ix + 477, with numerous illustrations. 1952. New York: Reinhold Publishing Corp. ($10.00); London: Chapman and Hall, Ltd. (80*s.*)

No extravagant claims are made in this book for the investment-casting process. The process is considered as a new foundry method with its own particular merits, and these merits are stated fairly in comparison with other casting techniques. The authors deprecate those claims to an order of accuracy or degree of surface finish which are so exaggerated that the term " precision casting " has been brought into disrepute.

A book addressed primarily to engineers will do a useful service if it can succeed in establishing " the elementary fact that investment casting is just another foundry process subject to most of the inevitable limitations attendant to the processing of liquid metals ". The chapters on design development and those factors which merit consideration when an investment casting is required to replace a fabricated component will also serve a most useful purpose to those interested in applying the process.

In general lay-out the book succeeds in taking the process step by step and establishing the vital importance of control at each one of them. There are many good photographs to support the well-laid-out text. American practice and materials are considered throughout, and the book includes no reference to British work, although, of course, certain American processes are operated by licensees in this country. It is admitted that the process is still in an active state of development and that changes in technique are inevitable. Perhaps undue emphasis is given to the Mercast process of producing ceramic shell moulds, and the two chapters on that subject tend to intrude into the general arrangement of the steps in the more usual investment-casting methods.

Due regard is given to the basic requirement of starting with a good die for the expendable pattern, and the available methods of die production are discussed. Thereafter follow the choice of the pattern material, the gating and clustering of patterns, and then through the actual investment to the final inspection of the castings. The heat-treatment of the castings and their machinability are also considered. All these operations receive, in general, a balanced consideration of the factors involved. No secret is made of the expensive nature of investment casting, but throughout the various stages the effect of the variable is often related to costs. The exception, so far as British practice is concerned, is that the relatively high price of ethyl silicate has accelerated the use of water-bond investment for high-temperature alloys. As well as the cost, the fire and explosion hazards of the ethyl silicate–alcohol mixtures should also have been pointed out.

Practically no alloy, ferrous or non-ferrous, has been developed specifically for investment casting. This book quotes some useful practical experience with materials in both fields. Their selection is dealt with adequately enough for most engineers. Metallurgically, some detail is lacking, and there are a few contradictions in the assessments of the American materials and practice considered, but the salient features all receive some attention. Control of grain-size is not so easy as the authors seem to suggest, nor is there any discussion of the effect of grain-size upon thermal shock resistance of the high-temperature alloys. The suggestion is wrong that cast materials have better fatigue-resistance than wrought.

The book will prove useful to all interested in either the actual investment-casting process or the possibilities of producing components by the method. Although further developments in technique are inevitable, this work will always provide a useful source of reference.

J. Hinde.

The Secondary Aluminium Industry in the U.S.A. (Report of Technical Assistance Mission No. 19). 27 × 21 cm. Pp. vii + 93, illustrated. 1952. Published for the Organization for European Economic Co-operation. Paris (6e): Presses Universitaires de France, 108 Boulevard Saint-Germain (350 fr.); London: H.M. Stationery Office. (7*s.*)

This Report is a record of a six-week visit made to the U.S.A. by a Team of 17 experts drawn from eight European countries, during the course of which 27 establishments connected in some way or other with the secondary aluminium industry were visited. The visit was made as a result of criticism by an American Mission to the effect that European countries could reduce demands on primary supplies by using more scrap locally. The timing of the visit was, perhaps, unfortunate, as there was in the autumn of 1950 an abnormal shortage of scrap in the U.S.A., and the Team had constantly to bear in mind this point. Taking account of this, the Team nevertheless concluded that the use of secondary metal in the United States is no more extensive than in Europe, except in the particular case of " Utility Sheet ". It was also concluded that few processes new to Europe were used, although individual members admitted learning a great deal in detail. Sentiments were recorded very similar to those noted by many Productivity Teams, namely, the fundamental difference between American and European economy based largely on the need to economize manpower in America and the need to save raw material in Europe.

After the introduction and summary there follow six chapters and two appendices giving the technical information obtained by the Mission. First, the secondary aluminium industry of the U.S.A. is surveyed in general terms, including bulk statistics of scrap in America and Europe, and a note on the organization of the industry. The second chapter deals with the processing of scrap, and includes brief notes on the types of melting furnace in use, fluxes, and the operations of casting ingots. A chapter on the production of castings is divided into three main parts, sand, permanent-mould, and pressure-die, and in each portion may be found useful items of scientific and technological information, illustrated with numerous line drawings and including notes on mould design, pouring methods, and the machinery employed. Next is a short chapter dealing with the production of billets and ingots for the manufacture of wrought products, and this is followed by a two-part chapter detailing the types of alloys, their specifications, and applications. The last portion is, of necessity, extremely compressed and on that account not particularly useful. The final chapter covers methods of production control, with a note on research which includes a description of a visit to an aluminium producer's laboratories.

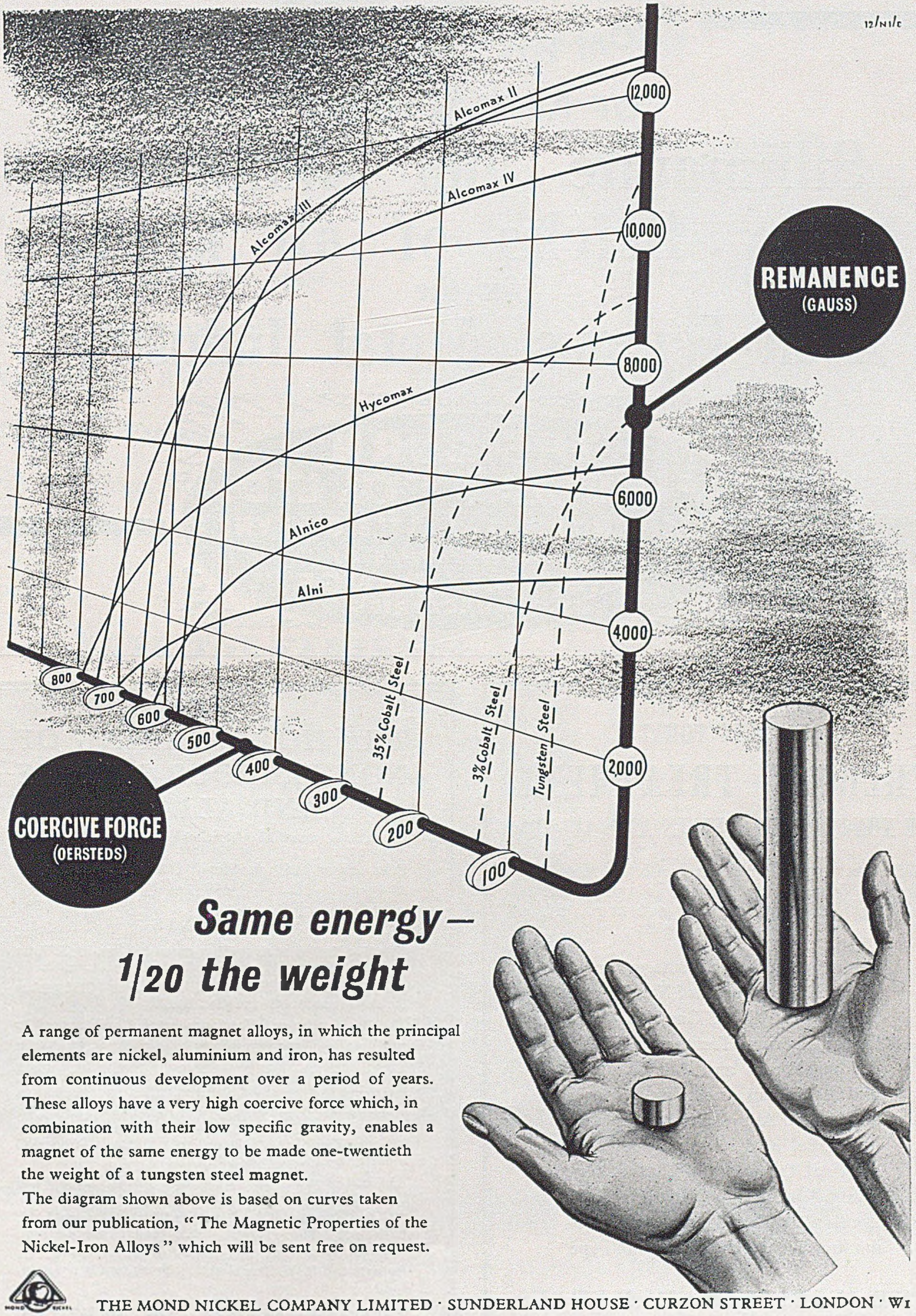
A particular item which the Report mentions as being specifically different from European practice is in the use of secondary metal for the so-called " Utility Sheet ", which is used as cladding for roofs and walls of American buildings. This material consists of re-melted scrap, chiefly from aircraft, clad with an aluminium alloy containing 1% zinc. This use of scrap is not, however, novel to the U.K. aluminium industry, as rather similar sheet was used here for prefabricated aluminium bungalows. This had a core of aluminium alloy similar in composition to D.T.D. 479—a war-time specification for secondary aluminium alloy ingot—but in this case the core was clad with a coating of pure aluminium. This use appeared to be new to a majority of the Team.

It will be correct to assume that the condensation which appears inevitable in a Report of this kind reduces its value, but all connected with the supply side of the aluminium industry should have a copy available for study and reference.

E. G. West.

METALLURGICAL ABSTRACTS

METALLURGICAL ABSTRACTS AUTHOR INDEX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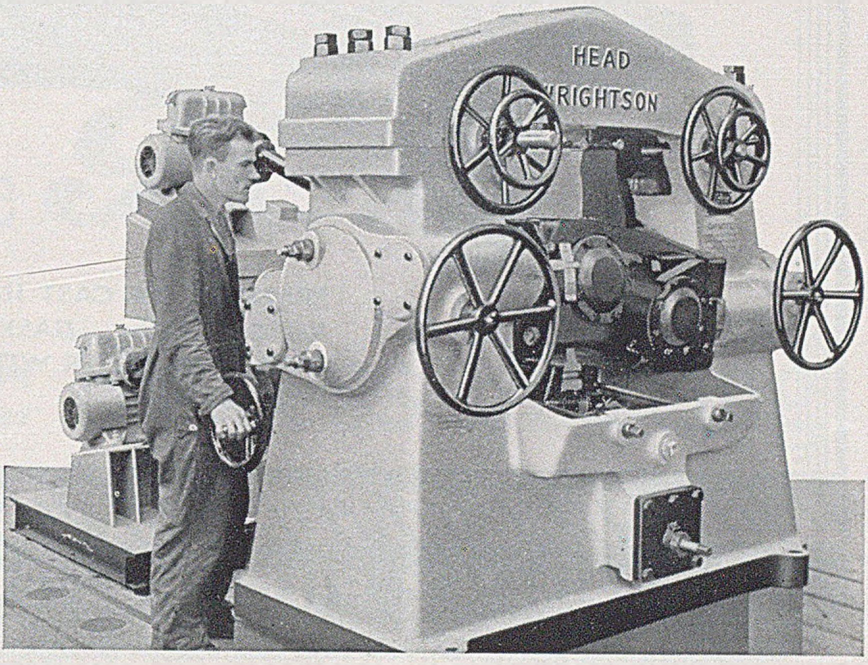
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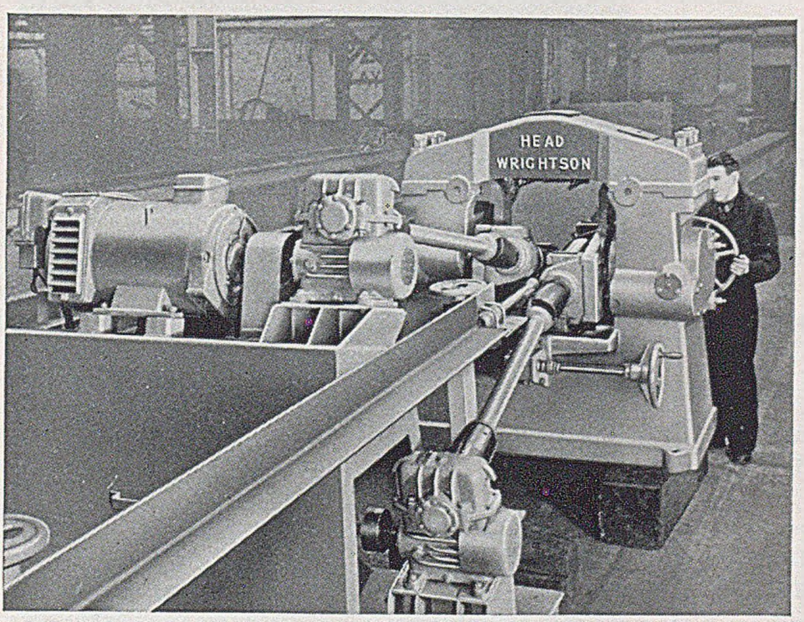
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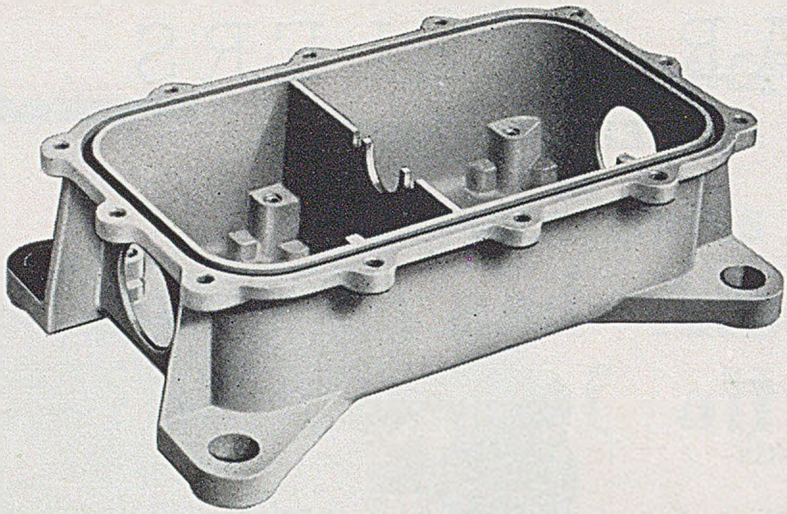

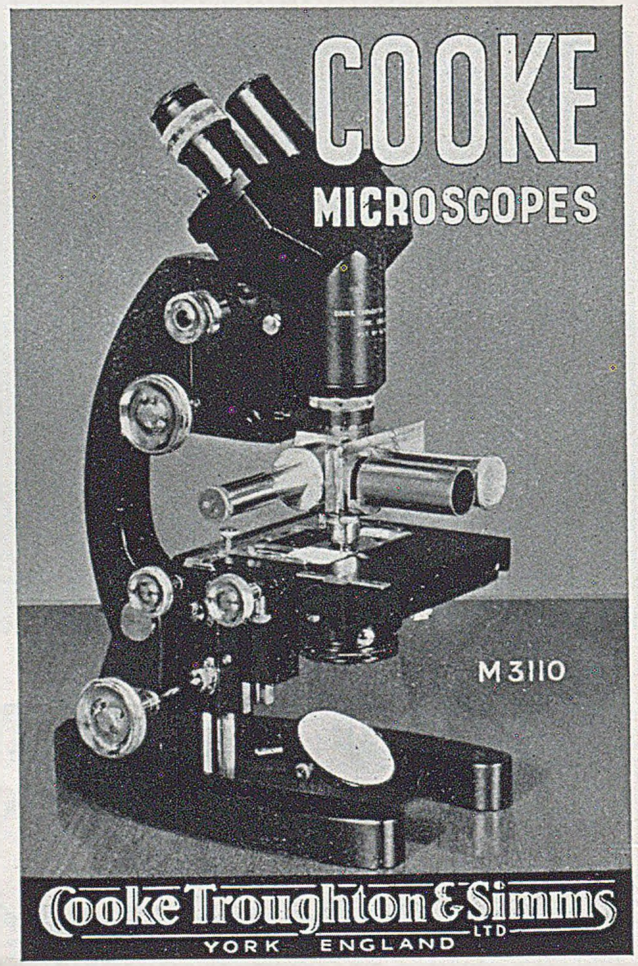

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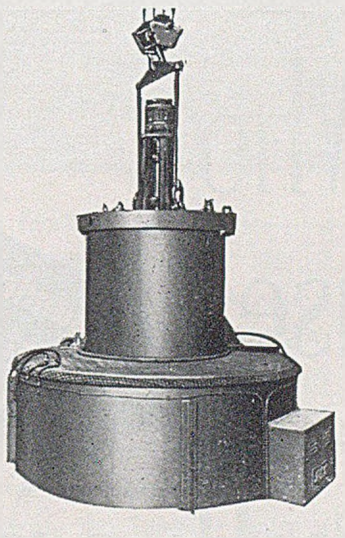

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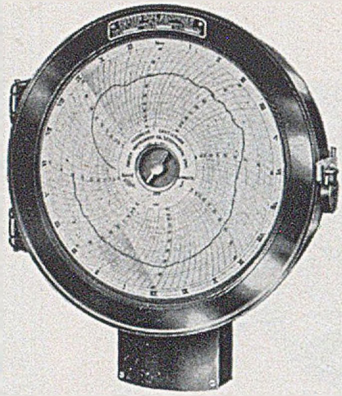

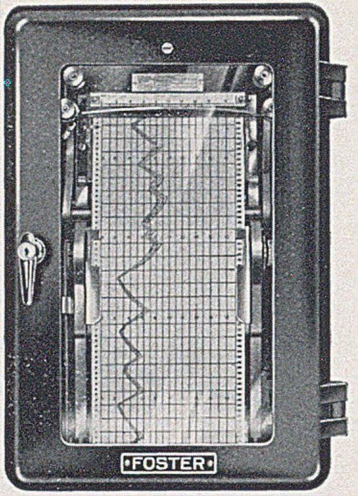
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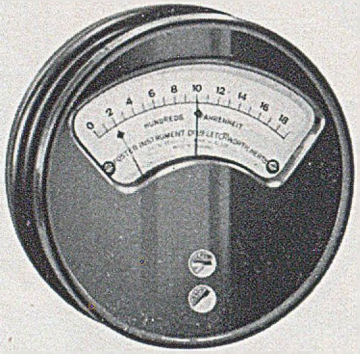

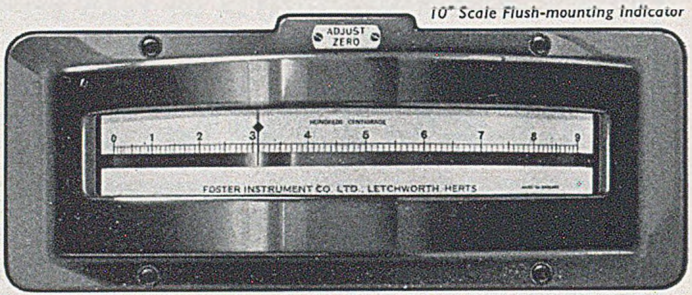
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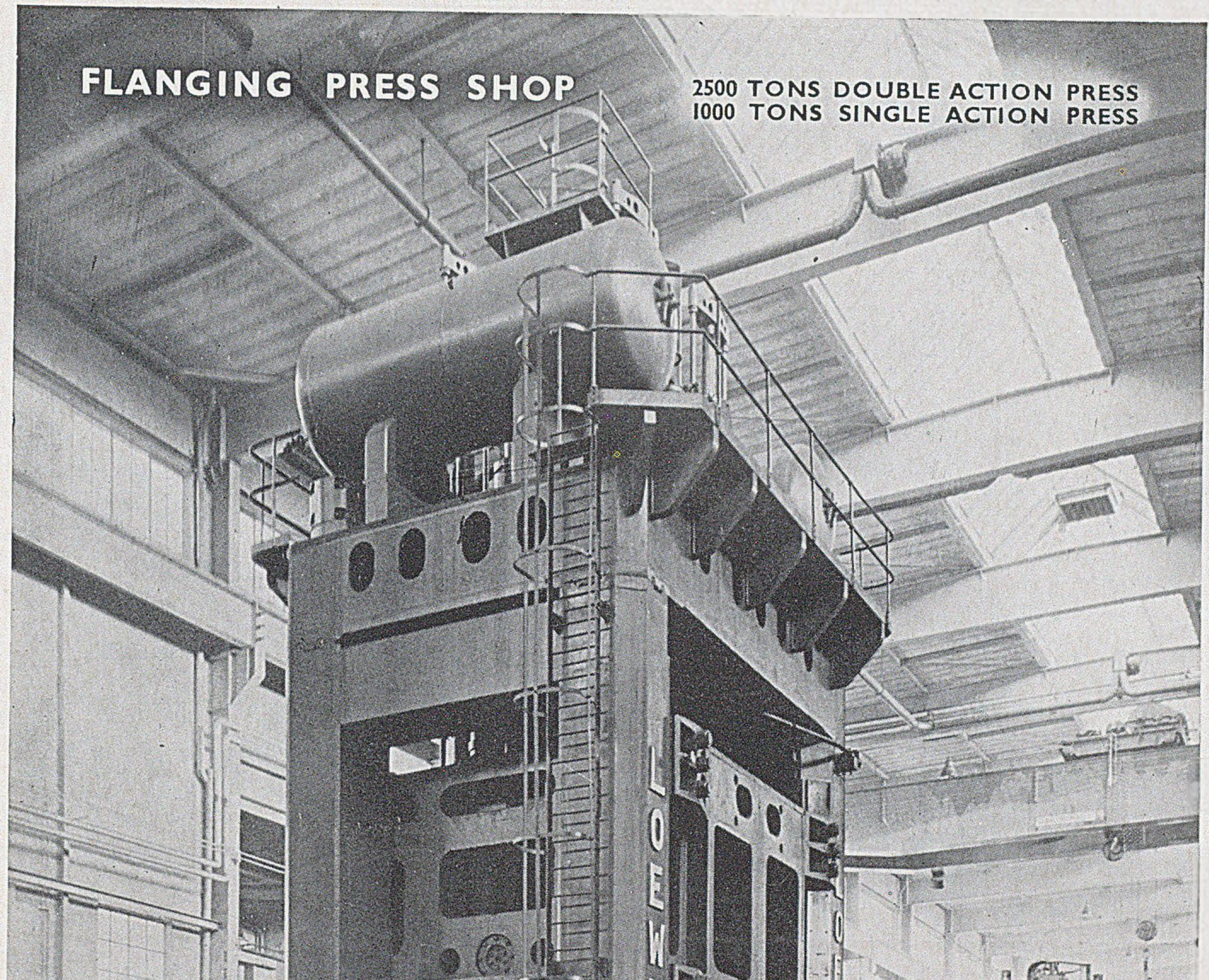
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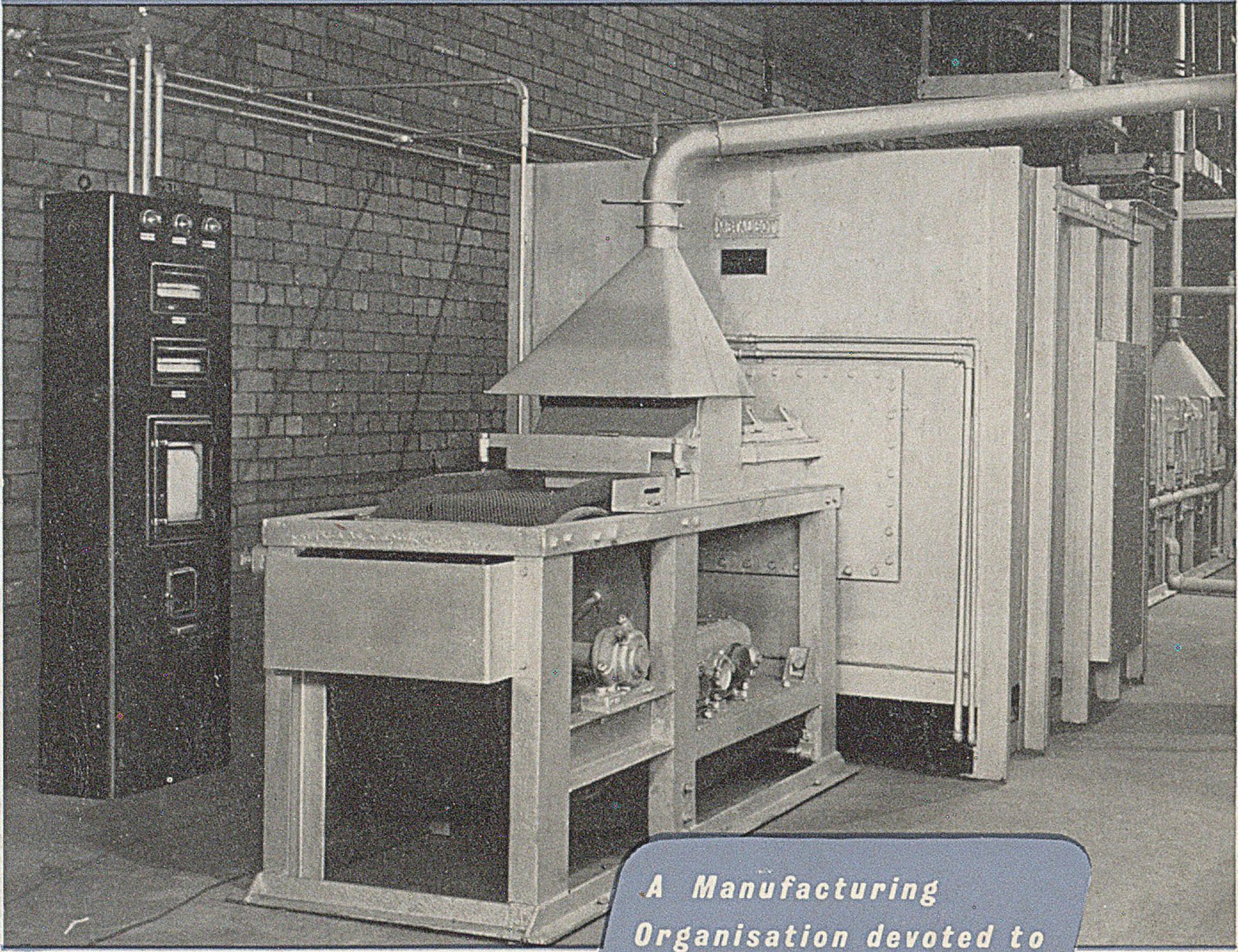

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