THE JOURNAL OF THE INSTITUTE OF Metals

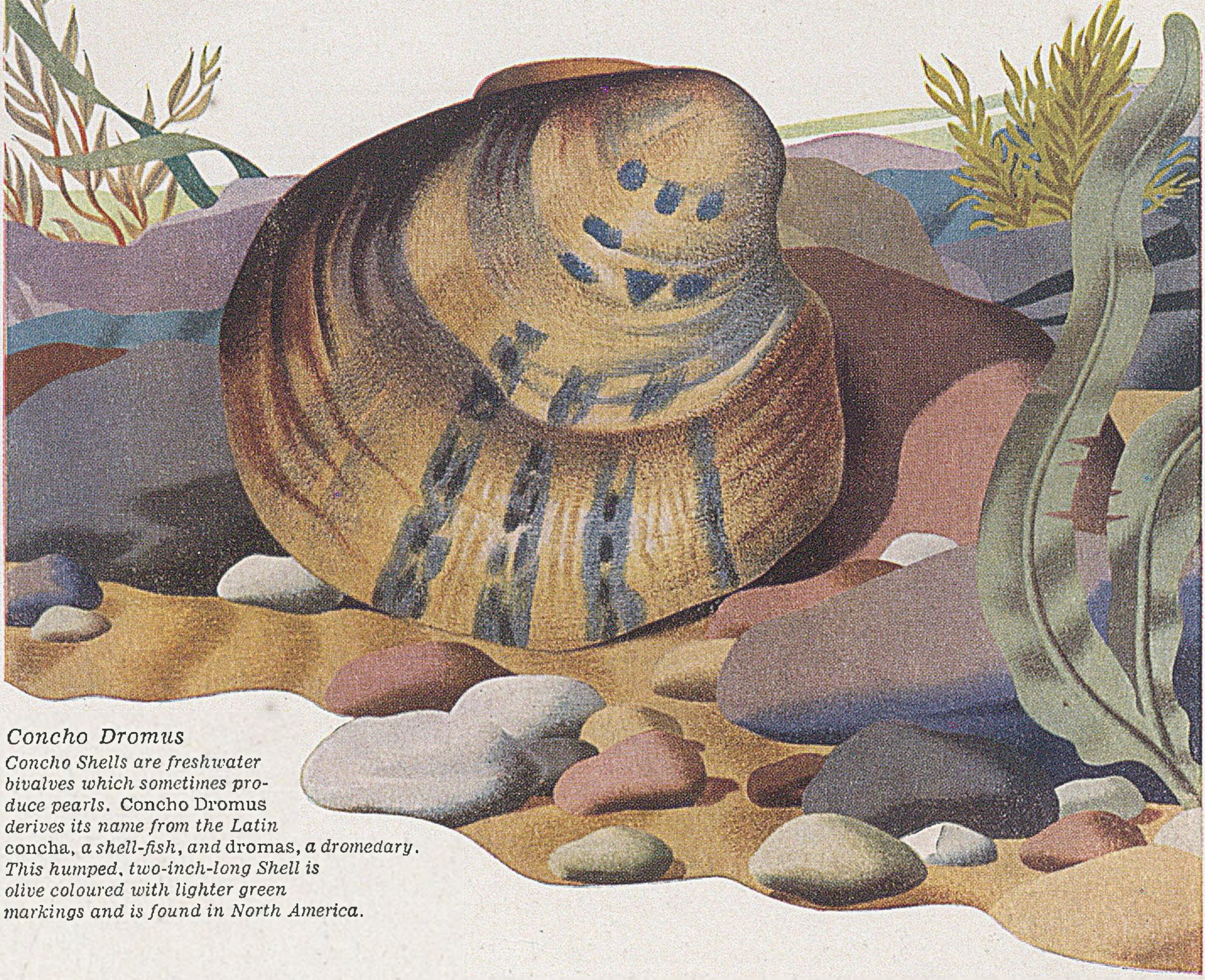
WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents:

PRICE: SEVEN SHILLINGS AND SIXPENCE

NOVEMBER 1951

Do your machining with this seashell



Concho Dromus

Concho Shells are freshwater bivalves which sometimes produce pearls. Concho Dromus derives its name from the Latin concha, *a shell-fish, and* dromas, *a dromedary. This humped, two-inch-long Shell is olive coloured with lighter green markings and is found in North America.*

SHELL DROMUS OILS are high grade soluble cutting oils incorporating efficient emulsifiers, which give stable emulsions showing a high degree of oil dispersion. The emulsions are suitable for the milder machining operations where cooling is more important than lubrication. These oils are available in opaque and clear grades, which give milky and translucent emulsions respectively.

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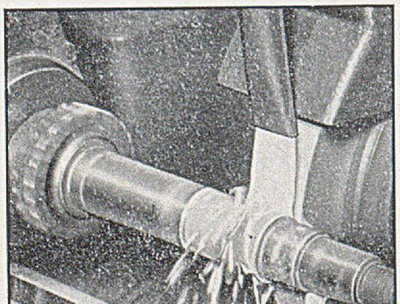


In common with all Shell Oils, Seashell Oils are produced as the result of world-wide research. This picture illustrates the use of the cutting fluid in cylindrical grinding.

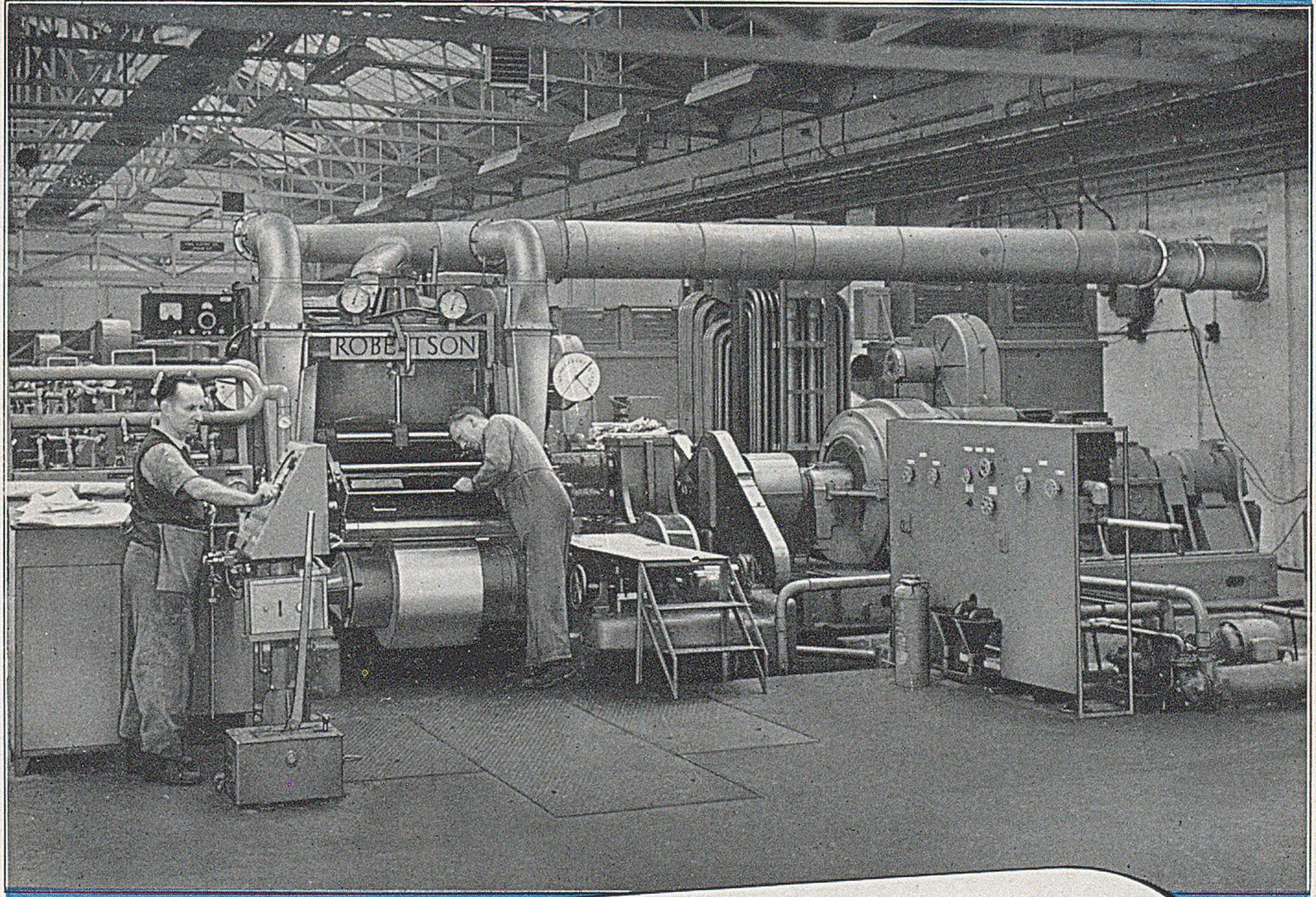
Shell *Dromus* Oil

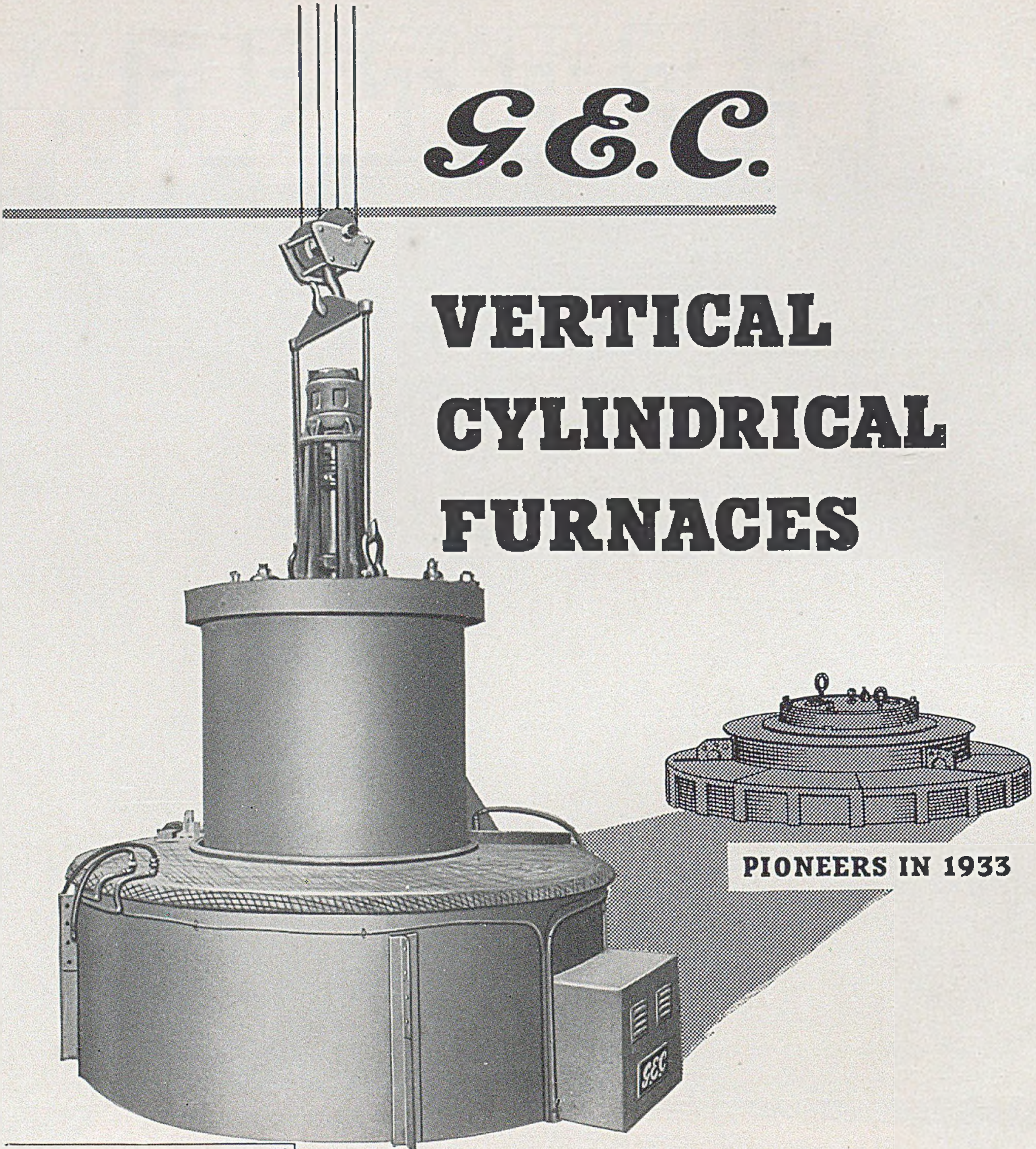
SHELL

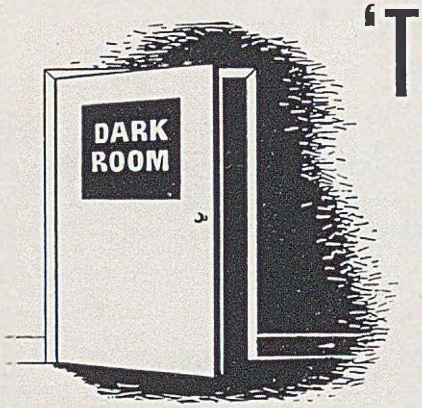
LEADERSHIP IN LUBRICATION

Inside front cover

November 1951


ROBERTSON





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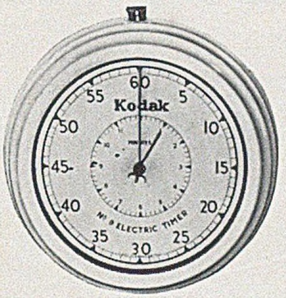

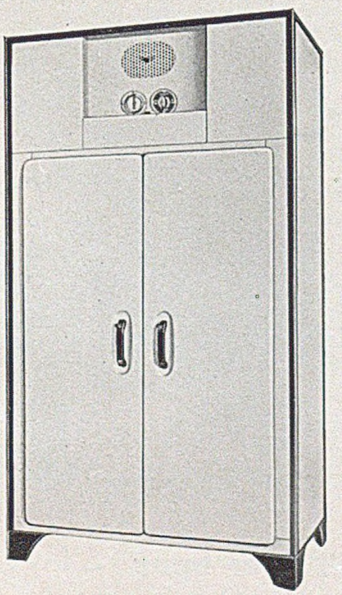

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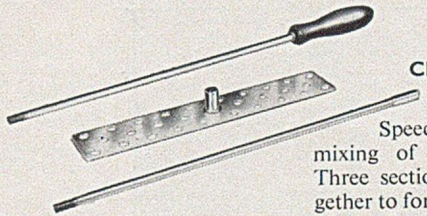

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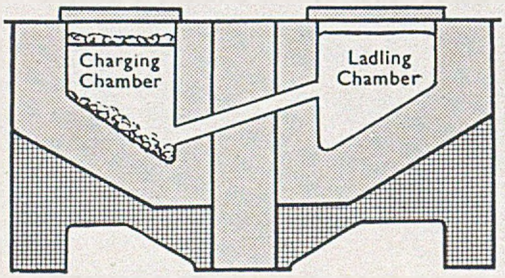
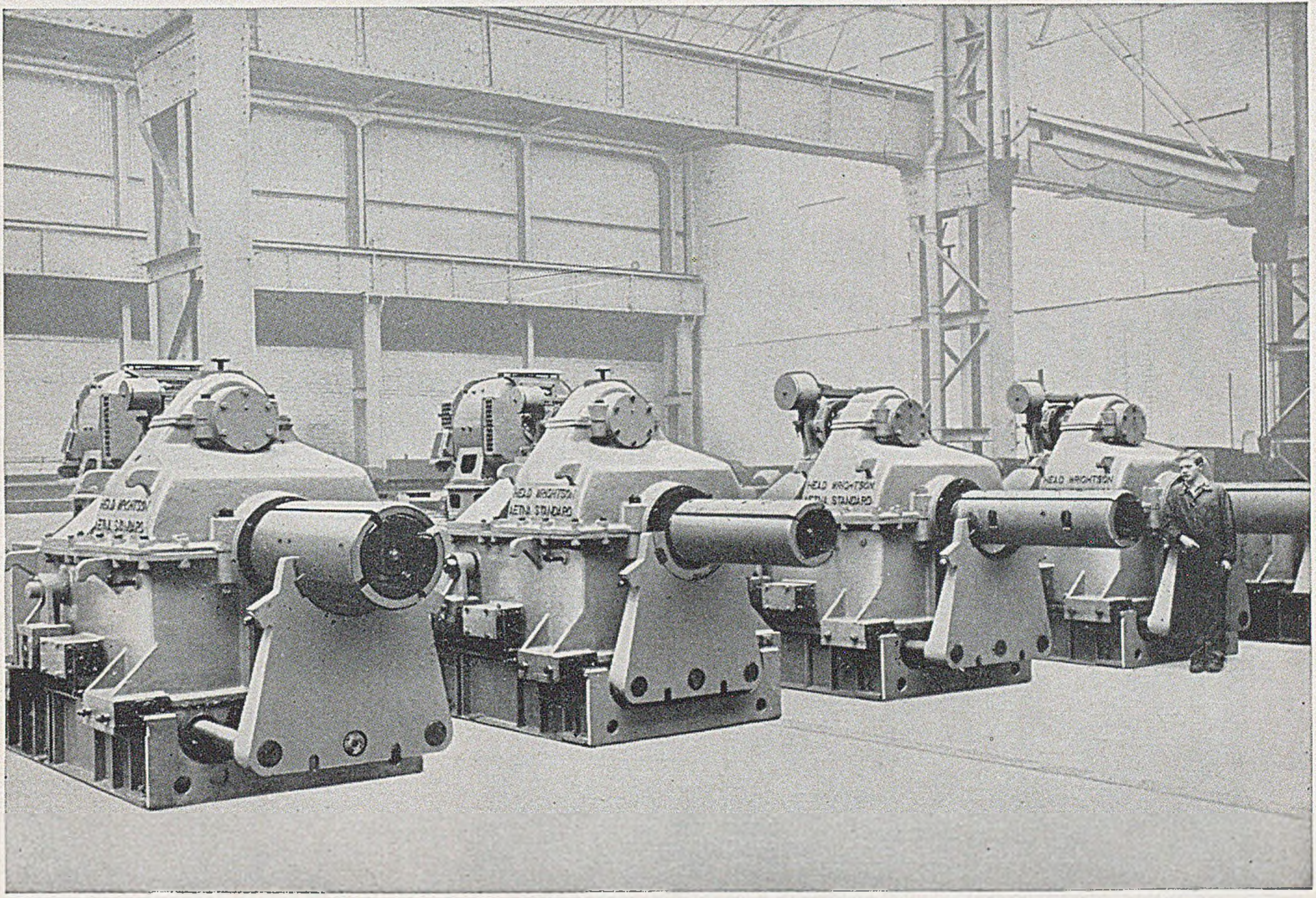


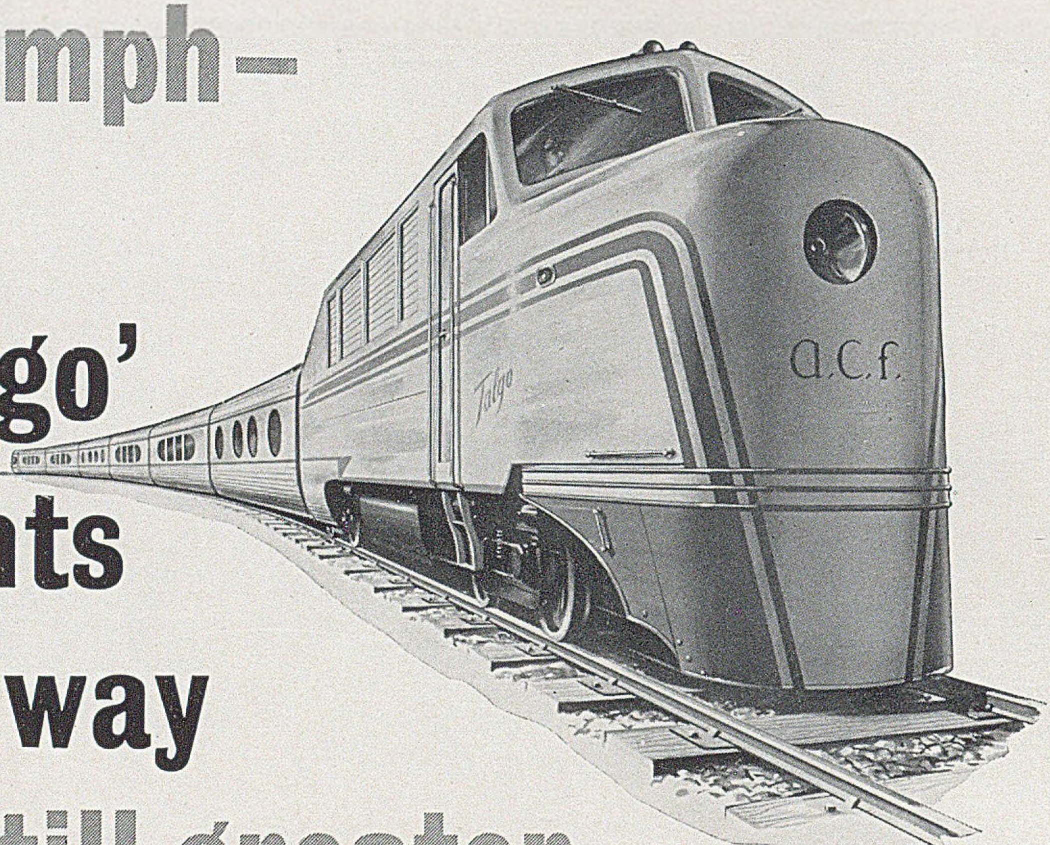
Charging Chamber
Ladling Chamber

EFCO
Lindberg



HEAD WRIGHTSON
AETNA STANDARD

an engineering triumph— the 'Talgo' points the way to still greater developments...



The "Talgo"
The new aluminium train
designed for service in Spain.

Nowadays we have hardly time to marvel at our own achievements before some new, some greater project is under way. Progress is inevitable. Just as inevitable as the fact that aluminium is part of this process of progress. It is happening all the time, and all the time aluminium is enabling it to happen. Engineers, technicians and designers are finding that the metal that helps them most is aluminium.

Progress has brought ever-increasing new uses of aluminium. Keeping pace with these developments requires adequate supplies of this light metal, produced in the most economical manner through an integration of the many stages of production from the mining of bauxite, shipping of raw materials and generating of hydro-electric power, to the ultimate extraction and fabrication of the metal—all assisted by research, technical and commercial services. This co-ordination is achieved through Aluminium Limited, the Canadian Parent Company, whose subsidiaries are located chiefly in the British Commonwealth.

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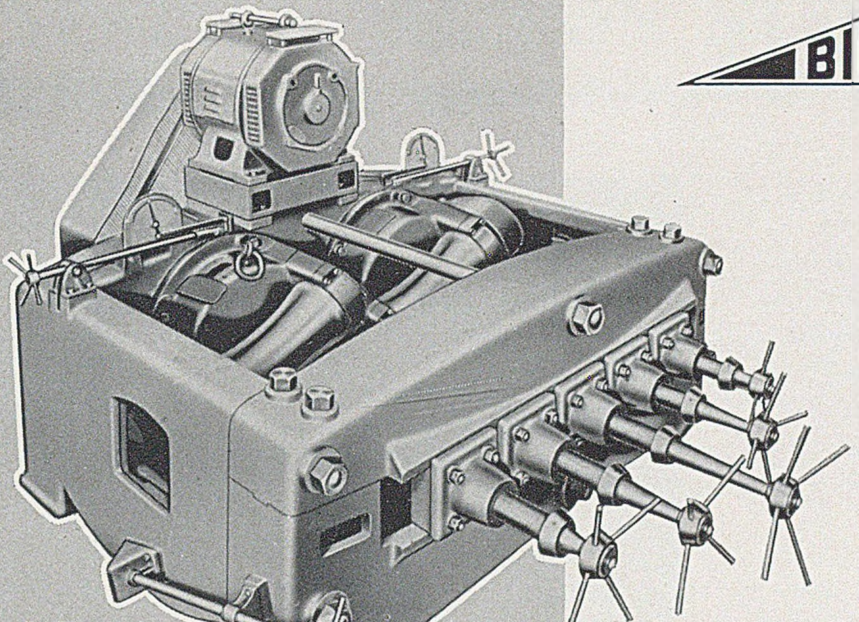

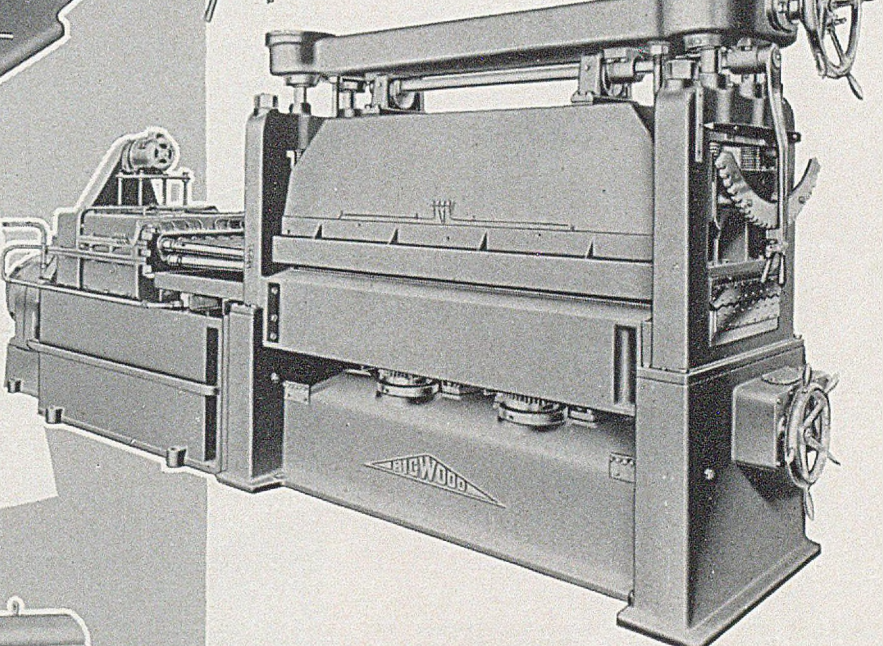
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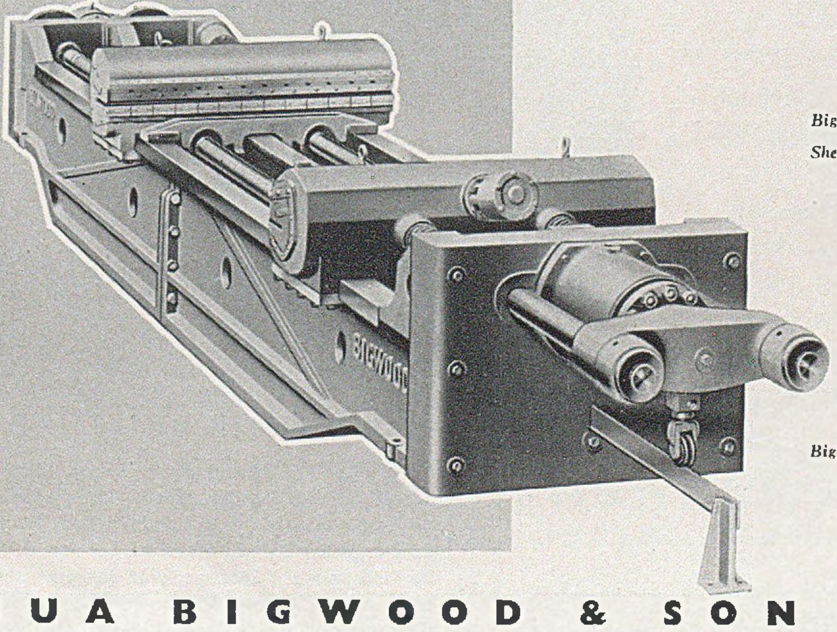
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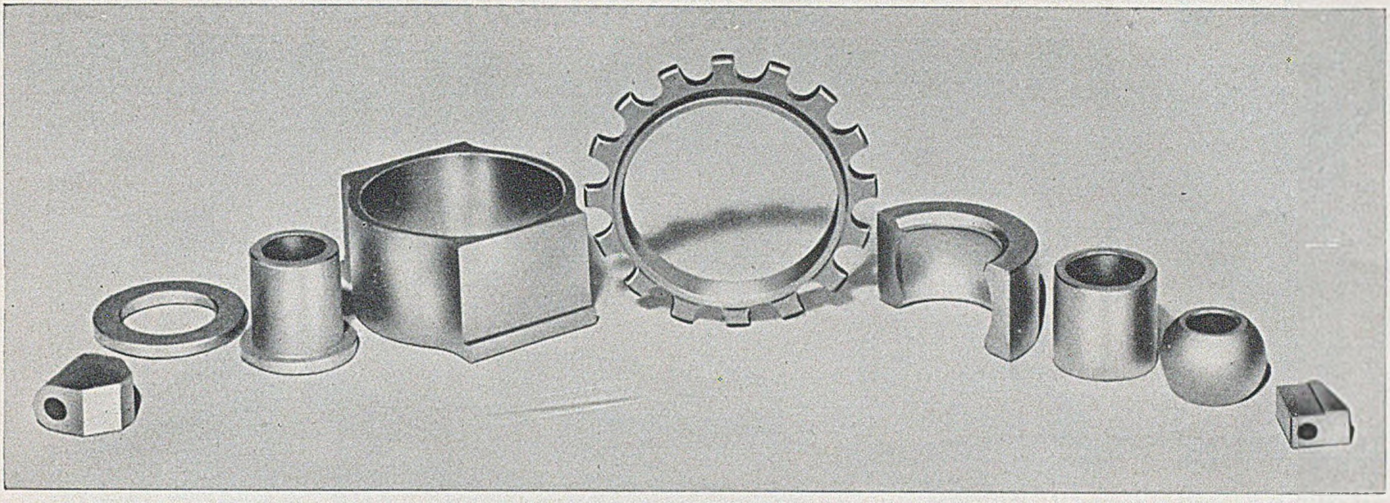

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Sintered metal filters

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Electronic High-frequency

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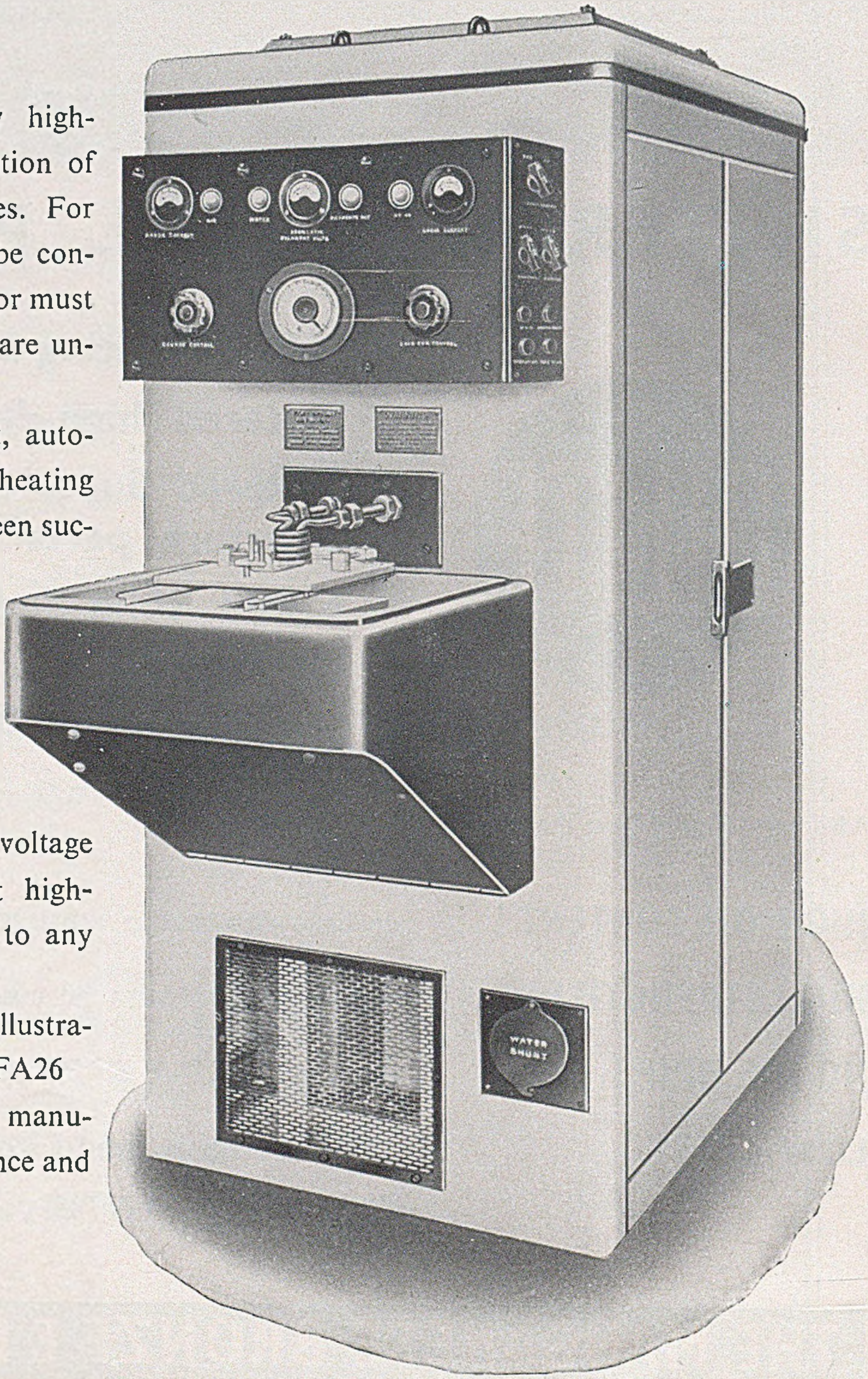
The equipment provides a fast, automatic, and accurate method of heating relatively small parts, and has been successfully applied to annealing, tin-flowing, bolt-head upsetting, surface-hardening, out-gassing the electrodes of valves, etc. The BTH Current Limit Control nullifies the effects of variations in supply voltage and enables maximum output high-frequency current to be preset to any desired value.

Available in sizes up to 25 kW. Illustration shows a 10 kW unit Type FA26

Other types of industrial heaters manufactured by BTH include resistance and dielectric heating equipment.



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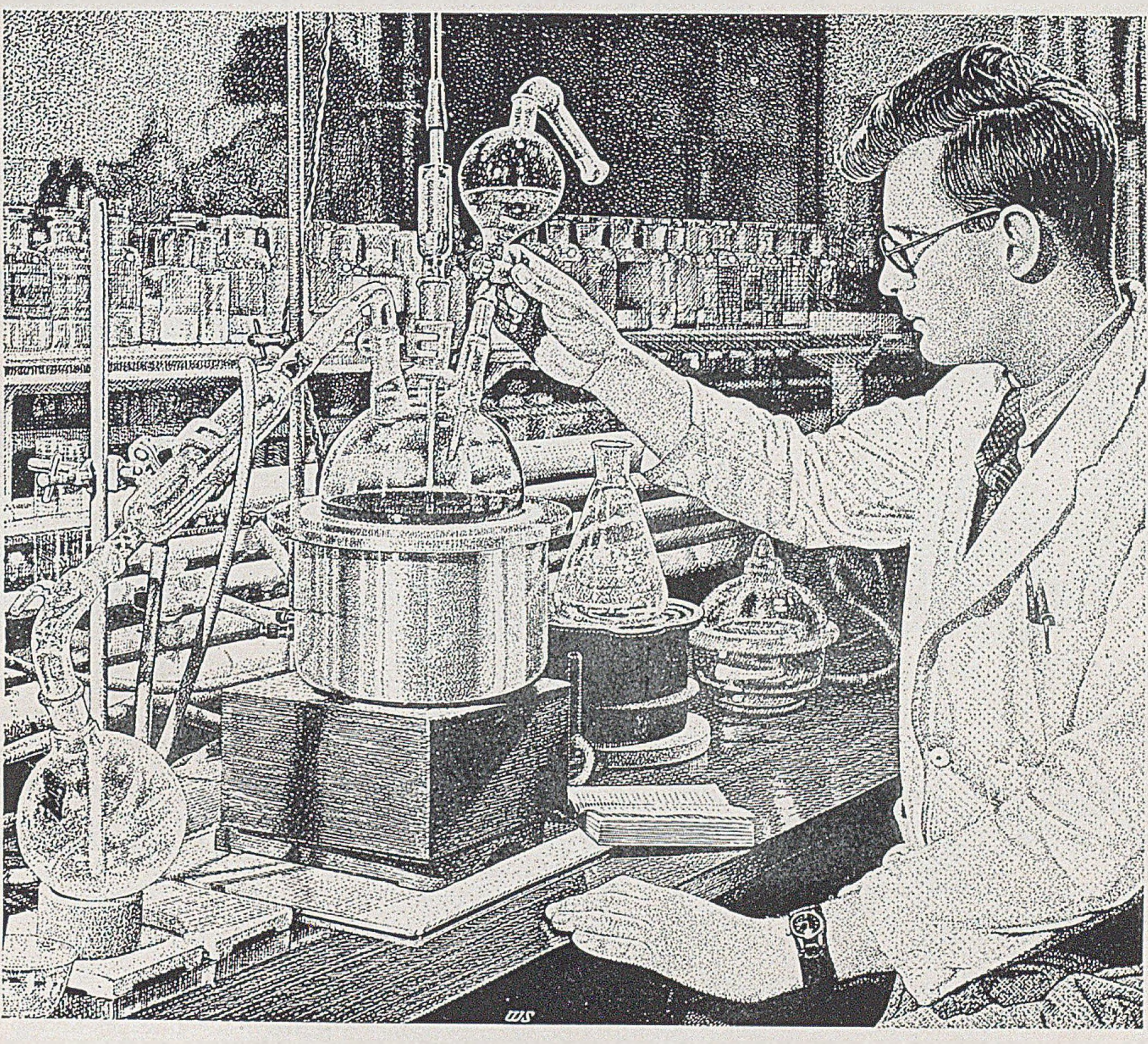
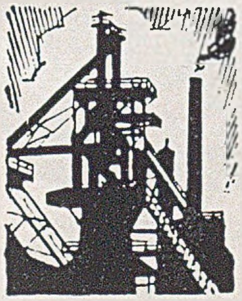
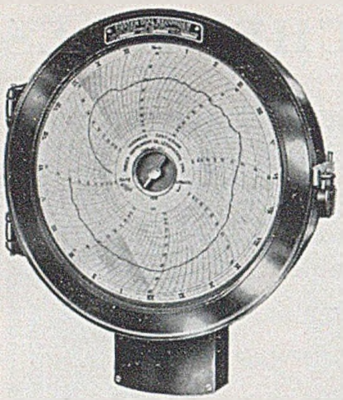




FOSTER

PYROMETERS



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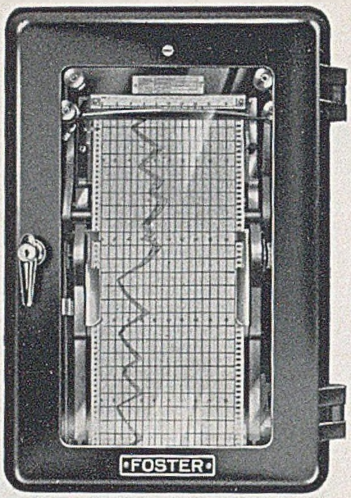

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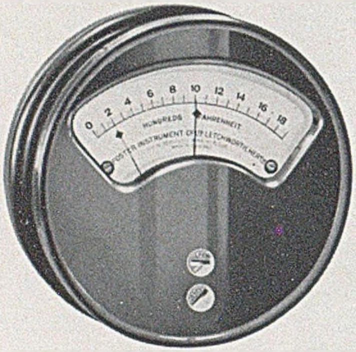

★ We also make the FOSTER INTROSCOPE for internal inspection of inaccessible parts, including pipe welds, diesel jets, etc.

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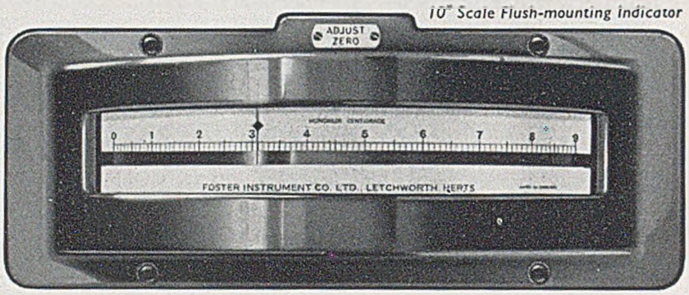

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While we appreciate the necessity to recover every pound of the scarce non-ferrous metals, we cannot refrain from reminding users of high quality zinc alloy pressure die castings that the fact still remains that such castings can ONLY be produced by using alloys which conform strictly to B.S.S.1004.

The present tendency to bring into use doubtful alloys makes it more necessary than ever for purchasers of zinc alloy pressure die castings to ensure that the pressure castings which they buy do in fact strictly conform to B.S.S.1004.

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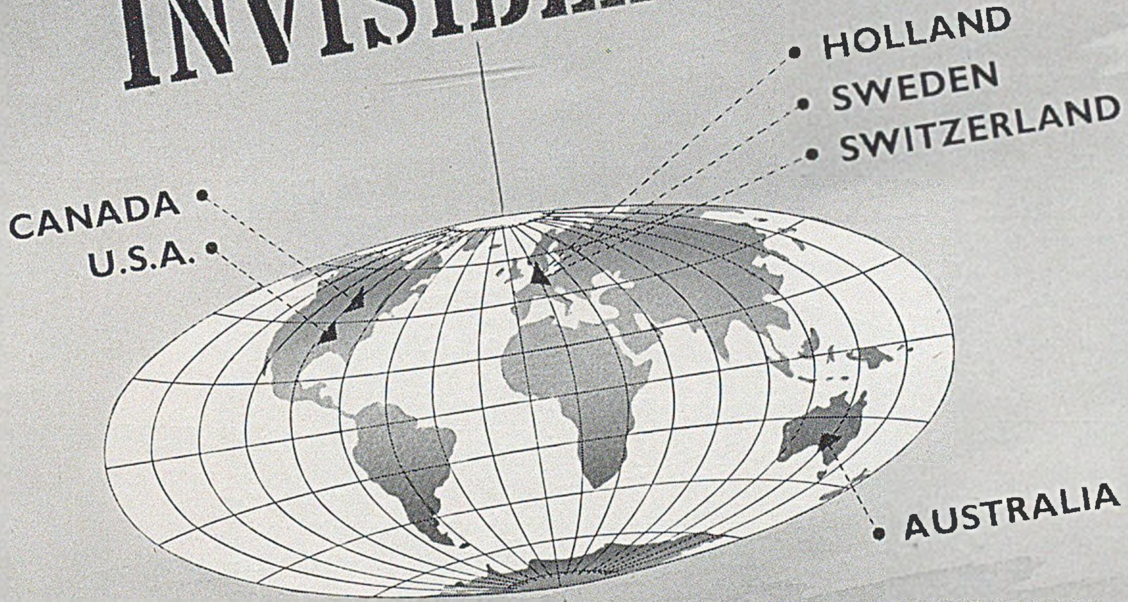
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RADIOLOGICAL EXAMINATIONS

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

and "X-ray quality" castings

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

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

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THE NON-DESTRUCTIVE TESTING OF METALS

By R. F. HANSTOCK, Ph.D., F.Inst.P.

(*Institute of Metals Monograph and Report Series, No. 10*)

Cloth, 170 pp., with 72 illustrations

Price 21s. or $3.50, post free

In this specially commissioned monograph, Dr. Hanstock, who is Chief Physicist to High-Duty Alloys, Ltd., has made a comprehensive survey of the various methods that are available for the non-destructive testing of metals, ranging from liquid-penetration methods of crack detection to the use of X-ray diffraction for the determination of grain-size and orientation and for the measurement of internal stresses. Special attention is devoted to damping-capacity measurements as a means of non-destructive testing, a field in which the author himself has worked extensively in recent years.

Throughout, emphasis is laid on the physical bases of the methods described, with specific examples of their application in many cases. The use of the book is facilitated by an Appendix in which are given in tabular form the properties that may be assessed by non-destructive means, the various methods applicable in each case, and brief comments on their suitability and limitations.

CONTENTS

Measurement of Thickness; Evaluation of Surface Finish; Detection of Cracks; Radiography; Ultrasonics; Dynamic Tests; Damping-Capacity Measurements; X-ray Diffraction, Electrical and Magnetic Methods; Methods of Analysis.

THE INSTITUTE OF METALS

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ZINC ALLOY DIE CASTING

makes all the difference!

The simple brass bulb horn was essential to early motorists, who gave it a proud position on their machines; and Africans valued them so much that they hoarded them instead of money! But ideas changed, and modern horns are so complex that they are only economical when they are zinc alloy die cast.

The windtone horn shown above is one of a pair that sound in harmony. Each is a good example of an intricate geometrical pattern which has been faithfully reproduced by die casting. The two parts of each horn are held tightly together by integral rivets on one engaging with cored holes on the other.

Some facts
about zinc alloy die casting

Speed of production is an outstanding feature of the die casting process — the shortest distance between raw material and finished product. Zinc alloys are the most widely used of all metals for die casting because they yield castings with the following qualities:

ACCURACY: Castings can be made practically to finished dimensions and need little or no machining.

STRENGTH: Good mechanical properties for stressed components.

STABILITY: Close tolerances are maintained throughout the life of the casting.

British Standard 1004

It is essential that alloys conforming to B.S.1004 should be specified for all applications.

The Association welcomes enquiries about the use of zinc alloy die castings. Publications and a list of Members are available on request.

ZADCA

ZINC ALLOY DIE CASTERS ASSOCIATION
LINCOLN HOUSE, TURL ST., OXFORD

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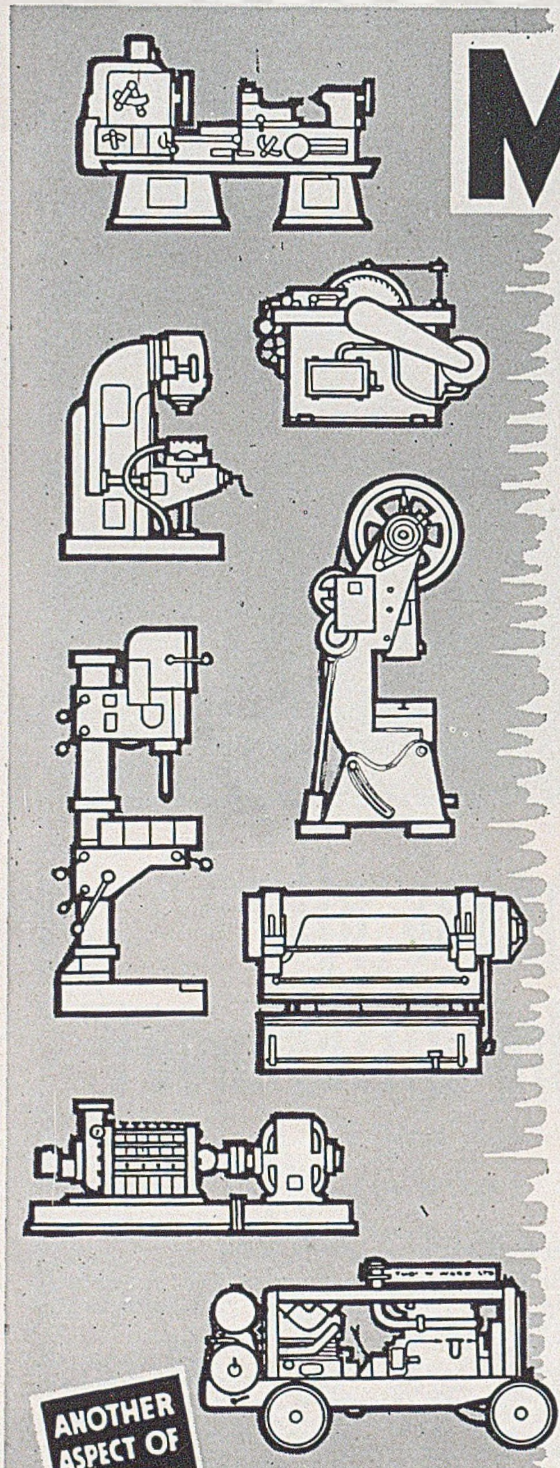

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1951, PART 3 PUBLISHED MONTHLY NOVEMBER

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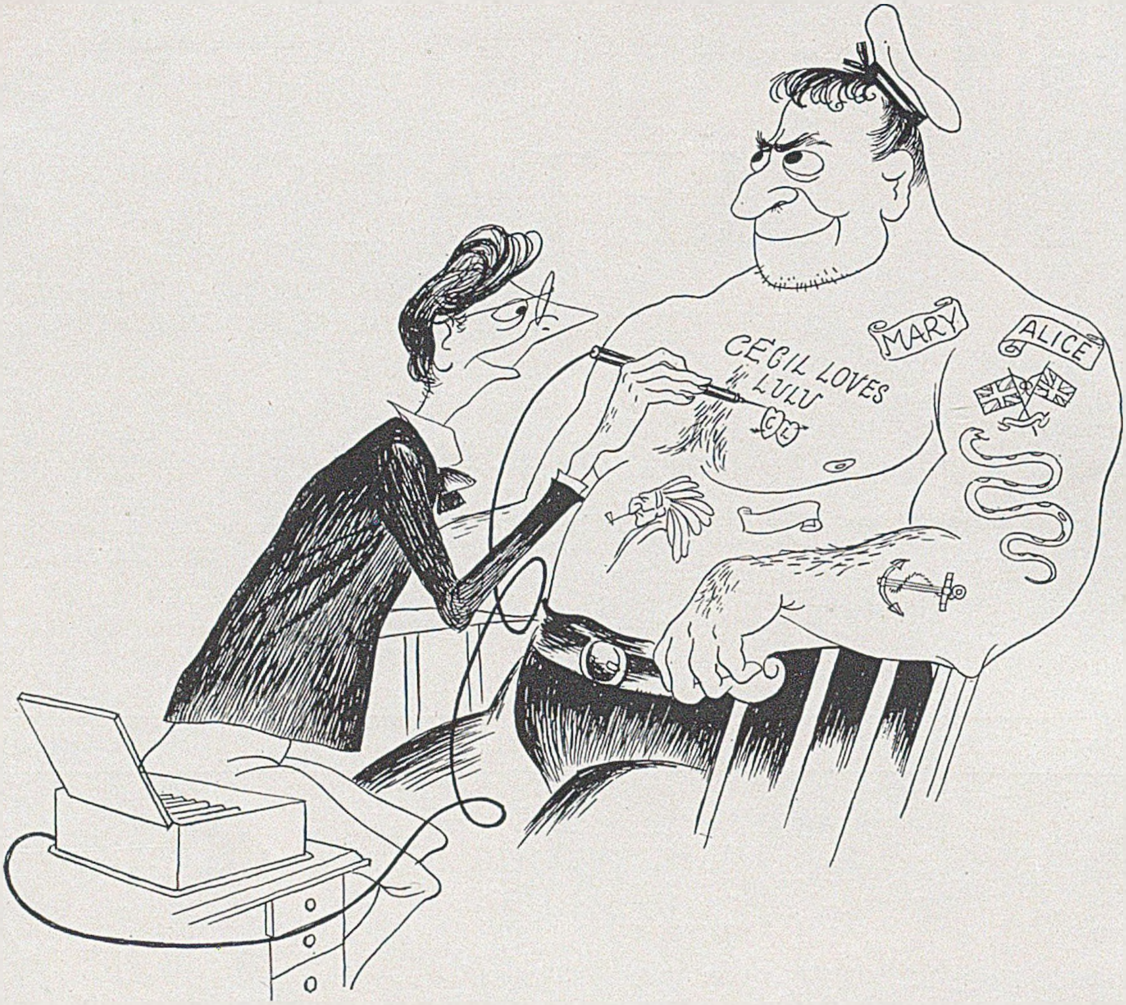
JOURNAL

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


CECIL LOVES LULU
MARY
ALICE

BULLETIN
OF THE INSTITUTE OF METALS

VOLUME I NOVEMBER 1951 PART 3

PERSONAL NOTES

Professor J. H. Andrew has been made an Honorary Vice-President of the Iron and Steel Institute.

Dr. Bani R. Banerjee has been appointed a Research Metallurgist in the Engineering Research Department of the Standard Oil Co., Chicago.

Mr. M. W. Barlow has resigned his position as Sales Manager of British Electro Metallurgical Co., Ltd., Sheffield, to join Foundry Services, Ltd., Birmingham, as Manager of their newly established Ferro Alloy Division.

Mr. E. H. Bucknall, of The Mond Nickel Co., Ltd., has recently spent some time in Washington as one of the British members of the Joint Sub-Committee (of the International Materials Conference) on Utilization of Manganese, Nickel, Cobalt, Tungsten, and Molybdenum. Mr. Bucknall was Chairman of the Group that drafted the Sub-Committee's report.

Mr. R. Butler has left the B.S.A. Group Research Centre, Sheffield, and is now in the Research Department of The Birmingham Small Arms Co., Ltd., Small Heath, Birmingham 11.

Mr. R. Caden has been appointed Officer in Charge, Metallurgical Laboratory, Engineering Department, H.M. Dockyard, Rosyth, Scotland.

Mr. A. Crosby has been awarded the B.Sc. degree of London University, with 2nd class Honours in Metallurgy.

Mr. J. K. Davies has recently left Magnesium Elektron, Ltd., and taken a post in the Experimental Department of The British Aluminium Co., Ltd., Falkirk.

Dr. D. W. Davison has joined the staff of T. S. Skillman and Co. Pty., Ltd., North Sydney, N.S.W.

Mr. A. P. Greenough has left the Royal Aircraft Establishment, Farnborough, to take up an appointment as Assistant Lecturer in Metallurgy at the University College, Swansea. He was recently awarded the M.A. degree of Cambridge University.

Mr. E. G. Hall is an Assistant Experimental Officer at the Armament Research Establishment, Woolwich, S.E.18.

Mr. Vernon Harbord has been elected President of the Institution of Mining and Metallurgy for the year 1952–53 and will take office at the Annual General Meeting to be held on 15 May 1952. He is a member of the well-known firm of consulting metallurgists Messrs. Riley, Harbord, and Law.

Mr. D. W. Hopkins has resigned his lecturership in the University College of Swansea to becаme Works Manager of Western Metallurgical Industries, Ltd., Neath, Glam.

Dr. Hsun Hu has completed his studies at Notre Dame University, and is now engaged as a metallurgist at the Institute for the Study of Metals, Chicago.

Mr. Frederick Jackson is now a metallurgist in the Newark Works of Kaiser Aluminum and Chemical Corp., at Newark, O., U.S.A.

Mr. S. R. A. Langford has left James Booth and Co., Ltd., and is now a metallurgist with J. Stone and Co. (Charlton), Ltd.

Mr. W. J. Lee-Bird has been appointed a metallurgist with Smiths Stamping Works, Ltd., Coventry.

Mr. S. E. Mills has graduated B.Met. at Sheffield University and is now metallurgist to the Bifurcated and Tubular Rivet Co., Ltd., Aylesbury, Bucks.

Mr. R. N. Parkins has recently been awarded the Ph.D. degree of the University of Durham.

Mr. E. R. Perry has taken up an appointment in the Metallurgy Group of the Research Laboratories of the General Electric Co., Ltd., Wembley.

Mr. S. F. Pugh has been appointed a Senior Scientific Officer in the Metallurgy Division of the Atomic Energy Research Establishment, Harwell.

Dr. W. I. Pumphrey, having returned from his visit to the United States as a Commonwealth Fellow, has been appointed Research Manager to Murex Welding Processes, Ltd., Waltham Cross, Herts.

Mr. R. W. Robinson has now left the Central Metallurgical Laboratory, Emsworth, and has taken up a new Admiralty appointment in H.M. Dockyard, Devonport.

Mr. B. G. Street has been awarded the Ph.D. degree of Liverpool University and has been appointed an assistant lecturer in the Metallurgy Department of that University.

Mr. J. R. Theobald is on a training course in production technology for about a year in the United States.

Mr. J. S. Thompson has been appointed Chief Metallurgist to Durham Chemicals, Ltd., Birtley, Co. Durham.

Mr. D. K. Worn has left the Research Laboratories of the General Electric Co., Ltd., Wembley, to take up an appointment in the Metallurgy Department, Nelson Research Laboratories, The English Electric Co., Ltd., Stafford.

Obituary

The Editor regrets to announce the death of:

Mr. Alfred Allcock, Chairman and Governing Director of Allcock and Co. (Metals), Ltd., Hockley Hill, Birmingham, on 20 September. Mr. Allcock was also Chairman of Barr Street Castings, Ltd., Birmingham, and a Founder Member and later President of the British Bronze and Brass Ingot Manufacturers' Association. He had been a member of the Institute since 1928.

JOINT ACTIVITIES

Beilby Memorial Awards

From the interest derived from the invested capital of the Sir George Beilby Memorial Fund, at intervals to be determined by the Administrators representing the Royal Institute of Chemistry, the Society of Chemical Industry, and the Institute of Metals, awards are made to British investigators in science to mark appreciation of records of distinguished work. Preference is given to investigations relating to the special interests of Sir George Beilby, including problems connected with fuel economy, chemical engineering, and metallurgy, and awards are made, not on the result of any competition, but in recognition of continuous work of exceptional merit, bearing evidence of distinct advancement in science and practice.

In general, awards are not applicable to workers of established repute, but are granted as an encouragement to younger men who have done original independent work of exceptional merit over a period of years.

The Administrators are empowered to make more than one award in a given year if work of sufficient merit by several candidates is brought to their notice. In 1950 two awards, each of one hundred guineas, were made to Mr. W. A. Baker and Dr. G. Whittingham.

Consideration will be given to the making of an award or awards from the Fund early in 1952, and the Administrators will therefore be glad to have their attention drawn to outstanding work of the nature indicated, not later than 31 December 1951.

All communications on this subject should be addressed to the Convener of the Administrators, Sir George Beilby Memorial Fund, Royal Institute of Chemistry, 30 Russell Square, London, W.C.1.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

Manchester Metallurgical Society

Officers for 1951–52

President: A. B. Ashton, M.Sc., F.I.M.

Ex-President: E. J. Heeley, Assoc. Met., F.I.M.

Vice-Presidents: H. Allison, B.Sc. Tech., F.I.M.; W. B. Wragge, B.Sc. Tech., A.R.I.C.; J. D. Hannah, M.Sc., F.I.M.

Council: R. S. Brown, M.B.E., F.I.M.; C. J. Bushrod, M.Sc.; G. A. Cottell, M.Sc., A.I.M.; E. Holland, B.Sc., A.I.M.; K. M. Entwistle, M.Sc., Ph.D.; E. L. Francis, M.Sc.; F. H. Poole, A.M.C.T.; R. Wallace; W. H. N. Whyman, B.Sc.

Hon. Secretary: J. A. Tod, B.Sc., F.I.M.

Hon. Treasurer: N. Youatt, A.M.I.Mech.E.

OTHER NEWS

Corrosion of Buried Metals

A new standard form for recording corrosion failures of buried pipes has been prepared by the British Iron and Steel Research Association's Sub-Committee on the Corrosion of Buried Metals. It is available from Mr. E. E. White, B.I.S.R.A., 140 Battersea Park Road, London, S.W.11, and is to be returned when completed to the Chemical Research Laboratory of the D.S.I.R., Teddington. The Sub-Committee has taken over from the Research Coordination Committee of the Institution of Water Engineers its work on the corrosion of buried pipe-lines, and the new form incorporates the changes which the Institution's experience has shown to be desirable.

The seriousness of the problem of underground corrosion, which has been estimated to cost the country £5,000,000 per year, was recognized in the report of the Ministry of Health's Departmental Committee last year, where the importance of full and accurate records being kept of all cases of underground corrosion was emphasized. The B.I.S.R.A. Sub-Committee, under the chairmanship of Mr. L. C. Whiskin of the Metropolitan Water Board and representing the Institution of Water Engineers, is now responsible for the work formerly carried out by the Departmental Committee. It is hoped that the new form will be widely used to report new cases of underground corrosion, especially in pipes laid since 1920.

International Congress on Analytical Chemistry

Some further details are now available concerning the arrangements for the International Congress on Analytical Chemistry which is to be held in this country next year.

The Congress will meet at Oxford during the period 4–9 September 1952, and the scientific sessions will be held in the rooms of the " Examinations Schools ".

Three main Congress Lectures by eminent chemists have been arranged and the programme for the scientific sessions has been divided on a basis of broad techniques.

Papers will be issued in pre-print form before the meeting, and the contributors will give only a brief summary of their papers, most of the time being given over to discussion. Arrangements have been made to publish the whole of the proceedings in a special number, or numbers, of the *Analyst*, as soon as possible after the Congress.

During the period of the Congress it is proposed to have in operation working demonstrations illustrating new techniques or special applications of older techniques in analytical chemistry. In addition to this, and quite separate from it, there will be a trade exhibition comprising apparatus and books.

Some visits have been planned, and at the week-end a number of excursions to places of interest will be arranged.

The Honorary Secretary of the Congress is Mr. R. C. Chirnside, F.R.I.C., Research Laboratories, The General Electric Co., Ltd., Wembley, Middlesex.

Report on Brassfounding

The Brassfoundry Productivity Team that visited the United States last year has recently published its report. Copies may be obtained from the Anglo-American Council on Productivity, 21 Tothill Street, London, S.W.1, or the Association of Bronze and Brass Founders, 25 Bennett's Hill, Birmingham 2, price 7*s.* 6*d.*, post free.

DIARY

Local Section Meetings

6 **November. Oxford.** "The Casting of Light Metals", by A. N. Turner and R. W. Gilkes. (Meeting place to be announced later. At 7 p.m.)

6 **November. South Wales.** "Pressing and Sintering of Metal Powders", by Dr. J. C. Chaston. (Metallurgy Department, University College, Singleton Park, Swansea, at 6.30 p.m.)

8 **November. London.** "Transformations in Metals", by Professor C. S. Barrett. (The Royal School of Mines, South Kensington, S.W.7, at 7 p.m.)

12 **November. Scottish.** "Metals for Gas Turbines", by Dr. J. M. Robertson. (39 Elmbank Crescent, Glasgow, C.2, at 6.30 p.m.)

22 **November. Birmingham.** "Metal Economics", by Professor A. J. Murphy. (James Watt Memorial Institute, Great Charles Street, Birmingham, at 6.30 p.m.)

23 **November. London.** Annual Dance. (4 Grosvenor Gardens, London, S.W.1.)

Other Societies

1 **November. Leeds Metallurgical Society.** "Some Structural Aspects of Recrystallization", by R. Eborall. (Chemistry Department, The University, Leeds 2, at 7 p.m.)

2 **November. Institution of Mechanical Engineers.** "Some Factors Affecting Wear on Cemented Carbide Tools", by Dr. E. M. Trent. (Institution of Mechanical Engineers, Storey's Gate, St. James's Park, London, S.W.1, at 5.30 p.m.)

6 **November. Institution of Engineers and Shipbuilders in Scotland.** "Some Aspects of Research in Friction and Wear", by Dr. F. T. Barwell. (39 Elmbank Crescent, Glasgow, C.2, at 6.30 p.m.)

6 **November. Institution of Production Engineers, Halifax Branch.** "The Production of Steel and Bronze Castings for Engineering Purposes", by G. L. Hancock. (The George Hotel, Huddersfield, at 7.15 p.m.)

7 **November. Institution of Production Engineers, Nottingham Branch.** "The Lost-Wax Process", by A. Short. (Victoria Station Hotel, Milton Street, Nottingham, at 7 p.m.)

7 **November. Manchester Metallurgical Society.** "What is a Dislocation?" by Dr. B. A. Bilby. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

9 **November. Liverpool Metallurgical Society.** "What is a Dislocation?" by Dr. B. A. Bilby. (Lecture Theatre, Electricity Service Centre, Whitechapel, Liverpool, at 7 p.m.)

13 **November. Institution of Works Managers, Birmingham Branch.** "Some Impressions Relative to Works Management Obtained as Leader of the British Non-Ferrous Metals Anglo-American Productivity Council Team", by W. F. Brazener. (The Grand Hotel, Birmingham, at 7 p.m.)

13 **November. Liverpool Metallurgical Society.** "Welding Metallurgy", by Professor E. C. Rollason. Joint meeting with the Liverpool and District Branch, Institute of Welding. (City Technical College, Byrom Street, Liverpool, at 7 p.m.)

14 **November. Institute of Welding, West of Scotland Branch.** "Argon-Arc Welding", by W. A. Woolcot. (39 Elmbank Crescent, Glasgow, C.2, at 7 p.m.)

15 **November. Institution of Mining and Metallurgy.** General Meeting. (Geological Society, Burlington House, Piccadilly, London, W.1, at 5 p.m.)

20 **November. Institute of British Foundrymen, East Anglian Branch.** "Core-Blower Applications and Operations", by G. W. Fairfield. (Public Library, Ipswich, at 7 p.m.)

21 **November. Institute of British Foundrymen, London Branch.** "Problems—Recruitment and Apprentices", by A. Talbot. (Waldorf Hotel, Aldwych, London, W.C.2, at 7.30 p.m.)

21 **November. Institute of British Foundrymen, North-East Scottish Section.** "Foundry Sand Control Technique", by W. Y. Buchanan. (Imperial Hotel, Arbroath, at 7.30 p.m.)

21 **November. Manchester Metallurgical Society.** "Spectrographic Analysis", by W. Ramsden. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

21–22 **November. Iron and Steel Institute.** Autumn General Meeting.

APPOINTMENTS VACANT

CHEMIST, recently graduated, wanted for work in connection with the control and development of works processes by manufacturers of Aluminium Sheet, Extrusions, and Paste. The Chemist's work would involve both organic and inorganic chemistry. Box No. 310, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGICAL CHEMIST. A rapidly expanding organization in the East Midlands requires a metallurgical chemist to undertake a variety of analytical work in the ferrous and non-ferrous fields. A knowledge of the properties and analysis of oils and greases would be an advantage. The nature of the post demands that only applicants who can work on their own initiative and take responsibility need apply. Full details of experience and present salary to Box No. 313, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGIST required with experience in metallography for Propeller Laboratory. Applicants should be Licentiate Members of the Institution of Metallurgists or should possess equivalent qualifications. Salary according to age, qualifications, and experience. Write giving details of qualifications, experience, age, and salary sought, to Box 311, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

PHYSICIST required by the Division of Atomic Energy (Production), Risley, to supervise research work connected with the mechanical properties of metals and alloys, including creep, fatigue, corrosion-fatigue, and stress-corrosion, and to be responsible for the interpretation of results according to modern physical conceptions. A large proportion of the work will be fundamental in nature.

Candidates must have a first or second class honours degree in physics or engineering or equivalent qualification. They should be at least 26 years old and have had not less than three years' experience in this type of research.

Salary will be assessed according to qualifications and experience within the range £720–910 p.a. Rates for women are somewhat lower. The post will carry F.S.S.U. benefits. Applications to Ministry of Supply, D.At.En. (P), Risley, Nr. Warrington, Lancs., stating post applied for.

RESEARCH ESTABLISHMENT in the Midlands requires a senior investigator to take charge of research on the welding of light alloys. Candidates should possess a good honours degree and research experience on welding. Box 312, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

NOTICE TO AUTHORS OF PAPERS FOR THE "JOURNAL"

AND CONTRIBUTORS TO DISCUSSIONS

1. **Papers will be considered for publication from non-members as well as members of the Institute.** They are accepted for publication in the *Journal* and not necessarily for presentation at any meeting of the Institute. MSS. should be addressed to The Editor of Publications, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

2. **Papers suitable for publication** may be classified as:

(*a*) Papers recording the results of original research.

(*b*) First-class reviews of, or accounts of progress in, a particular field.

(*c*) Papers descriptive of works methods, or recent developments in metallurgical plant and practice.

(*d*) Papers in classes (*a*), (*b*), and (*c*) above, previously published in languages other than English, French, German, or Italian, if of sufficient merit.

3. **Manuscripts and illustrations** should be submitted in duplicate. MSS. must be typewritten (*double-line spacing*) on one side of the paper only, and authors are requested to sign a declaration that neither the paper nor a substantial part thereof has been published elsewhere. Exceptions may be made in certain cases where a paper has been published in a language other than English, French, German, or Italian (see 2(*d*) above). MSS. not accepted are normally returned within 6 months of receipt.

In the interests of economy, all papers must be written as concisely as possible; in general, internal research reports are not in suitable form for publication as papers in the *Journal*. All but the simplest mathematical expressions should be written by hand, with capital and small letters clearly distinguished. Superscript and subscript letters should also be plainly indicated. Greek letters and special signs should be identified in the margin. For style, spelling, and abbreviations used, any recent issue of the *Journal* may be consulted.

4. **Synopsis.** Every paper must have a synopsis (not exceeding 250 words in length) which, in the case of a paper reporting original research, should state its objects, the ground covered, and the nature of the results. The synopsis will appear at the beginning of the paper, and should be in a form suitable for use by abstracting organizations. Extracts from a "Guide for the Preparation of Synopses" drawn up by the Abstracting Services Consultative Committee are reproduced below.

5. **References** must be collected at the end of the paper and must be numbered in the order in which they occur in the MS. Initials of authors must be given, and the Institute's official abbreviations for periodical titles (as used in *Metallurgical Abstracts*) should be employed, where known. References to papers should be set out in the style:

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

References to books should be in the following style:

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

6. **Illustrations.** Each illustration must have a number and description; only one set of numbers must be used in one paper, and it is desirable to number the half-tone illustrations consecutively, rather than to intersperse them with the line figures. The captions should be typed on a separate sheet.

The set of **line figures** sent for reproduction must be drawn (about twice the size to appear in the *Journal*) in Indian ink on smooth white Bristol board, good-quality drawing paper, co-ordinate paper, or tracing *cloth*, which are preferred in the order given. Co-ordinate paper, if used, must be blue-lined, with the co-ordinates to be reproduced finely drawn in Indian ink. Curves should be drawn boldly (i.e. at least twice the thickness of the frame). Experimental points should be indicated by open or closed circles, triangles, squares, &c. (preferably not crosses). Curves should be broken on each side of such symbols and plenty of allowance should be made for closing up in blockmaking. All lettering and numerals, &c., should preferably be in *pencil*, so that the Institute's standard lettering may be affixed, and ample margins must be left outside the framework of the figures to enable this to be done. The second set of line illustrations may be photostat copies.

Photographs must be restricted in number, owing to the expense of reproduction, and photomicrographs should be trimmed to the smallest possible of the following sizes consistent with adequate representation of the subject: 4 in. deep by 3 in. wide: 2 in. deep by 3 in. wide: 2 in. square. Magnifications of photomicrographs must be given in each case. Photographs for reproduction should be loose, not pasted down (and not fastened together with a clip, which damages them), and the figure number and author's name should be written on the back of each. Captions should be given to the photomicrographs, but these should be kept as brief as possible.

Because of the present high cost of printing and paper it is imperative that authors restrict illustrations (particularly photographs) to the absolute minimum deemed necessary to support their argument. Only in exceptional cases will illustrations be reproduced if already printed and readily available elsewhere.

7. **Tables or Diagrams.** Results of experiments, &c., may be given in the form of tables *or* figures, *but* (unless there are exceptional reasons) *not both*. Tables should bear Roman numbers, and each should have a heading that will make the data intelligible without reference to the text.

8. **Corrections.** A certain number of corrections in proof are inevitable, but any modification of the original text is to be avoided. Since corrections are very expensive, the Institute reserves the right to require authors to contribute towards their cost if the Editor deems them to be excessive. The Institute also reserves the right to require a contribution to the cost of remaking any block where this is necessitated by an error on the author's part.

9. **Reprints.** Individual authors are presented with a maximum of 25, and two or more authors with a maximum of 50 reprints from the *Journal*, without covers. Limited numbers of additional reprints can be supplied at the author's expense, if ordered before proofs are passed for press. (Orders should preferably be placed when submitting MSS.)

10. **Discussion.** Except in the case of special symposia, shorthand records of discussions are not taken at meetings. Written discussion may be submitted on any paper, preferably typewritten (*double-line spacing*). References should be given in the form of footnotes. Paragraphs 6 and 7 above are also applicable to such contributions. Reprints of discussion cannot be supplied to contributors.

GUIDE FOR THE PREPARATION OF SYNOPSES

(*As recommended by the Abstracting Services Consultative Committee*)

1. **Purpose.** The synopsis is not part of the paper; it is intended to convey briefly the content of the paper, to draw attention to all new information, and to the main conclusions. It should be factual.

2. **Style of writing.** The synopsis should be written concisely and in normal rather than abbreviated English. It is preferable to use the third person. Where possible use standard rather than proprietary terms, and avoid unnecessary contracting.

It should be presumed that the reader has some knowledge of the subject, but has not read the paper. The synopsis should therefore be intelligible in itself without reference to the paper; for example, it should not cite sections or illustrations by their numerical references in the text.

3. **Content.** The title of the paper is usually read as part of the synopsis. The opening sentence should be framed accordingly and repetition of the title avoided. If the title is insufficiently comprehensive, the opening should indicate the subjects covered. Usually the beginning of a synopsis should state the objective of the investigation.

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

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

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

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

4. **References.** If it is necessary to refer to earlier work in the summary, the reference should always be given in full and not by number. Otherwise references should be left out.

When a synopsis is completed, the author is urged to revise it carefully, removing redundant words, clarifying obscurities, and rectifying errors in copying from the paper. Particular attention should be paid by him to scientific and proper names, numerical data, and chemical and mathematical formulæ.

22nd AUTUMN LECTURE

1329

ELECTROCHEMISTRY AND THE SCIENCE OF METALS *

By Professor ROBERTO PIONTELLI †

SYNOPSIS

After giving an historical account of the origin and growth of electrochemistry, the author surveys the present state of knowledge on various aspects of electrochemistry that are closely related to metals. Particular attention is given to the question of the seat of electromotive force.

I.—INTRODUCTION

First, I should like to express my appreciation of the honour of having been invited to deliver this Autumn Lecture to the Institute of Metals. My choice of subject has been made on the one hand because electrochemistry is one of the few topics which I feel capable of treating at length; and on the other because it affords me an opportunity of directing the attention of an audience thoroughly versed in the science of metals to the problems of the electrochemist.

Every technologist will readily admit the important, in some respects unique, part played by electrochemistry in the field of metallurgy. The chemical action of electricity covers a wide range of free energies of reaction, and thereby exercises a predominating influence in certain directions, as, for instance, in the production of the less noble metals. The speed of such action and also its selectivity may be easily regulated, thus making possible a very high standard of purity of electrometallurgical products. The value of electrical instruments in control operations and in recording essential data does not need to be stressed. The measurement of electromotive forces at various temperatures affords a means of direct determination of free-energy and entropy changes in chemical processes, thereby supplying valuable data unobtainable by any other means.

Electrochemical methods are, moreover, finding wider and wider application in research on such subjects as the structure, reactivity, and catalytic power of metal surfaces; the mechanism of nucleation, growth, and breakdown of metal crystals; the genesis and significance of " secondary " structures; the existence and conditions of formation of unstable intermetallic phases; metal-gas systems, &c. Even so brief a summary indicates the importance of electrochemistry in extraction, refining, protection, finishing, analysis, research, and many other fields. The sciences of metals, metallurgy, and electrochemistry are natural allies and the bonds between them are certainly not destined to relax.

For the future progress of both theoretical and applied electrochemistry it is in fact essential that the bonds should be drawn ever closer, and that the mechanism of their electrochemical behaviour should become as much a part of the study of metals as that of their mechanical and physical properties.

What I have to say will therefore constitute primarily an appeal, recalling that made by Sir Robert Boyle in the middle of the seventeenth century to the physical and corpuscular philosophers, urging on them a close and mutually profitable collaboration, inasmuch as it had been fully demonstrated how "experimenta chymica per corpuscularium doctrinam feliciter explicari; ita vicissim huic ex illis vel lucem vel confirmationem non raro conciliari posse ".[1]

To demonstrate more clearly the essential features of present-day problems, I should like to draw on historical sources to trace their origin and development.

The lack of collaboration so greatly deplored by Boyle did not by any means denote, either in that or the preceding epoch, inactivity on the part of the philosophers of either of the two groups, in that very field of metals that we, too, have so closely at heart. It was the feverish activity of the alchemists that stored up the rich but confused fund of knowledge which made possible the development of modern chemistry; while the imaginative spirit of the " corpusculares " was also in ceaseless turmoil, endeavouring to establish in " formulations of everlasting philosophy " the protean nature of metals. To both groups the electrochemical problems of metals presented themselves in a particularly complicated form, in so far as behaviour towards atmospheric and chemical agents was concerned. It had long been known that iron coming into contact with waste water from copper mines was

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* Delivered at the Annual Autumn Meeting, Venice, 17 September 1951.

† Laboratori di Elettrochimica, di Chimica Fisica e Metallurgica e di Metallurgia del Politecnico, Milan.

transformed into copper. Such observations had led in distant times to the drawing up of the scale of "metallic perfection" headed by gold, and had been the source of endless illusions and misconceptions to those who, in good or in bad faith, saw in displacement reactions proof of the transmutability of metals—a road minutely explored with the object of appeasing the eternal lust for gold.

II.—HISTORICAL SURVEY

I should like to survey briefly a few theories of the nature of metals from the wealth of historical material that I have accumulated.

First, there is the astrological conception—probably the most ancient—which attributed the behaviour of every metal to the more or less direct influence of a planet with which it was in affinity. Such affinity, for instance, was attributed by one school of thought to an infinite number of small corpuscles emerging from the planet and the metal, and possessing a form suitably adapted to the porosity of the planet and metal in question, but to no others. Let us not, however, too readily jeer at and condemn this apparently fantastic idea, lest the physics of cosmic rays should sooner or later disclose some phenomenon of a similar nature.

For centuries the most widely accepted explanation of the metallic state was based upon the blending of "constituent principles" (of Aristotelian inspiration): mercury, sulphur, and subsequently, salt. The purity of these materials, their relative proportions, the degree of their blending (accomplished in the depths of the earth but under the constant influence of the planets) determined the perfection of the various metals and the degree of their respective sympathies and antipathies.

The spread of clearly defined corpuscular theories during the second half of the seventeenth century—just at the time of Boyle's appeal—is worthy of note. The behaviour of individual metals was attributed to the form and compactness of the constituent corpuscles, which affected the size and shape of the pores accessible to reactive particles of the chemical agents. Behaviour therefore depended not only upon the penetrative properties of the agent, but also upon the internal and external structure of the metal.

The development of the phlogiston theory, however, represented a retrogression, inasmuch as the behaviour of metals was then attributed to the qualities of the phlogiston each possessed. Towards the end of the eighteenth century chemical doctrine was being radically revised. Bitter but fruitful disputation between Galvani and Volta and the ensuing clash between electricity and physiology definitely marked the advent of electrochemistry. Studies of the behaviour of metals by Fabbroni * (1792), Ritter,[3] and others disclosed further and vaster fields open to electrochemistry. The discovery of the electric pile (1800) saw the problem of the day—the origin of galvanic fluid—extended to the working mechanism of the new apparatus. Volta had to oppose not only the supporters of "animal electricity", but also those of the "chemical theories"; moreover, he had to refute † the extreme opinions attributed to him—opinions which are still quite generally attributed to Volta to-day.

In a galvanic circuit Volta himself perceived a state of "electric tension" set up as a result of the formation of various heterogeneous contacts. The absolute or relative importance of the contributions of such contacts to the tension, and also the mechanism of these contributions, formed the crux of Volta's problem.

The supporters of the chemical theories refused to acknowledge any source other than the chemical effects at the metal/solution contacts. Volta was opposed to such an interpretation, being profoundly convinced of the importance of intermetallic contacts, and also because of the possibility of such chemical effects manifesting themselves independently of the galvanic phenomena, or even as a result of the circulation of the current.‡

The embittered nature of the controversy, together with the keen interest aroused by the phenomena involved—such as that of the decomposition of water (Nicholson and Carlisle[5]; 1800)—resulted in the advent of electrometallurgy (Cruickshank and Brugnatelli; 1800) passing almost unnoticed.

In his "History of Electricity"[6] Priestley attributes to G. B. Beccaria, the Turin physicist, the first experiments (about 1756) on the reduction of metallic compounds by friction machines. Although those experiments,[7] in which the thermal action of the

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* "I have also observed that the alloys utilized for the welding of the copper plates on the sliding roof of the Florence Observatory have rapidly deteriorated, and transformed themselves into white oxide at all points of contact—welded joints —with the copper plates."[2]

† "Je n'ai jamais attribué aux métaux exclusivement la faculté d'inciter le fluide électrique par leur contact mutuel, lorsqu'ils sont de différentes espèces; ayant reconnu, et prouvé par un grand nombre d'expériences directes, que cette faculté appartenait, sans exception, à tous les conducteurs; et que si elle était en général plus marquée entre les métaux, elle ne laissait pas de se manifester dans le contact d'un métal . . . avec un de la seconde, ou conducteur humide. . . ."

"L'action électrique qui se déploie par le contact de plusieurs solutions salines avec les métaux—principalement de certains acides avec certains métaux, et des alkalis concentrés avec presque tous les métaux—est souvent plus forte et plus marquée que celle qui se déploie par le contact mutuel de deux métaux peu différents entre eux."[4]

‡ "L'oxidation est en partie indépendante de l'action galvanique, ou pour mieux dire électrique; car elle est l'effet chimique ordinaire de tel ou tel fluide sur tel ou tel métal; elle en dépend en partie, en tant que le courant électrique modifie singulièrement cette oxidation; en l'augmentant beaucoup dans le métal d'où le courant sort pour passer dans l'eau ou tout autre liquide oxydant, et en la diminuant ou supprimant tout à fait dans le métal où le courant électrique entre, et où le gaz hydrogène se développe. Ainsi donc le courant électrique exerce une action oxydante, et une désoxydante, suivant qu'il passe d'un métal dans un liquide, ou du liquide dans le métal; mais cette action n'est nullement la cause du courant, elle n'en est que l'effet."[4]

current certainly played a decisive part, contain in some respects the first germs of electrometallurgy, they remained undeveloped, and the true path to metallurgical applications was disclosed only with the discovery of the electric pile. In London Cruickshank, employing the Voltaic apparatus with the object of obtaining oxygen free from hydrogen, decided to eliminate the latter by utilizing it to reduce dissolved metallic compounds. Using lead acetate, he observed the formation on an electrode of tiny needles of lead and later of dendritic crystals; using copper sulphate, he observed the separation of copper.[8] Brugnatelli, who had the good fortune to work at Pavia, in close collaboration with Volta, and who concentrated on obtaining practical results rather than on establishing general laws, recorded the phenomena of the cathodic separation of silver, mercury, zinc, and copper.[9] At the time when Davy was entering upon his series of researches, which proved so valuable not only for the results obtained but also for the perfection of their technique, and the clarity of their interpretation, Brugnatelli [10] began work on galvanotechnical problems. This led to the discovery of his galvanic " gilding " process (1803), to the metallizing of carbon with gold, silver, and copper (1816), and to the successful electrodeposition of alloys of copper with gold, silver, and tin (1818).

In Faraday, the physical and chemical schools of thought, which had been becoming more and more divergent, met and temporarily achieved an harmonious union. " What is the source of power in a Voltaic pile ? " he wrote. " This question is at present of the utmost importance in the theory and to the development of electrical science." And again : " All the facts show us that power, commonly called chemical affinity, can be communicated to a distance through the metals . . . that, in other words, the forces termed chemical activity and electricity are one and the same."[11] Faraday's contribution was indeed fundamental, and served to clothe with a true theory the ideas (bearing in part the highly original imprint of Davy) of the supporters of the chemical theory, whose numbers had in the meantime been augmented by such scientists as Oersted, De la Rive, Becquerel, Pouillet, Schönbein, and others. Divergencies of opinion nevertheless continued with increasing acrimony.

The activities of the physicists were, however, directed more and more towards new and promising fields of research in electrodynamics and electromagnetism; and, despite the fact that the results of Cruickshank and Brugnatelli were almost overlooked, applications of the electrochemistry of metals proceeded to increase and multiply.

The names of De la Rive, Jordan, Spencer, Elkington, Ruolz, Jacobi, Smee, Murray, and numerous others are still fresh in our memories, and it may be of interest to note that in his early days (about 1830) Sir Henry Bessemer did some valuable pioneer work in this field. Nobili [12] and Marianini [13] published important theoretical and applied studies of the subject; and mention should be made of the name of Zantedeschi,[14] who was active in Venice and foremost among the pioneers in the art of galvanoplastics.

In the field of classical research the study of the Voltaic apparatus continued to provide a never-ending subject of discussion. The theoretical and experimental research work of Helmholtz, Joule, Clausius, Lord Kelvin, and others, provided a perfect working test-bench for the new doctrines of thermodynamics. From the science of energy as laid down by Ostwald,[15] a severe criticism emerged of the views of Volta and his followers in regard to the importance of intermetallic contacts. However, other authors, in particular Helmholtz, by virtue of the electro-molecular-forces hypothesis and the electrical double-layer theory of contacts, had confirmed the importance of Volta's views, in a form which has continued to arouse interest down to the present day.

More recently, new chemical and physical concepts—in particular Arrhenius' theory of electrolytic dissociation and van't Hoff's osmotic theory of solutions—seem likely to shed important new light on the problem. Nernst's " osmotic theory of the pile " is confined substantially to contacts involving liquids (the contribution of the intermetallic contact being supposed to be represented by thermoelectric effects). While avoiding a crude conception of ionic-exchange processes, Nernst's theory attempts to explain, in terms of the solution pressure on the one hand and of the osmotic pressure on the other, the opposing chemical tendencies which the difference in electric potential is called upon to counterbalance.

The development of the new electron theories, and in particular their application to the phenomena of electronic emission from hot bodies (Richardson [16]), leads, however, to the conclusion that, in view of the different " affinities " of the various metals for electrons, the exchanges of electrons between metals must represent a definite amount of chemical energy. The origin of the electrical work accomplished by a Voltaic battery is essentially chemical (as thermodynamics requires), but henceforth all the particles constituting the electric fluid increasingly assume the aspect of a chemical species, thereby fully justifying the hypothesis of electromolecular forces which are also active at the intermetallic contact.

III.—THE SEAT OF ELECTROMOTIVE FORCE

Chemical processes in a galvanic circuit can no longer be considered as exclusively confined to the contacts between metals and solutions, for the exchange of electrons at contacts between dissimilar metals forms an integral part of the overall chemical process in the battery, in the same way as does the exchange of ions between metals and solutions. If therefore Volta's substantial reasons for opposing chemical doctrines are demolished, the bases of the refutation of the ideas of Volta and his followers on the contribution of intermetallic contacts are likewise destroyed. The curve of the progress of human knowledge is by no means a smooth one, but has many ups and downs. On the basis of experimental values of the work of extracting electrons some (so-called "neo-Voltian") physicists, are again seeking in the intermetallic contact the main or sole source of electromotive force.

The methodical revision of ideas, which is universally adopted in physics, when applied to electrochemical problems, leads to certain conclusions, which at first sight may appear discouraging, although they have already been clearly dealt with by Gibbs.[17] "The consideration of the electric potential in an electrolyte, and especially the consideration of the difference of potential in electrolyte and electrode, involves the consideration of quantities of which we have no apparent means of physical measurement, while the difference of potential in pieces of metal of the same kind attached to the electrode is exactly one of the things which we can and do measure."

And so we now come to the present position, which in certain respects may seem paradoxical: for while some physicists maintain that the problem of the seat of electromotive force is quite real, and one which may in some cases be regarded as solved, or as amenable to solution either experimentally (by various electrocapillary and electrokinetic methods) or theoretically (on the basis of our knowledge of the structure of the phases); others hold that the problem is devoid of any concrete physical significance.[18]

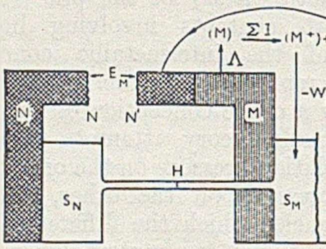

FIG. 1.—Cell for the Study of the Electrochemical Behaviour of a Metal.

$$E^{\circ}_{M} \simeq \Omega_{M} + K.$$

$$\Omega = \frac{1}{z}(\Lambda + \Sigma I - W_{M^+}) = \frac{1}{z}(z\phi_0^{[M]} + \phi_{M^+}^{[M]} - W_{M^+})$$
$$= \frac{1}{z}(U^{[M]} - W_{M^+})$$

In order to get a clearer picture of the situation, let us endeavour to follow a course that cannot, from the very outset, be condemned on purely logical grounds.

Consider a system such as that shown in Fig. 1, which we will suppose to be isothermal, and in which the metal (M), which it is proposed to study, is immersed in a solution (S_M) of one of its salts. The solution is, in turn, connected (by means of an "interliquid contact") with a reference electrode (N/S_N). As a precautionary measure, we shall choose physically identical materials for the two terminals of the circuit. Ordinary temperatures are considered.

Under these conditions it is possible to give the problem a fairly general significance. In fact, experiment furnishes us, by well-known direct methods, with the electric potential difference, E_M, between the terminals of the galvanic cell.

The series of equilibrium values E^{0}_{M}, determined for cells corresponding to the so-called "standard" conditions, constitutes the well-known "electrochemical series" or "scale of nobilities".

How is E^{0}_{M} made up of the contributions from the various contacts? Apart from the lack of an experimental method of determination, this problem is complicated by the absence of agreement on what constitutes such a "contribution". Any solution is of an essentially speculative character, but considering the extent and nature of past efforts in this direction, it seems unfair to ignore the problem altogether.

Let me therefore reaffirm an opinion that I have already discussed in detail elsewhere.[19, 20] When considering any one of the components of the circuit, isolated and electrically neutral, we find ourselves confronted with the problem (well defined both from the physico-structural and thermodynamic points of view) of the work needed to deprive it of its various constituent particles. As far as electrically charged particles (ions and electrons) are concerned, the forces—to the existence of which such work corresponds—may be supposed to be partly of an electrical and partly of a chemical nature.

This division is essentially arbitrary in character, inasmuch as according to the scale of atomic and subatomic dimensions at least one part of the forces we normally consider as being chemical appears to be of an electrical nature.[19] It is not my intention to enter deeply into such a difficult matter, and I shall confine myself to the view accepted by many writers, according to which a prudent application of mathematical electricity of the macroscopic type (Maxwell) is admissible, even internally, to a material.[19-22] The contour surface will be assumed to be uniform.

The average value of the internal potential of the phase is then essentially determined by the electric double layers existing on its surface. Such double layers may be present both for intrinsic structural reasons (for example, owing to the unsymmetrical configuration of the electronic atmospheres of the peripheral atoms), and as a result of a chemical alteration of the surface (adsorption of gas, oxidation, tarnishing, &c.), which may lead to the formation of new phases.

Let us now consider what happens when two different phases—initially isolated and electrically neutral—come in contact with each other without mixing. Owing to the different "affinities" (in the thermodynamic sense of the term) in respect of the various kinds of particles, we may suppose, generally, work to be available in the system that is liable to lead to an exchange of particles between the two phases. Such a mechanism (and in particular that of rearrangement of the interphase region and of the distribution of charges on the free surface of individual phases) will continue to operate until the available work is used up. The masses involved in these exchange processes being extremely small, the internal chemical properties of the phases and the double layers on their free surfaces, will remain practically unaltered after the contact. The establishment of equilibrium is, however, essentially the result of these changes in the distribution of electrical charges.

The final distribution of these charges proves to be complicated, but it is always such as to satisfy the requirements necessary under equilibrium conditions to attain uniformity in the average electric potential in each of the phases in contact, and to neutralize the work available for the process of exchange of the particles between the phases.

At a given point in the interior of the phase, the electric potential is now determined by: (*a*) the double layers on the free surface of the phase which already existed before contact; (*b*) the free charges that have been superimposed on such layers as a result of the contact; and (*c*) the distribution of charges originating in the interphase region.

Turning from consideration of an internal point of a phase to an external one, though always one immediately adjacent to the surface (normally distances of the order of 10^{-4} cm. are considered, namely those just out of the practical radius action of the so-called " image forces "), the variation of the electric potential is due exclusively to the double layers, either in the initial state or after the contact. Between two external points, immediately adjacent respectively to the two phases, the potential difference before the contact is zero, while after the contact an electric-potential difference exists, directly measurable by experiment, commonly called the " Voltaic potential difference ". This Voltaic potential difference is very simply related to the work of extraction of the particles exchanged, and it expresses also the variation of the potential difference between internal points of the phases, with respect to the condition before the contact. This quantity, which thus represents the electrical consequences of the establishment of contact equilibrium, and the work initially available for which it must compensate, can therefore be reasonably supposed to be a measure of the " contribution " made to the overall electromotive force of the circuit by the contact itself, an assumption corresponding closely to Volta's original ideas.

Let us briefly examine the principal implications of this assumption. It is well known that E_M is the sum total of the Volta potential differences corresponding to the various contacts. The fact that a significant contribution to the electromotive force of a Daniell cell may derive from an intermetallic contact which is not a site of chemical changes, does not present any difficulty, inasmuch as the exchange of electrons between the phases, which is concentrated at the contact itself, is also a chemical process, and is governed by affinity.

The fact that the individual Volta potential differences depend on atmospheric action at the surfaces of the phases, whereas such action does not enter into the electromotive force of the circuit (at identical terminals), also presents no difficulty, when we consider that each phase takes part in at least two contacts (one of them possibly with the atmosphere), the net contribution of which is independent of the influence of the double layers on the free surfaces. Such a contribution consisting of two contacts involving the same phase, also eliminates the indeterminateness inherent in the arbitrary value of the internal potential of the individual phases.

It thus seems to be proved that the problem of the seat of electromotive force is a definite one, and one which leads to-day to a somewhat Salomonic solution, inasmuch as it attributes importance to intermetallic contacts as well as to contacts between metals and liquids, and takes account also of the part played by the chemical changes at the surface due to the atmosphere, not as being the source of power, but as the cause of existence of or of variation in the surface double layers.

However, the indeterminateness inherent in the " absolute " electric potential differences, and in determining the part played in the above contributions by the internal properties or by the surface ones, still persists.

As may be gathered from the above outline, such a problem not only presupposes a definition (an arbitrary one, of course), but it admits only of theoretical solutions which could be based solely upon an adequate knowledge of internal structure and surface properties of the various phases, and in particular of the metallic ones.

Such knowledge, therefore, is one of the possible objectives that the collaboration between electrochemistry and the science of metals might seek to reach; it is a speculative aim and a somewhat distant one at that; but the road which leads to it would certainly reveal new horizons in the classical problems of thermionic and photoelectric emission, adsorption, catalysis, heterogeneous reactions, and so on.

IV.—ELECTROCHEMICAL PROPERTIES OF METALS

Let us now pass to the fundamental problem of the electrochemical behaviour of the various metals in relation to their other chemical and physical characteristics. We are not yet able to predict this behaviour on the basis solely of the electronic structure of the free atoms, but we may characterize the behaviour by means of properties more closely related to the atomic structure.

If we compare different galvanic cells, which vary only because of the nature of the metal M (and hence of the salt dissolved in S_M), the differences in the values of E_M are essentially to be attributed to the contacts M/N and M/S_M, as the contribution of the liquid/liquid contact may be considered practically constant (and in any case very small). The problem of the relative values of E°_M and of their relation to the properties of the various metals, thus leads back to the total contribution of the two above-mentioned contacts. In consequence of the processes that take place at them (exchanges of electrons at the intermetallic contact and of ions at the other contact), the phase M loses or gains atoms.

We may arrive indirectly at the same result (for the anodic direction of the reaction), if we assume the evaporation of the atoms M, their transformation into ions and electrons, and the passage of the former into S_M and of the latter into N. To a rough approximation (neglecting in particular some entropy changes) which nevertheless is sufficiently close for our present purpose, the part involving total chemical affinity of transformation, and therefore also the part of E°_M which depends upon M, may be considered to be measured by the quantity Ω_M (the so-called " nobility index " of the metal M), which may be expressed in one of the equivalent forms: [19, 20]

$$\Omega_M = \frac{1}{z}(\Lambda + \Sigma I - W_{M^+}) = \frac{1}{z}(z\phi_0^{[M]} + \phi_{M^+}^{[M]} - W_{M^+})$$
$$= \frac{1}{z}(U^{[M]} - W_{M^+})$$

where z is the valency of the " normal " ionic form of M; Λ is the energy required for the sublimation of M; ΣI is the sum of the ionization potentials of the free atoms, and W_{M^+} the chemical energy developed in the solvation of the gaseous ion; $\phi_0^{[M]}$ and $\phi_{M^+}^{[M]}$ are the " works of extraction " from the metal, of electrons and ions, respectively; while $U^{[M]}$ represents the energy necessary for dissociating the crystal lattice into ions and electrons.

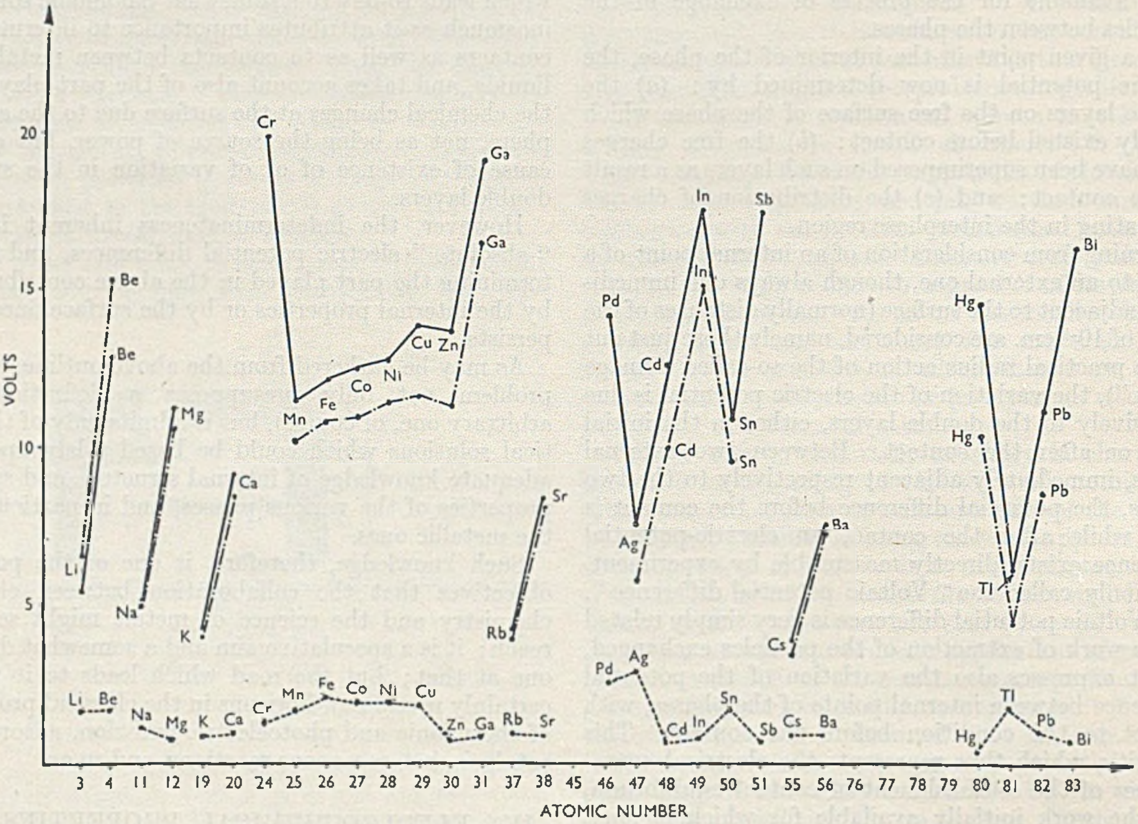

FIG. 2.—Values of Λ/z (- - - -), $\Sigma I/z$ (——), and W_{M^+}/z (— · —) as Functions of the Atomic Number. The abscissa is not uniform.

The values (in volts) of the ratios of some of these quantities to z* for numerous metals, are shown in Fig. 2. It should be noted that: (1) Λ expresses the affinity of the lattice for the atoms M; (2) ΣI measures the affinity of the individual atom for the valence electrons; and (3) W_{M^+} expresses the chemical affinity of the solution S_M for the ions M^+. Only the relative values of W_{M^+} can be indirectly determined by experiment, but for our present purposes, which are purely comparative, this does not matter.

The two largest terms (ΣI and W_{M^+}) often counteract one another (in a certain sense the process of solvation tends to restore the electronic configuration of the atom); for this reason Λ is often of decisive importance. The works of extraction $\phi_0^{[M]}$ and $\phi_{M^+}^{[M]}$, taken separately, cannot generally be considered as " purely chemical " work, for they comprise a chemical part and an electrical part (we here encounter once more the indeterminateness which obliged us

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* Taken from various sources [23] and in part original.

to abandon the problem of the absolute potential differences).

Taking the sum $U^{[M]}$, which corresponds to the loss from the metallic phase of equal charges of opposite sign, and which expresses the resistance of the lattice to dissociation into ions and electrons, we eliminate the electrical terms and thus obtain a quantity expressing the part of the active chemical affinity of the circuit, that depends solely upon M. This quantity, together with W_{M^+} (which also depends on the solution), determines the relative degree of nobility of M.

This fundamental property is thus correlated with others of more obvious significance. Fig. 3 shows the good agreement existing between the values of Ω_M and of E°_H.* In this figure are recorded together the values (both as divided by z and in volts) of the heats of formation (at 25° C.) of the oxides of M assuming the usual valency z $\left(\frac{-\Delta H_{\text{ox.}}}{z}\right)$. These values vary inversely with those of Ω_M. It is easy to see that the nobility index of the various metals also expresses, from a thermodynamic point of view (and in a relative way), their tendency to resist chemical attack from various sources and thus the predisposition of the metal not to return, as soon as a chance presents itself, to the "natural" forms of chemical bond from which the efforts of the metallurgist have torn it.

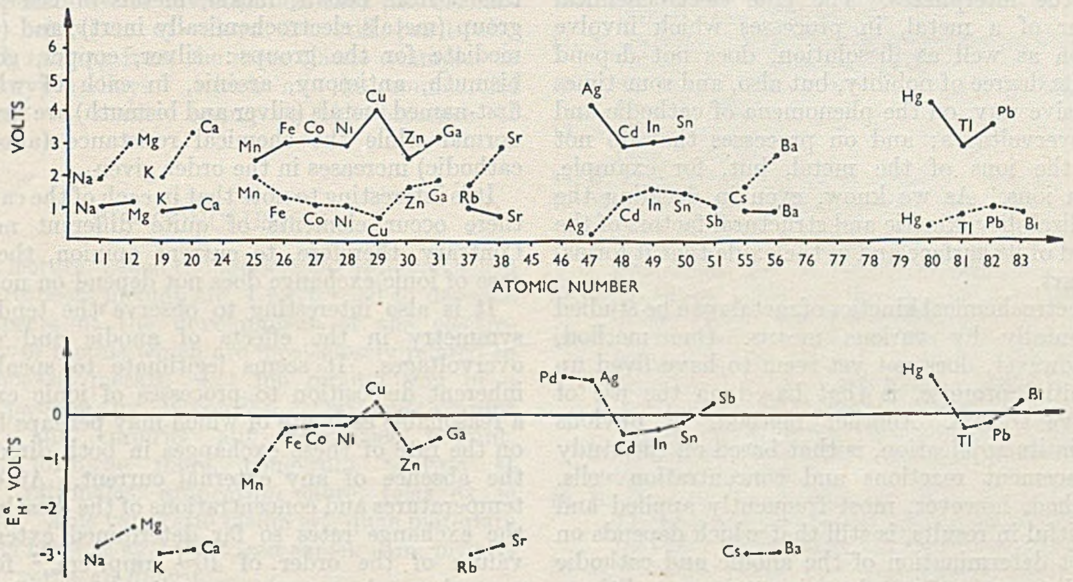

FIG. 3.—Values of Ω_M, $-\Delta H_{\text{ox.}}/z$, and E°_H as Functions of the Atomic Number. The abscissa is not uniform.

$$—— = \frac{1}{z}(\Lambda + \Sigma I - W_{M^+}). \qquad - - - = -\Delta H_{\text{ox.}}/z.$$

These reactions may be considered as the sum of two partial reactions, in which the metallic phase cedes ions and electrons, respectively, and in particular the thermodynamic affinity of reaction which depends on M can be measured, in a relative way, by Ω_M. When, however, the final state of the ion is not that of plain solvation, it may be necessary to replace W_{M^+} by the heat of the reaction of formation, starting from the gaseous ion, of certain aggregates.

This holds good in the formation of complex ions in solution; and also in the formation of solid oxides, starting from the metal and gaseous oxygen. In the latter case, however, experiment shows (see Fig. 3) that the inverse variation of the heat of oxidation and the degree of nobility is well preserved, indicating that W_{M^+} runs parallel with the thermal output of the reaction between the ion M^+ and the oxygen ion in the gaseous state.

Allow me to draw an analogy, which is perhaps rather far-fetched. A metallic phase is like a community which may be considered to be composed of individuals of two species (electrons and ions) and of the families (atoms) formed by these individuals. When the phase is engaged in an electrochemical process, it is fighting, so to speak, on two fronts, against the attractions exercised by its surroundings. The ability of the aggregate to resist, upon which its degree of nobility depends, is determined by (*a*) solidarity between families (Λ); (*b*) the resistance of a family when alone (ΣI); and (*c*) by the disposition of the individuals, when free, not to become too easy a prey to their surroundings ($-W_{M^+}$). We have noticed the important part played by solidarity in deciding the fate of the aggregate.

V.—KINETIC ASPECTS OF ELECTROCHEMICAL BEHAVIOUR

Up to now we have considered exclusively the thermodynamic aspects of the electrochemical behaviour of metals. From a practical point of view, and in my opinion also from the theoretical standpoint,

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* Values of E°_H referred to the "normal hydrogen electrode" at 25° C.

the kinetic aspect is of still more lively and real interest.

In all physicochemical problems, thermodynamic affinity, referred to the internal properties of homogeneous phases, taken in large masses, supplies adequate, but not overwhelming conditions for equilibrium. It is a measure of the availability of work, which is indispensable for producing a transformation but which may not be sufficient to initiate it; while the speed of transformation depends also on the other kinetic factor, the " resistance to reaction ".

In electrochemistry, these resistances are closely bound up with the structural properties of the phases, and of the interphases. The true electrochemical behaviour of a metal, in processes which involve deposition as well as dissolution, does not depend only on its degree of nobility, but also, and sometimes in a decisive way, on the phenomena of cathodic and anodic overvoltages; and on processes that do not involve the ions of the metal, but, for example, hydrogen ions. As we know, even in deciding the true oxidizability, kinetic and structural factors of the metal and of its surface and of the oxide that it forms, play a part.

The electrochemical kinetics of metals can be studied experimentally by various means. One method, which, however, does not yet seem to have lived up to its initial promise, is that based on the use of radioactive tracers. Another method, of obvious interest in its application, is that based on the study of displacement reactions and concentration cells. The method, however, most frequently applied and most fruitful in results, is still that which depends on the direct determination of the anodic and cathodic polarization curves of metals under various conditions.

It should here be pointed out that the uncertainties which exist with regard to the absolute potential differences of contacts (intermetallic and metal/liquid) no longer figure in the case of polarization voltages. Everything leads us to believe that intermetallic contacts are unpolarizable, for which reason " overvoltages " can with certainty be attributed to metal/liquid contacts, and (under suitable conditions) they supply a measure of the " chemical resistance " which hinders the processes of ionic exchange.

It thus becomes possible to apply to metals the method so often used in chemistry to obtain evidence of the orientation of bonds; namely, that of examining the degree of ease with which they break down to give place to new configurations. In metals, the energy of cohesion and the work of extraction, although of rather indeterminate significance, are the result of a number of factors. Also the kinetics of the exchange processes of ions makes their bond conditions important, both in the lattice and in the solution; in addition, the surface condition of the metal, and, in general, the structure and properties of the interphase, are important.

The information regarding chemical resistance in the process of ionic exchange that can be obtained from the study of overvoltages, is therefore still difficult to interpret, quite apart from the present difficulties of experimental technique. However, it is possible to say that this direction of research is full of promise.

In spite of the serious lack of experimental data at our disposal at present, concordant results yielded by the various methods of investigation mentioned above, show that the chemical resistances in the ionic-exchange processes are: (1) negligible in the case especially of mercury, lead, tin, cadmium, thallium (the so-called metals of normal behaviour); (2) important, or very large, for the transition metals that have been studied adequately up to the present time: iron, cobalt, nickel, metals of the platinum group (metals electrochemically inert); and (3) intermediate for the groups: silver, copper, gold, and bismuth, antimony, arsenic, in each of which the first-named metals (silver and bismuth) are practically normal while the chemical resistance (anodic and cathodic) increases in the order given.

It is interesting to note that in each of the categories, there occur elements of quite different nobilities. Contrary, therefore, to current opinion, the kinetic ease of ionic exchange does not depend on nobility.

It is also interesting to observe the tendency to symmetry in the effects of anodic and cathodic overvoltages. It seems legitimate to speak of an inherent disposition to processes of ionic exchange, a reasonable estimate of which may perhaps be based on the rate of these exchanges in both directions in the absence of any external current. At ordinary temperatures and concentrations of the dissolved ions, the exchange rates so far determined extend from values of the order of 10^{-1} amp./cm.,^{2} for some normal metals, to very uncertain values, often lower than 10^{-10} amp./cm.,^{2} for the inert ones.

The properties of the metallic phase which are relevant to the kinetics of the ionic-exchange processes seem to be: (1) the possible existence of *d* sub-shells which are incomplete in the free atom (or the purely metalloid character of the bonds, as is the case of arsenic); and (2) the minimum value of the interatomic distances in the lattices. The first property seems essential for the appearance of electrochemical inertia, which increases, in each group, with decrease of the interatomic distances. Both for the metals of intermediate behaviour and for the normal ones, the minimum interatomic distance seems to afford again a characteristic indication of the degree of normality.

The existence of an electrochemical inertia which is essentially determined by the nature of the bonds in the phases (crystal lattice and solution) also seems to be of importance in explaining the phenomena of passivity of metals; not so much in the sense of identifying inertia with passivity, as in finding in inertia the ideal basis for establishing states of passivity, which always seem liable to lead back to the surface condition, as Faraday foretold and U. R. Evans and others have clearly demonstrated.[24]

A serious gap in the study of the electrochemical kinetics of metals is the fact that up till now experiments have been concerned almost entirely with polycrystalline electrodes, and the existing data represent mean values, whose interpretation is by no means simple, of those corresponding to the various orientations of the lattice planes. The new and fairly simple methods for preparing single crystals are now opening up a very interesting field of investigation.

Another field which has been very little explored, owing to its inherent difficulty, though well deserving of study, is that of the part played, both in anodic and cathodic processes, by the so-called "secondary structures."

The possibilities of investigation into the electrochemical behaviour of alloys are still practically unlimited, ranging from the more classic examples to those "crypto-alloys" which all the commercial metals are in reality owing to the presence of impurities.

VI.—CONCLUSION

In this lecture, which though lengthy is nevertheless in many respects incomplete, I have tried to describe those aspects of the development of the electrochemistry of metals which are most closely related to the science of metals. Let me remind you of a striking passage by Bacon[25]: "Qui tractaverunt scientias, aut empirici aut dogmatici fuerunt. Empirici, formicae more, congerunt tantum et utuntur; rationales, aranearum more, telas ex se conficiunt. Apis vero ratio media est, quae materiam ex floribus horti et agri elicit, sed tamen eam, propria facultate, vertit ad digerit."

The role of the bee is well suited to the electrochemistry of metals because of its character of an intermediate science. To extend the metaphor, the means of stimulating this increasingly fruitful collaboration between electrochemistry and the science of metals which I have predicted to-day, would perhaps be the opening to the electrochemical bees of the hothouses in which the beautiful flowers of the science of metals are grown.

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HIGH-TEMPERATURE THERMAL ANALYSIS USING THE TUNGSTEN/MOLYBDENUM THERMOCOUPLE *

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SYNOPSIS

A technique which enables thermal analysis to be carried out at temperatures up to 2000° C. has been developed. High-frequency induction heating is used, and the temperature is measured by means of a tungsten/molybdenum thermocouple, for which a calibration curve is given. The method has been applied to the determination of the freezing point of pure chromium, which is shown to be 1845° ± 10° C., and the liquidus curve for the solid solution of manganese in chromium.

I.—INTRODUCTION

DURING an investigation of the properties of several chromium-base alloys it became necessary to determine the freezing point of chromium and the liquidus curves of the systems under examination. It was decided that the induction heating method would be a satisfactory means of attaining the high temperatures involved and that a thermocouple would be the best means of measuring the temperature. The use of an optical pyrometer was dismissed because of vapour effects and of the personal factor in taking readings.

The work reported in the present paper includes the calibration of the tungsten/molybdenum thermocouple and examples of its use in determining the freezing point of chromium and the liquidus curve of the chromium–manganese system. The results are of special interest because of their recent determination by Carlile *et al.*,[1] using an optical pyrometer.

Some notes on the preparation of pure chromium have been included because of the importance of impurities in the determination of the freezing point and liquidus curves.

II.—THE TUNGSTEN/MOLYBDENUM THERMOCOUPLE

The use of tungsten/molybdenum thermocouples is not new, but there has been considerable disagreement between the temperature/thermo-e.m.f. curves given for them by the various previous users, and they have had the reputation of being unreliable. The early workers [2-6] have shown, however, that the thermoelectric power of the thermocouples is sufficiently great, above 1250° C., to allow reasonably accurate temperature measurement. It has been found that, with the better materials now available, a reproducible calibration curve can be established, and that after annealing the couples are quite stable. Although the tungsten and molybdenum wires are brittle after annealing, and require careful handling, their fragility is not a severe disadvantage to a careful worker.

The tungsten and molybdenum wires used were 0·05 in. in dia. and were obtained from commercial sources. Spectrographic examination indicated the presence of a trace of molybdenum in the tungsten and faint traces of magnesium and iron in the molybdenum, but the purity of both wires was of a high order. The wires were annealed in an atmosphere of hydrogen before use by passing a current of 65 amp. through them for 15 min. Thermocouples were made by welding in a tungsten arc. No protective atmosphere was used, but the operation was so rapid that very little oxidation took place, and good clean welds were obtained.

The thermocouples were calibrated over the temperature range 800°–2200° C. in two stages. From 800° to 1700° C. the temperature/e.m.f. relationship was established by direct comparison with a platinum/platinum–rhodium thermocouple; above 1700° C. the temperatures were measured by an optical pyrometer of the disappearing-filament type.

1. CALIBRATION FROM 800° TO 1700° C.

When using tungsten/molybdenum thermocouples, a protective atmosphere is necessary to prevent oxidation of the couple elements. Since the couples were required principally for use in hydrogen, the calibration was carried out under the same conditions. Although hydrogen is a more suitable medium than either air or vacuum in which to use platinum/

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platinum–rhodium thermocouples at high temperatures,[7] it involves the danger of contamination of the thermocouple if it is used in conjunction with silicon-bearing refractories in an apparatus in which carbon and sulphur are likely to be present.[8] In calibrating the tungsten/molybdenum thermocouple against a platinum / platinum – rhodium thermocouple, this danger was avoided by the use of a heating apparatus in which no refractory materials were present.

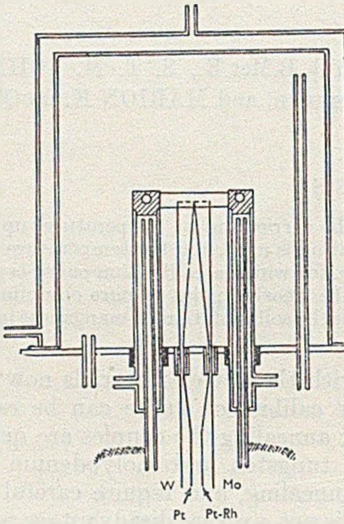

Fig. 1.—Molybdenum Strip Heating Apparatus for Calibration of Thermocouple from 800° to 1700° C.

A diagram of the apparatus is shown in Fig. 1. The heater consisted of two molybdenum strips 0·008 in. thick and about ⅜ in. wide, arranged in parallel in the form of a bow between two water-cooled electrodes spaced about 2½ in. apart. The strips, which were heated directly by resistance, acted as radiation shields as well as heaters, and very high temperatures were quickly and easily attained in the small space between them. In this space the hot junctions of the thermocouples were arranged. In order to ensure that the junctions of the two couples being compared were at the same temperature, the four wires were welded together to form a single bead. The electrodes were mounted in a base-plate, on to which was sealed a water-cooled copper bell which covered the heater assembly. The cold junctions of the thermocouples were allowed to remain at room temperature, which, during the course of the experiments, differed very little from 21° C. Any alternating currents picked up by the thermocouples were filtered out by means of a network of condensers and inductances.

The heating current was supplied by a transformer giving an output ranging from 0 to 15 V. At any transformer setting, thermo-e.m.f. readings were taken for the two couples alternately. The readings were sometimes found to vary slightly owing to fluctuations in the mains voltage, but these effects were overcome by taking a number of readings for each couple and using the means of all the readings taken. The thermo-e.m.f. readings given by the platinum/platinum–rhodium thermocouple were converted to temperatures, and plotted against the corresponding thermo-e.m.f. reading given by the tungsten/molybdenum thermocouple. A smooth and reproducible curve was obtained from 800° to 1700° C.

2. Calibration from 1700° to 2200° C.

Since there was no alternative to optical or radiation pyrometry for measuring temperatures above the range of the platinum/platinum–rhodium thermocouple, the disappearing-filament type of optical pyrometer was chosen as the most reliable and convenient means. The uniform temperature enclosure in which the thermocouple junction was heated was formed by three molybdenum blocks about ¾ in. square and ⅜ in. thick, arranged as shown in Fig. 2. The thermocouple elements were not welded together, but were bent to a right angle at a distance of about $\frac{3}{16}$ in. from one end, and were dropped through holes in the lower molybdenum block so that the short bent ends lay along a groove in the surface of the block running between the two holes, the remainder of the wires emerging downwards. Since the wires were not in contact with any part of the block other than the groove, the temperature of the hot junction could be safely assumed to be that of the grooved part of the block. The other blocks were placed on top of the grooved one and the three held together with a wrapping of molybdenum sheet. Holes drilled in the centres of the upper blocks formed a cavity which radiated effectively as a black body.

The blocks were mounted in a sealed silica tube. A totally reflecting prism above a window in the upper cover-plate of the tube was arranged so that the optical pyrometer might be conveniently sighted on the black-body cavity. The effect of the window and prism on the optical-pyrometer readings was measured by means of a series of temperature determinations made on a calibrated tungsten-strip lamp with and without the window and mirror in the optical system, at temperatures from 1000° to 2000° C. A correction curve was thus established which enabled the optical-

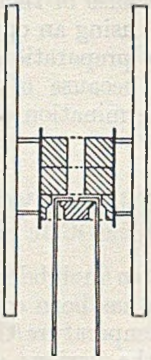

Fig. 2.—Molybdenum-Block Heater Assembly for Calibration from 1700° to 2200° C., Showing Arrangement of Thermocouple Elements.

pyrometer readings to be simply converted to true temperatures.

The molybdenum-block assembly was heated directly by high-frequency induction, any high-frequency currents picked up by the thermocouple being filtered out. Purified hydrogen was passed slowly through the apparatus during the experiment. The cold junctions of the thermocouple were again allowed to remain at room temperature. Using a pyrometer newly calibrated against a tungsten-strip lamp, it was found that the temperature/thermo-e.m.f. relationship obtained in this way agreed well with that obtained by comparison with a platinum thermocouple from 1000° to 1700° C. The calibration was then continued to 2200° C.

3. The Temperature/Thermo-E.M.F. Curve

The calibration curve obtained for the tungsten/molybdenum thermocouple is shown in Fig. 3. At lower temperatures the curve is parabolic in form, but above 1500° C. it becomes linear. Below 1250° C. the thermo-e.m.f. generated is positive, decreasing to zero at 1250° C. and thereafter becoming increasingly negative. The reproducibility of the curve is about ± 3° C. below 1700° C. and ± 5° C. above this temperature. Its accuracy has been checked at the platinum point (1774° C.) by carrying out a cooling curve on a melt of pure platinum. The thermocouple gave a reading of 1776° C. at the freezing point, and its errors, therefore, lie within the limits of accuracy of the calibration of the optical pyrometer at this temperature.

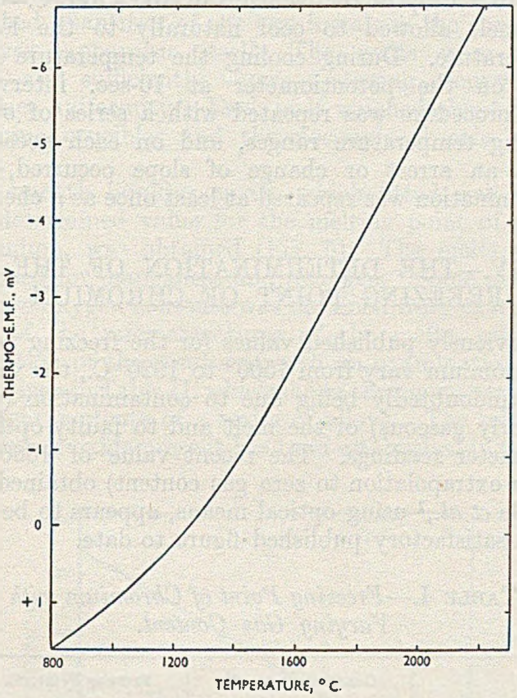

Fig. 3.—Calibration Curve for Tungsten/Molybdenum Thermocouple.

It has been found that couples made from the same batch of material give reproducible calibration curves, and it is not necessary to calibrate each couple individually once the standard curve has been established. Furthermore, results obtained on a new batch of material indicate that there is no significant difference between its calibration and that of the earlier batch. Repeated heating to temperatures of about 2000° C. does not affect the couple once it has been annealed.

4. Contamination of the Thermocouple

Contamination of tungsten/molybdenum thermocouples is observed if they are heated in contact with beryllia, metallographic examination showing that severe intergranular penetration occurs. When the use of beryllia is unavoidable, it has been the practice to calibrate the couples before and after each significant experiment. Alumina has been found to cause no contamination of the thermocouple, but its low melting point (2050° C.) limits its usefulness. Experiments have recently been carried out on the use of magnesia. After heating to about 1900° C. in contact with magnesia, the thermocouple bead shows slight intergranular contamination near the surface, but this seems to have no marked effect on the calibration of the thermocouple. The contamination is very much less than that caused by beryllia under similar circumstances.

III.—METHOD OF THERMAL ANALYSIS

The heating unit was a 20-kW. high-frequency generator manufactured by Amalgamated Wireless (Australasia), Ltd. The operating frequency is in the range 300-500 kc./s., depending on the characteristics of the work being heated. The power input to the work coil can be varied from about 20 to 100% of the power available for a given set of electrical conditions by a controlled positive voltage applied in opposition to a permanent negative bias on the grids of the oscillator valves.

The furnace arrangement is shown in Fig. 4. The alloy under investigation was placed in the crucible, *A*, which was supported at the mid-section of a cylindrical molybdenum heater, *B*, by means of an upturned crucible resting on a molybdenum wire grid, *C*. The heater was 8 in. long and 1·2 in. in dia., and was fabricated from 0·008-in.-thick molybdenum sheet by longitudinal spot welding with a tantalum strip interlayer. The heater was supported in turn on a section of molybdenum strip, *D*, resting on a fired alumina insulating piece, *E*. The whole of this assembly was contained in a silica tube, *F*, 30 in. long by 2 in. inner dia. The heater was located centrally in this tube by means of four molybdenum wire spacers, *G*, and was energized by induction from the work-coil, *H*, of the generator. The tube, *F*, was closed with water-cooled brass end-plates, *K*.

The thermocouple bead was protected by a sheath, *N*, which was held centrally in the crucible by a molybdenum wire spacer, *P*. This spacer also served to locate the crucible centrally in the heater. The thermocouple wires were led out through a twin-bore silica insulator, *R*, to a pair of shielded extension leads, *S*, and thence through a radio-frequency filter, *T*, to a Negretti and Zambra quick-reading potentiometer. The filter consisted of an array of radio-frequency chokes and condensers which was necessary to eliminate stray e.m.f. induced in the thermocouple by the H.F. current. Extension of the cylindrical heater well beyond the limits of the work-coil helped considerably to reduce these e.m.f.'s and also had the advantage of enabling the grid, *C*, and wire spacers, *G*, to be located outside the central hot zone. Failures of the heater at these points were frequent during earlier work when heating tubes 2·5 and 4 in. long were used. As a final precaution against inductive effects, one side of the thermocouple circuit was earthed, and the thermocouple wires were kept close together over the whole of their length, thus eliminating inductive "loops" in the vicinity of the work coil. The possibility of interfering e.m.f. could always be checked by momentarily shutting off the power supply and observing whether or not there was a corresponding jump in the potentiometer reading. With the arrangement shown in Fig. 4, it was usual to find no interference of any kind. An atmosphere of purified hydrogen was maintained throughout a determination, the gas passing through at a slow constant rate.

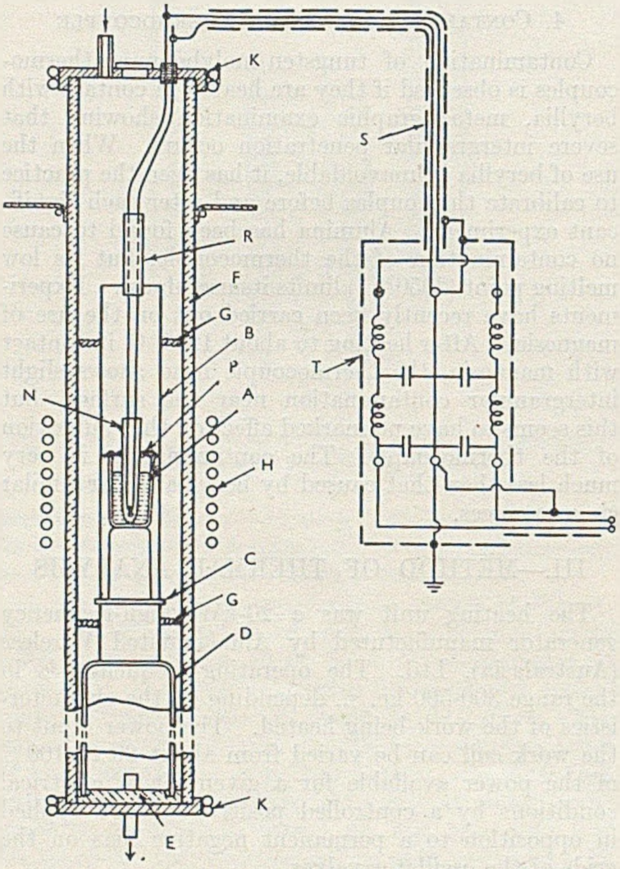

Fig. 4.—Furnace Arrangement for Thermal Analysis.

For experiments in which the maximum thermocouple reading did not exceed about 1750° C., the crucible and thermocouple sheath were made of alumina. For maximum temperatures up to 1900° C., a beryllia crucible was used, and above this temperature both crucible and sheath were of beryllia or magnesia. When a combination of alumina and beryllia was used, care was taken to prevent contact between them in the hot zone, thus avoiding the formation of a low-melting mixture. Whenever a beryllia sheath was used it was necessary to recalibrate the thermocouple immediately after the experiment, as noted before (Section II, 4).

To determine a cooling curve after the melting of an alloy, the power input to the work-coil was quickly reduced to a predetermined setting—which would lead to a drop of temperature of 100°–200° C.—and the melt allowed to cool naturally to the lower temperature. During cooling the temperature was read on the potentiometer at 10-sec. intervals. This procedure was repeated with a series of overlapping temperature ranges, and on each occasion when an arrest or change of slope occurred, the determination was repeated at least once as a check.

IV.—THE DETERMINATION OF THE FREEZING POINT OF CHROMIUM

Previously published values for the freezing point of chromium vary from 1500° to 1920° C., the variation undoubtedly being due to contamination (particularly gaseous) of the melt and to faulty optical-pyrometer readings. The recent value of 1860° C. (after extrapolation to zero gas content) obtained by Carlile *et al.*,[1] using optical means, appears to be the most satisfactory published figure to date.

Table I.—*Freezing Point of Chromium with Varying Gas Content.*

Sample	Gas Content, wt.-%		Freezing Point, °C.	
	O_2	N_2	1st run	2nd run
1	0·02	0·014	1839	1841
2	0·02	0·033	1835	...
3	0·02	0·076	1834	1832
4	0·12	0·029	1831	...
Carlile *et al.*[1]	0·14	0·011	1845	

Using the apparatus and method described in the Appendix with melts of approximately 35 g. weight, a beryllia crucible, and an alumina sheath, the freezing points shown in Table I were determined. For comparative purposes the recent value obtained by Carlile *et al.*[1] is also given.

Extrapolations of these four results to zero gas contents have given the freezing point of pure chromium as 1845° C. Taking into account the experimental errors, this value is estimated to be true to ± 10° C.

The determination of nitrogen was made by a modification of the Allen distillation method. Oxygen was determined by the chemical method described in the Appendix, a correction being made for the insoluble nitrogen content.

V.—THE LIQUIDUS CURVE OF SOME CHROMIUM-MANGANESE ALLOYS

The method of thermal analysis described in Section III has also been used for the study of the liquidus curve of chromium-manganese alloys containing up to about 60 at.-% manganese. Electrolytic chromium and manganese were used, the chromium being prepared as described in the Appendix.

The alloys were prepared by melting together weighed quantities of the two metals in the thermal-analysis apparatus itself. The melts were very fluid, and sufficiently vigorous stirring occurred to produce uniformity. Each melt weighed about 30 g. Alumina crucibles and thermocouple sheaths were used.

A smooth liquidus curve, which fitted well with the determined value for the melting point of pure chromium, was obtained (Fig. 5). The melts were analysed for manganese and nitrogen. Some idea of relative oxygen contents was obtained from examination of the microstructures of the melts, and an indication of the results of such an examination has been given in Table II, which shows the manganese and nitrogen contents of a series of alloys. The relative amount of oxide present has been established in terms of the average amount present in the unalloyed chromium.

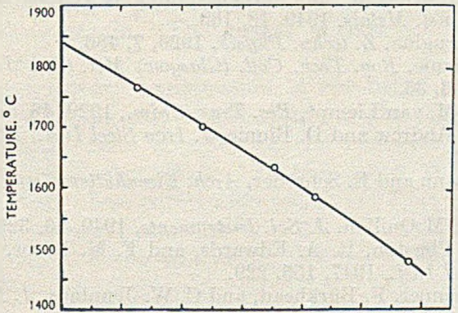

Fig. 5.—Liquidus Curve for the Solid Solution of Manganese in Chromium.

As a matter of interest, the approximate difference between the present results and those recently presented by Carlile *et al.*[1] for the same alloys have been included. An explanation of the observed differences was sought, unsuccessfully, in the gas contents of the alloys. No correlation was found between the gas contents of the alloys and the differences between the two sets of results. It may be seen in Table II that two alloys of closely similar manganese content show the same value for the liquidus temperature, even though the nitrogen content of one is more than twice that of the other. Furthermore, a specimen which shows one of the greatest differences between the present results and those of the earlier workers, has appreciably less oxide in its microstructure than any other specimen. The explanation is probably to be found in the different methods of temperature measurement used, the arrest points obtained by the authors having been obtained by thermocouple measurements, whereas Carlile *et al.* used an optical pyrometer.

Table II.—*Liquidus Temperature and Nitrogen and Relative Oxide Content of Chromium–Manganese Alloys.*

Mn Content, at.-%	N_2 Content, at.-%	Relative Amount of Oxide Present as Compared with Amount in Unalloyed Cr	Difference Between Authors' Determination of Liquidus Temp. and that of Carlile *et al.*,[1] °C.
57·8	0·027	Less	22 lower
42·0	0·014	Slightly less	35 ,,
41·5	0·038	,, ,,	35 ,,
35·3	0·018	Very much less	30 ,,
23·4	0·052	The same amount	35 ,,
12·5	0·037	,, ,, ,,	20 ,,

ACKNOWLEDGEMENTS

The work described above was carried out as part of the high-temperature materials programme of the Aeronautical Research Laboratories, Department of Supply, Melbourne. The authors wish to thank the Chief Scientist of the Department for permission to publish.

All the chemical analyses were done by Mr. E. J. T. Lumley of these laboratories, to whom acknowledgement is made. The authors are indebted to Dr. N. P. Allen of the National Physical Laboratory for the vacuum-fusion analyses of the chromium.

APPENDIX

The Preparation of Pure Chromium

As high-purity chromium was not readily available, it was necessary to investigate means of producing the metal in the quantities required for the high-temperature materials programme of the Aeronautical Research Laboratories. Two methods were considered: (i) the electrolysis of chromic acid and (ii) the reduction of chromic oxide. The former was chosen, since it will produce, using suitable apparatus, metal containing only non-metallic impurities. The reduction of chromic oxide has one serious disadvantage—the probability of contamination of the product by the reducing element.

Electrolytic Production of Low-Oxygen Chromium

The apparatus employed was of the type used in normal electroplating, except for the following features:

(*a*) The acid bath and anode were made from pure lead, since antimonial lead was found to contaminate the electrodeposit with approximately 0·2% antimony.

(*b*) Polyvinyl chloride tubing was used for stopping off the cathode—a $\frac{5}{8}$-in.-dia. copper tube—since the commercial waxes available were unsuitable for the plating temperatures employed.

The bath contained 250 g./l. commercial chromic acid and 2·5 g./l. sulphate ion. Its composition was controlled chemically and additions made as required. The solution level was maintained constant, and the temperature thermostatically controlled to ± 1° C.

Spectrographic examination of chromium produced in this bath detected no metallic impurities. Gaseous impurities were hydrogen and oxygen; the latter was believed to be present in the form of hydrous chromic oxide. The hydrogen was readily evolved on heating to 600° C., leaving oxygen as the remaining impurity. Brenner *et al.*[9] have shown that the oxide content of electrolytic chromium varies with plating conditions; temperature and current were therefore varied to determine the conditions which would yield the lowest oxygen content in the " as-deposited " condition. The results are given in Table III.

TABLE III.—*Variation of Oxygen Content with Temperature and Current Density.*

Current Density, amp./ft.²	Temperature, °C.	Oxygen, wt.-%
1100	68	0·23
1100	73	0·16
1100	75	0·13
1100	80	0·06
1100	83	0·08
850	80	0·07
960	80	0·07
1100	80	0·06
1250	80	0·06

From these results it is evident that the oxygen content varied only slightly with current density, but fell rapidly with increase in temperature. On this basis the conditions chosen for the production of all the chromium used in the programme were a temperature of 80° C. and a current density of 1100 amp./ft.².

Oxygen was determined by chemical means, which consisted essentially of dissolving a 1-g. sample in 10% hydrochloric acid, then determining the amount of chromium in the residues and converting to oxygen. This first necessitated a vacuum treatment at 850° C. to convert the soluble oxygen content of electrodeposited chromium into insoluble chromic oxide (Cr_2O_3). The results of vacuum-fusion determinations as compared with the results of chemical analyses are given in Table IV. The low figures obtained by chemical analyses of samples *A* and *C* in this table indicate the need for vacuum treatment before analysis.

TABLE IV.—*Comparison of Oxygen Content Determined by Vacuum-Fusion Method and Chemical Analysis.*

Sample	Condition	Oxygen, wt.-%	
		Vacuum Fusion	Chemical Analysis
A. Low-O_2, bath temp. 80° C.	As deposited.	0·06	0·02
B ,,	Sample *A*, vacuum treated 1 hr. at 850° C.	0·05	0·05
C. High-O_2, bath temp. 68° C.	As deposited.	...	0·04
D. ,,	Sample *C*, vacuum treated 1 hr. at 850° C.	0·23	0·22

The nitrogen content of the chromium was less than 0·001%, as determined by vacuum fusion.

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SLIP BANDS AND HARDENING PROCESSES IN ALUMINIUM*

By A. F. BROWN,† M.A., Ph.D.

SYNOPSIS

Slip bands on aluminium increase in number during plastic deformation and, at the same time, further slip occurs within each band. At higher temperatures and lower rates of deformation, as well as with increasing strain under all conditions, the latter process becomes increasingly predominant. This is interpreted on the basis of the fine structure of slip bands which has been resolved by the electron microscope. The differences in density and inner structure of slip bands formed under different conditions are compared with the differences between stress/strain curves, and it is shown that slip which forms a new band involves much more macroscopic hardening than slip within an existing band. A consequence of this conclusion is that a mechanical equation of state can exist only at very small strains.

I.—INTRODUCTION

When a metal crystal is deformed, the deformation is at first elastic and obeys Hooke's law: stress is proportional to strain. If the stress is progressively increased, the elastic limit is reached, after which the metal does not return to its original shape when the stress is released and increase of stress does not lead to a proportional increase of strain. The metal is then said to deform plastically.

In the elastic range, deformation is homogeneous, all atomic planes being strained alike, but in the plastic range all the deformation is concentrated on a few atomic planes; the crystal acquires a stepped appearance and the steps are visible on the surface as slip bands. However, in spite of the change in the character of the deformation, the elastic limit is not in general marked by an abrupt change in the slope of the stress/strain curve, and the stress needed to produce further plastic shear continues to rise with increasing strain. The metal is said to be work-hardening.

A satisfactory explanation of the origin of plastic work-hardening and of the shape of the stress/strain curve must take account of the observed inhomogeneity of plastic strain. In this paper it is shown that a very close correlation exists between the shape of the stress/strain curves and the differences in appearance of the slip bands formed under different conditions of temperature and strain rate.

In Section II of the paper experiments on aluminium are described which relate the amount of work-hardening to the number of slip bands and to the amount of slip within each. Plastic strain may arise either as a result of slip which produces new slip bands or as a result of slip producing extra shear on existing bands.

In Section III the experimental results are all summed up in the form of a simple picture of the variation of stress in the space between slip bands. The sites where slip is observed most frequently to take place are indicated as places where the internal stresses are small. From this picture the result is obtained that slip on existing bands contributes little to the macroscopic stress. It is shown in Section IV that the form of the stress/strain curve under different conditions of temperature and strain rate can be accounted for in terms of this picture; a quantitative expression for the curve is obtained and gives good agreement with the evidence available for aluminium. Finally, in Section V, it is shown that a mechanical equation of state cannot exist under conditions which permit the strain within a slip band to increase after the formation of the band.

In a later paper it will be shown that similar conclusions are valid for several metals other than aluminium.

II.—EXPERIMENTS ON THE FORM AND DISTRIBUTION OF SLIP BANDS

1. Yamaguchi's Experiment

The classic experiment to show how the distribution of slip bands in aluminium is related to the amount of work-hardening was made by Yamaguchi in 1928.[1] The method was to polish single crystals of the purest aluminium then obtainable and extend them in a tensile testing machine. The experiment was stopped at intervals, the stress and strain recorded, and a count made of the number of slip bands per unit length in the direction normal to the slip planes. Slipping on one system only was allowed and if, with increasing strain, the specimen reached an orientation such that a second slip system became active, the experiment was stopped. With some specimens shear strains of 70% were reached.

Yamaguchi found that the number of slip bands per unit length was not proportional to strain, as would

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be the case if each band contributed a fixed amount of slip to the strain. However, the number of bands was related to the shear stress by an equation of the form:

$$\tau - \tau_0 = kN$$

where N is the number of bands per mm. counted as described above, τ is the corresponding yield stress, and k is a constant of magnitude about 5×10^{-3} kg./mm.^{2}. τ_0, the yield stress corresponding to the elastic limit, is here defined as the stress which produces the first slip band. A curve of τ against N plotted from a few of Yamaguchi's measurements is shown in Fig. 1 (*a*).

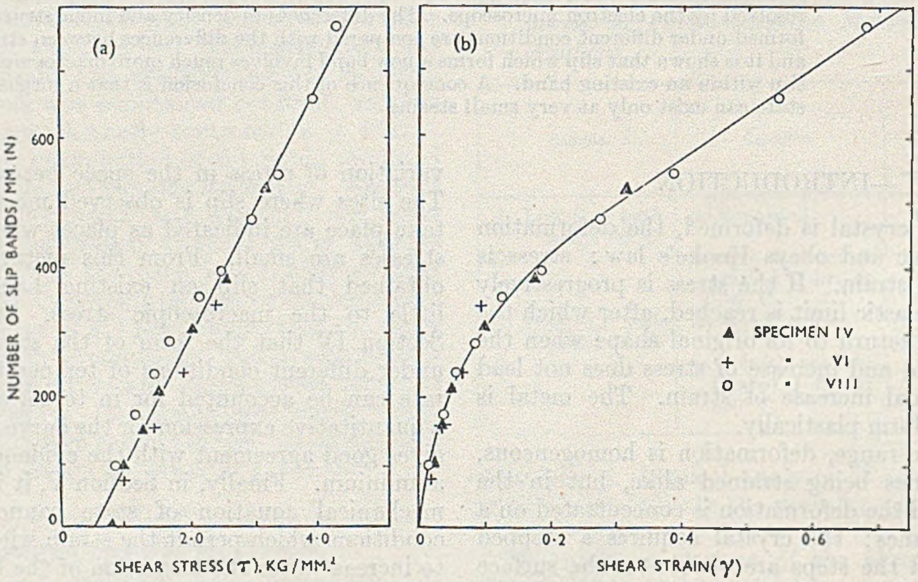

Fig. 1.—Curves from Results Obtained by Yamaguchi,[1] Relating the Density of the Slip Bands to (*a*) Stress and (*b*) Strain.

Yamaguchi's results are given in great detail in his paper, and it is possible to obtain additional information from them. Thus the relationship between N and the shear strain (γ) has been extracted and is shown in Fig. 1 (*b*); this curve has the shape of the stress/strain curve. In Fig. 2, the amount of slip per band has been calculated from Yamaguchi's results and plotted as a function of the macroscopic shear strain. The resulting curve shows immediately that fresh slip occurs on existing bands at the same time as new bands are being formed.

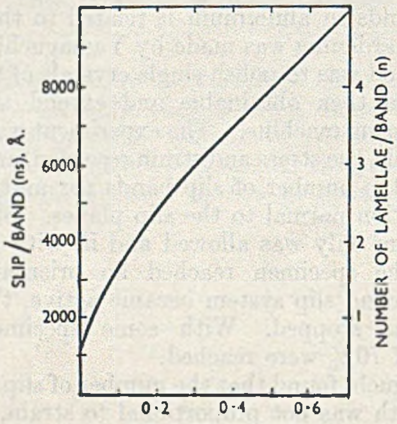

Fig. 2.—Curve from Results Obtained by Yamaguchi,[1] Relating Amount of Slip Per Band to Shear Strain.

It has been objected that Yamaguchi's procedure of counting slip bands could not give significant results, since the slip bands did not run right across the crystal but stopped or even ran into each other; however, it can be shown from Yamaguchi's photographs that the number of slip bands counted crossing a mark parallel to the axis of the specimen is independent of the position of the mark, provided that the number of bands is sufficiently great. Moreover, it is now possible to measure directly the amount of slip per band and to show that the values thus obtained agree well with those calculated from Yamaguchi's figures on the assumption that all bands make the same contribution to the macroscopic strain, at a given stress.

A recent attempt [2] to repeat Yamaguchi's experiments using aluminium of the highest modern purity gave the result that, up to 2% strain, both stress and strain are almost linear functions of the number of bands. (The authors of the paper are, however, concerned with the deviations from linearity.)

2. Fine Structure of the Slip Bands

The resolution of the light microscope is not sufficient to give definite information about the profiles of slip bands, i.e. the amount of slip occurring, or the number of crystallographic planes active in each. Recently, however, Heidenreich and Shockley [3] have shown by means of the electron microscope that the slip bands on aluminium can be resolved into bundles of finer lines as shown in Fig. 4. (Plate XXIV). These are interpreted as being the traces of slip lamellæ such as are shown schematically in Fig. 3. The slip distance per lamella, measured in the slip plane and slip direction is about 2000 Å., while the width of the lamellæ, i.e. the distance between active slip planes, is of the order of 200 Å.

Extended observations have now been made with the electron microscope on the variation with strain of the separation and fine structure of slip bands in aluminium deformed at different temperatures. The methods employed and the results obtained have been described elsewhere.[4] The results for room-temperature deformation agree with those given by Heidenreich and Shockley, and no evidence was found of a finer structure. If such a structure exists, it must have dimensions of the order of a few atoms.

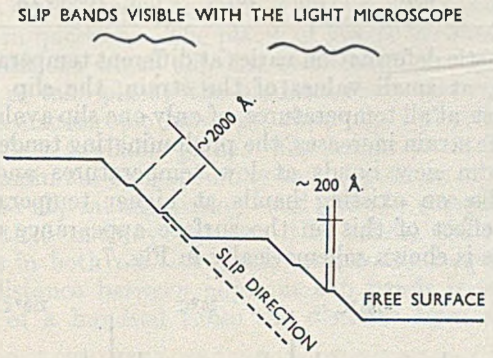

Fig. 3.—Schematic Representation of Fine Structure of Slip Bands.

Within the limits of accuracy with which electron micrographs from metal replicas can be measured,[4] each lamella slips over its neighbours by approximately the same distance; in aluminium of the highest purity, measured distances range from 1600 to 2200 Å. There is no evidence either for uncompleted slip processes or for multiple slip processes on the same plane. Thus it may be concluded that slip takes place, under normal conditions, in avalanches and that the distance moved as a result of one avalanche is about 2000 Å.

Experiments on aluminium of lower purity [5] have given values of the slip distance between 1000 and 1200 Å. Thus, while the slip distance per lamella is rather less than in the pure metal, the range of values is still small. There is again no evidence that slipping of the lamellæ relative to each other can be otherwise than by avalanches.

At increasing values of the strain, the electron microscope shows that there are more lamellæ in each slip band, but that the amount of slip corresponding to the formation of each lamella is unchanged (within the limits specified above) from the smallest strains up to strains of the order of 90%. Thus the additional slip which Fig. 2 shows to occur within existing slip zones takes place by avalanches of length always about 2000 Å. on planes spaced some 200 Å. apart and leading to the formation of new lamellæ.

The numbers on the right-hand axis of Fig. 2 represent the number of lamellæ in each band, and are obtained from the figures on the left-hand axis by assuming that each lamella contributes just 2000 Å. of slip and that all bands in a uniformly sheared single crystal have the same number of lamellæ. The latter assumption is justified by experiment; the number of lamellæ in bands on a specimen deformed 40% will be normally about four, the observed range being 3–5. Thus bands formed late in the deformation rapidly catch up in total slip with those formed in the early stages. Experiment gives reasonable confirmation of the number of lamellæ calculated in the above manner.

At magnifications sufficient to resolve the lamellæ, the field of view of the electron microscope is not large enough to take in more than one or two slip bands. Thus, it is not possible to show in one photograph how the distribution of bands and, at the same time, their internal structure vary with strain. However, the evidence obtained from a number of electron micrographs of individual bands is shown diagrammatically in Fig. 6 where the lamellar spacing has been exaggerated some ten times relative to the band spacing. The diagram indicates that, at small strains, the bands are widely spaced and contain only one avalanche each. As the strain increases, more bands are formed in positions approximately midway between existing bands, and at the same time, fresh avalanches occur in the bands, giving rise to lamellæ.

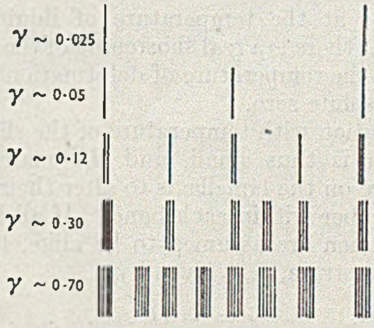

Fig. 6.—Development of a Field of Slip Bands at Room Temperature.

3. Effect of Temperature on the Slip Bands

The light microscope reveals differences between the slip bands on specimens which have been deformed at different temperatures. Bands formed in deformation at the temperature of liquid air appear as closely spaced, fine lines. As the temperature of deformation is raised, the spacing for a given strain increases until at high temperatures the bands are widely spaced and broad. Thus, with increasing temperature the same strain is concentrated into fewer and fewer bands.

The electron microscope [4] also enables this observation to be interpreted on the basis of formation of lamellæ. At low temperatures there are many bands, each containing few lamellæ, while as the temperature of deformation is raised fewer bands are observed, with a corresponding increase of the number of lamellæ in each. The distance slipped in the formation of each lamella, i.e. the avalanche distance, does not alter appreciably with temperature. Values for this distance obtained by averaging observed values for many specimens are given in Table I.

TABLE I.—*Distance Slipped [Avalanche Distance] in Formation of Each Lamella.*

Deformation Temperature, °C.	Avalanche Distance, Å.	
	Mean	Limits Observed
−180	1500	1200–1800
20	2000	1600–2200
250	2200	1700–2500

Column 2 in this table gives the mean value obtained from 20 to 30 specimens for each temperature. Column 3 gives the limiting values measured; the spread is due in great part to the difficulty of measuring poorly contrasted lines on micrographs. The significance of the difference between the mean slip distances at 20° and 250° C. is doubtful, particularly since mean values higher than 2200 Å. are not found in experiments at temperatures above 250° C. However, experiments at the temperature of liquid helium [5] confirm that there is a real shortening of the avalanche distance as the temperature of deformation is reduced towards absolute zero.

The variation with temperature of the slip distance per lamella is thus small, and the main effect of temperature on the lamellæ is to alter their grouping, i.e. the number within each band. If all bands in a given specimen are assumed to be alike, the macroscopic shear strain, γ, is given by:

$$\gamma = ns/D \quad . \quad . \quad . \quad . \quad . \quad (1)$$

where n is the number of lamellæ per band, s is the avalanche distance, i.e. slip distance per lamellar slip, and D is the spacing of the slip bands measured, as usual, in a direction normal to the slip planes.

If, now, the small variations between the values of the avalanche distance, s, at temperatures between − 180° and 250° C. are neglected:

$$n/D = \text{a constant for any given strain.}$$

How far this relationship is realized in practice is shown in Table II, where observed values of the number of lamellæ per band (n) and the slip-band spacing are tabulated for various temperatures of deformation. The figures are for aluminium specimens which have undergone a shear strain of about 40%.[4] Greater accuracy of agreement than that shown in Table II cannot be expected owing to the difficulty of obtaining truly representative fields of slip bands with the small field of view of the electron microscope.

TABLE II.—*Variation of Number of Lamellæ per Band and the Slip-Band Spacing with Deformation Temperature.*

Temperature of Deformation, °C.	Approx. Distance between Neighbouring Bands, μ	No. of Avalanches per Band (n)
−180	½–1	1–2
20	2	3–4
250	4	5–6
500	10	About 12

Thus the experimental evidence is that the geometry of plastic deformation varies at different temperatures. First, at small values of the strain, the slip bands consist, at all temperatures, of only one slip avalanche. As the strain increases, the predominating tendency is to form new bands at low temperatures and new lamellæ on existing bands at higher temperatures. The effect of this on the surface appearance of the bands is shown schematically in Fig. 7.

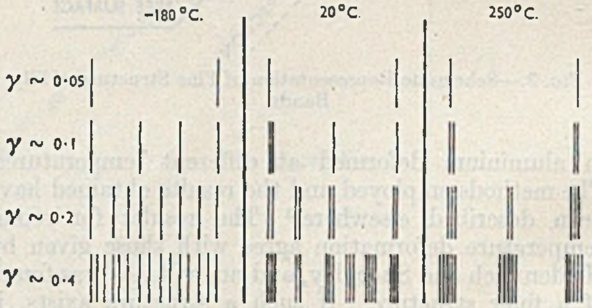

FIG. 7.—Comparison of Development of Slip Bands at Different Temperatures of Deformation.

4. THE EFFECT OF VARYING THE RATE OF DEFORMATION ON THE SLIP BANDS

When the rate of deformation is reduced, the slip bands are found to be further apart so that, for a given total strain, there is more slip in each band. An extreme case is illustrated in Figs. 5 (*a*) and (*b*) (Plate XXIV): here a single crystal of aluminium has been extended at 450° C. at a constant strain rate of 1% per day. Fig. 5 (*a*) is a low-magnification photograph of the gross slip bands, some of which were found by differential focusing to be 10 μ in vertical height. Fig. 5 (*b*) shows part of the lamellar structure of one of these bands. In such conditions of high temperature and slow strain the lamellæ are no longer parallel, but the amount of slip corresponding to each is the same and is about 2000 Å. Although several days have elapsed between the formation of some of the lamellæ shown, and in spite of the high temperature, there is no evidence that more than one slip process has anywhere occurred on one slip plane.

III.—THE RELATIONSHIP OF SLIP BANDS TO WORK-HARDENING

After plastic deformation has begun and slip bands have appeared, further increase in the strain is by means of slip processes that take place, either on planes approximately midway between, or on planes situated very close to, the existing slip bands. Slip on other planes is unlikely, especially on those planes which have already slipped. These experimental observations are summarized in Fig. 8, which indicates what must be concluded to be the variation of hardness between two slip bands. The distance between the bands is plotted horizontally, while the vertical co-ordinate represents hardness, defined as the stress needed to force a slip process across the plane in question. The places of lowest hardness are the positions where slip has been found to be most likely to take place; the hardness there corresponds to the macroscopic hardness, since slip will occur on the softest available plane. The height of the hardness minimum corresponding to the position of fresh lamellar slip on an existing band must be comparable with that midway between the bands, since slip occurs in both positions with comparable frequency. The distance between adjacent slip bands is of the order of a hundred times the distance from a slip band to the nearest plane of lamellar slip, so it is necessary, in the figure, to exaggerate the latter distance. Fig. 8 also shows the positions where slip is never observed to occur; these are the slip planes themselves and the hardness maximum on the other side of the lamellar trough. The full-line curve can be regarded as the sum of the separate hardening contributions of each of the two bounding slip zones. These latter curves are shown by dotted lines and represent the manner in which the hardening falls off with distance from the slip band. There are similar curves for the hardening on each side of every slip band.

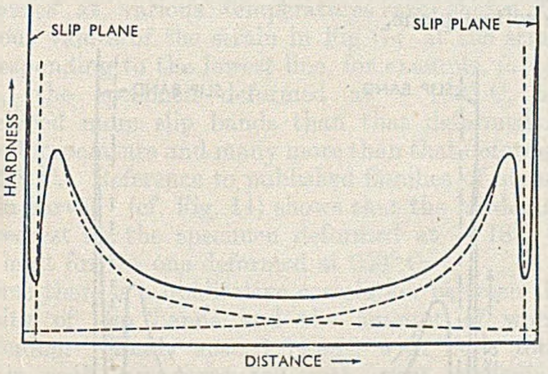

Fig. 8.—The Variation of Hardness Between Two Slip Bands.

1. The Reason for Planes of Easy Slip

The passage of a slip avalanche leaves the slip plane so disordered that slip never takes place on it again. At the same time internal stresses are set up throughout the metal which, as a result becomes harder throughout; higher stress is now required to force another slip avalanche through it. The stress required falls off with distance from the original slip plane in a manner such as is shown in Fig. 9. This curve is of the same form as the dotted curves of Fig. 8, but without the minimum close to the slip plane. Fig. 9 is not to be regarded as a curve giving the diminution of internal stress with distance from a single disordered plane, since the internal stresses that produce hardening cannot be all of one sign. On the contrary, Fig. 9 shows the diminution with distance of the hardening effect of internal stresses of rapidly varying sign produced by dislocations created during the slip avalanche but not on the apparent slip plane. It could perhaps be regarded as the envelope of the local stress peaks. The analytical shape of the curve will be discussed later.

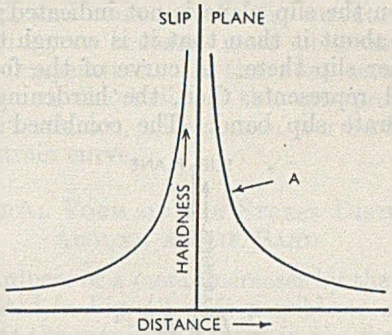

Fig. 9.—The Hardening Due to a Slip Process.

Now experiment shows that there are planes of easy slip in positions where, as a result of the slip process, there should be the very high hardening stresses indicated in Fig. 9. By some process, therefore, the hardest parts must be softened. The means by which this occurs may be self-annealing, as discussed by Paterson and Orowan[6]; as soon as the internal stresses reach a critical magnitude, depending on the temperature, self-annealing would lead to softening. If this critical value corresponds to the point *A* (Fig. 9), then self-annealing would be expected to result in a hardness curve of the form shown in Fig. 10, in which the former slip plane has now become a plane of relatively easy slip. However, experiment shows that the slip plane itself does not become a plane of easy slip, and the result of the softening

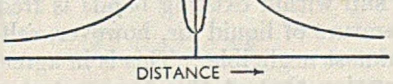

Fig. 10.—The Effect of Self-Annealing on Hardness Curve.

process is represented by the curve of Fig. 11. The hardness corresponding to the two troughs in this curve is comparable with the hardness midway between the bands. The amount of hardness remaining on the slip plane is not indicated; no more is known about it than that it is enough to prevent any further slip there. A curve of the form shown in Fig. 11 represents, then, the hardening effect of each separate slip band. The combined hardening effect of two adjacent slip bands is thus obtained by summing the ordinates of two such curves, with their origins spaced apart by the inter-band distance. The result is the full curve shown in Fig. 8.

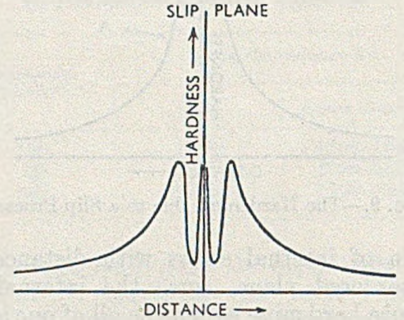

Fig. 11.—The Effect of Self-Annealing on Hardness Curve When the Slip Plane Remains Hard.

If each new slip process causes a hardening effect such as is represented by the curve of Fig. 11, then a new slip process occurring in the central trough of Fig. 8 will produce the result shown in Fig. 12. The addition of the new hardening results in a rise of the inter-band trough, and it becomes unlikely that a further new band can be formed until all the soft spots, close to the slip bands, are used up. A slip process, which occurs on one of these soft spots, has the same hardening effect on its own slip plane as a slip process which forms a new band. Its effect on the macroscopic hardness is different, however, since the metal on each side of it has reached limiting hardness and cannot be hardened further. The result is that the trough in which the slip occurred is wiped out, and there is no effect outside this trough. Thus the macroscopic hardness is unaltered, except by the circumstance that the elimination of one soft spot reduces the probability that further slip can find a spot of comparable hardness. The trough may, of course, be re-formed by further self-annealing.

The critical stress for self-annealing depends on temperature. At high temperatures the hardening due to a slip process readily reaches the critical value and then slip within existing bands is frequent. At the temperature of liquid air, however, self-annealing must be almost negligible. This is in agreement with the observation that at this temperature slip rarely takes place in existing bands until plastic shear strains of the order of 40% are reached.

As the strain increases each new slip process adds to the general hardening of the metal. Thus, at higher strains, it is more likely that the stresses produced by a slip process, added to the higher general stresses, will be sufficient for self-annealing. Slip within existing bands, therefore, becomes more frequent as the strain becomes greater.

The contribution of a slip process to the macroscopic hardening will not vary greatly with the temperature of deformation. For slip takes place by avalanches occurring on the slip plane in a very brief space of time; the slip distance is limited by disordering which stops the avalanche. Since, as was shown above, the slip distance does not vary greatly with temperature, the hardening cannot vary greatly with temperature. That is, the curve in Fig. 9 is similar at all temperatures. At sufficiently high temperatures the hardening represented by this curve is modified, in the vicinity of the slip plane, by the self-annealing processes discussed above. In the region midway between slip bands, however, the hardness will be unaffected by self-annealing, since the stresses there are far below the critical value. It is the hardness in this region which determines the macroscopic hardness, for it must be assumed that slip will always take place on the softest plane.

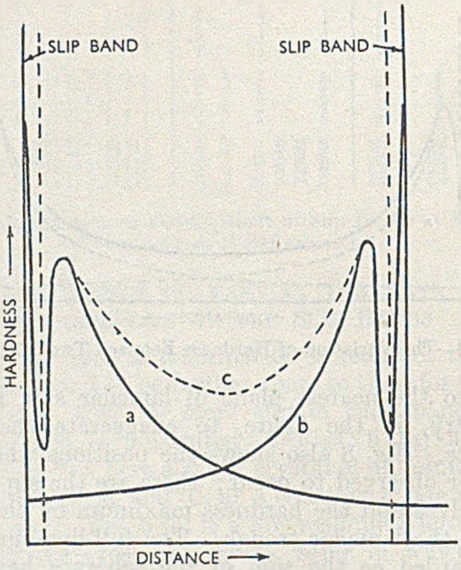

Fig. 12.—The Hardening Effect of a New Slip Band.

Curve (*a*): Hardness produced by slip band on left.
Curve (*b*): Hardness produced by slip band on right.
Curve (*c*): Combined effect of (*a*) and (*b*).

Similarly, since the slip distance is independent of strain, the effect of a slip process on the macroscopic hardness is also independent of strain.

2. The Relationship Between the Appearance of the Slip Bands and the Shape of the Stress/Strain Curve

The stress/strain curve is the macroscopic expression of the effect of the microscopic slip processes on the mechanical properties of the metal. Thus, if it is correct that the hardening effect of a slip process which forms a new slip band is much greater than that of a similar slip process occurring within an already existing band, then there ought to be a close correlation between the density of slip bands and the amount of hardening. Moreover, since the hardening effect of a slip process does not vary greatly with the temperature of deformation, it may be expected that the differences observed between stress/strain curves corresponding to different temperatures must be related to the observed differences in the way these slip processes are grouped into bands.

The initial parts of the plastic stress/strain curve follow on without change of slope from the elastic region, and so are found to be [7] of similar gradient at all temperatures (cf. Fig. 14). Correspondingly, the slip bands resulting from small strains are similar at all temperatures, and consist of a single slip process (Fig. 7). As the strain is increased, differences appear between stress/strain curves measured at different temperatures. At the same time, the corresponding distributions of slip bands begin to differ; slip on existing bands is more common at high temperatures than at low. The differences in disposition of slip processes at various temperatures are shown for various values of the strain in Fig. 7; at the strain corresponding to the lowest line, for example, (about 40%) the specimen deformed at − 180° C. has developed more slip bands than that deformed at room temperature and many more than that deformed at 250° C. Reference to published families of stress/strain curves [7] (cf. Fig. 14) shows that the hardening is greatest for the specimen deformed at − 180° C. and least for the one deformed at 250° C.

Thus there is a qualitative correlation between the density of slip bands and the amount of work-hardening. Closely spaced bands, with little total slip in each, are associated with great hardening, while widely spaced bands with many slip processes in each correspond to little hardening. A quantitative correlation between hardening and the density of slip bands is provided by the relation already quoted which Yamaguchi [1] obtained for aluminium deformed at room temperature :

$$\tau - \tau_0 = kN$$

The simplest interpretation of Yamaguchi's equation is that work-hardening is a property solely of the density of slip bands, not of the amount of slip in each. Thus the contribution to the macroscopic hardness is made by the first slip process in a band. Subsequent slip processes, occurring on planes within the band which have recovered, do not contribute at all. Thus when slip occurs within an existing band, the strain increases without any corresponding increase in stress.

This circumstance provides a qualitative explanation of the curvature of the plastic stress/strain curve; at all strains a given stress increase produces the same number of new slip bands, but as the strain increases the amount of slip on existing bands increases relative to the amount of slip forming new bands. Hence the strain produced by a given increase of stress increases progressively with strain. The curvature of the stress/strain curve is greater in proportion as the strain is produced by slip within existing bands rather than by formation of new bands. As shown above, this tendency increases as the temperature of deformation is increased, and consequently the curvature of the stress/strain curve becomes more pronounced with increasing temperature. In the next section it is shown how, on the basis of this interpretation of Yamaguchi's relation, an equation can be found for the stress/strain curve.

3. ANALYTICAL FORM OF THE STRESS DISTRIBUTION AROUND A SLIP BAND

If the hardness of a metal increases by the mechanism illustrated in Fig. 12, it is possible to obtain an equation for the empirical curves in the figure valid at places not too close to slip bands. Yamaguchi's result, that a doubling of the number of bands results in a doubling of the hardness, requires that the hardening due to a slip process falls off approximately reciprocally with distance from the slip plane. The hardening effect must be appreciable at distances of the order of the band spacing, i.e. several microns.

IV.—AN EQUATION FOR THE STRESS/STRAIN CURVE

Consider a metal crystal of height 1 mm. measured normal to the slip planes and apply to it a steadily increasing shear stress until the first slip band appears. The stress required is the critical shear stress, τ_0. The resulting strain is made up, partly of a strain, γ_0, corresponding to the first slip process in the band and partly of the strain corresponding to subsequent slip processes within the band. If the latter contribution to the strain is denoted by $\gamma_0 f_0$ then :

$$\gamma = \gamma_0 + \gamma_0 f_0.$$

At low temperatures $f_0 = 0$.

At this point in the deformation, the stress (τ_1) to produce a second slip band is given by Yamaguchi's relation :

$$\tau_1 - \tau_0 = k$$

The strain increase is again made up of two components : γ_0, corresponding to the slip process forming the second slip band, and $\gamma_0 f_1$, corresponding to subsequent slip processes which occur on the two slip bands at stress τ_1. Proceeding thus, by increasing the stress at each stage just sufficiently to produce one more slip band and observing the resulting strain at each stage, after N slip bands have been formed the stress is :

$$\tau_N - \tau_0 = Nk \quad . \quad . \quad . \quad . \quad (2)$$

while the strain is :

$$\gamma_N = \gamma_0\Big(N + \sum_{r=0}^{N} f_r\Big) \quad . \quad . \quad . \quad (3)$$

Equations (2) and (3) are the equations of the stress/strain curve. k is found by experiment and γ_0 is the shear strain corresponding to one slip process of distance s per mm. That is, if s is measured in Ångstrom units:

$$\gamma_0 = s \times 10^{-7} \quad . \quad . \quad . \quad . \quad (4)$$

and with $s = 2000$ Å.:

$$\gamma_0 = 2 \times 10^{-4}$$

The first term in equation (3) gives the contribution of the bands, while the Σ term gives the contribution of the second and subsequent slip processes in each band. The function f is a measure of the relative importance of the two slip mechanisms.

With sufficient accuracy, the sum in equation (3) can be replaced by an integral, which gives the following strain equation:

$$\gamma_N = \gamma_0\Big(N + \int_0^N f(x)dx\Big) \quad . \quad . \quad (5)$$

In the absence of theory, f is available to fit Yamaguchi's results to the equations; if $f(x)$ is taken to be βx^m, where β and m are constants at a given temperature, equation (4) is integrable, giving (substituting for γ_0):

$$\gamma_N = s \times 10^{-7}(N + \alpha N^{m+1}) \quad . \quad . \quad (6)$$

In Fig. 13 this curve is plotted with $s = 2000$ Å., $\alpha = 1{\cdot}6 \times 10^{-6}$, and $m = 2{\cdot}25$. Superposed on the curve are the values measured by Yamaguchi for five specimens: his results for the specimens he numbered iv, vi, and viii are plotted directly and fit equation (6) well. In the case of specimens v and vii his results do not fit the equation with $s = 2000$ Å., but fit it well if s is taken to be 850 Å. for specimen v and 1500 Å. for specimen vii. In order to show all five specimens fitting the same curve, Yamaguchi's measurements of strain for specimens v and vii have been multiplied by 2000/850 and 2000/1500 respectively, and the values thus obtained plotted in Fig. 13. Values of s used in fitting Fig. 13 are given in Table III; the range of s required is comparable with the limits found for the avalanche distance in aluminium of different purities. For comparison Table III also gives values of k recalculated from Yamaguchi's results.

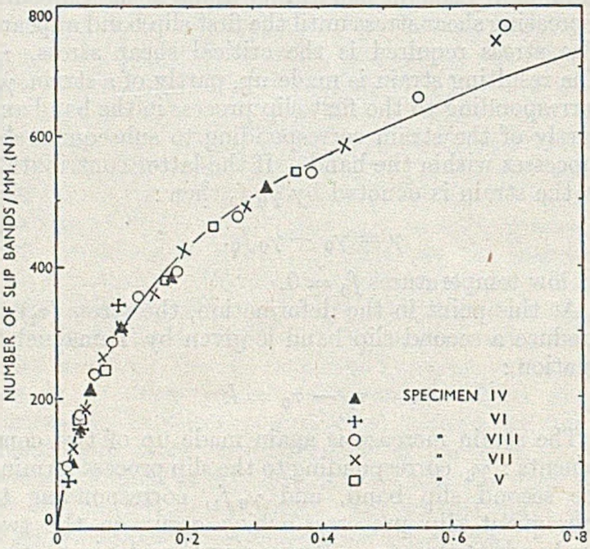

Fig. 13.—Comparison Between Yamaguchi's Results and Equation (6).

TABLE III.—*Comparison of Avalanche Distance* (s) *and the Rate of Hardening* (k).

Specimen No.	s, Å.	k, kg./mm.^{2}
iv, vi, viii	2000	$5{\cdot}0 \times 10^{-3}$
vii	1500	$4{\cdot}8 \times 10^{-3}$
v	850	$4{\cdot}0 \times 10^{-3}$

If the assumption that the differences between Yamaguchi's specimens lie in differences of unit slip distance is correct, then the table shows that the rate of hardening (k) as a function of N is not very dependent on the slip distance, but that shorter slip distances are apparently associated with slightly less hardening. If slip occurs by means of avalanches, this result implies that, as the avalanche passes over the slip plane it produces work-hardening; the avalanche is stopped, not when it has produced a definite amount of slip but when it has produced a fairly definite amount of hardening.

1. Stress/Strain Curves at Different Temperatures

Fig. 14 illustrates graphically the mechanism just discussed. The steeply sloping parts of the curve marked "20° C." show the hardening effect of the formation of successive slip bands, while the parts drawn horizontally show the effects of the addition of subsequent slip processes to these bands. Since the relative probability of slip within a band as opposed to the formation of new bands increases with strain, the horizontal portions become longer as the strain becomes greater. The result of this construction is the stepped curve indicated, which in the limit of small steps becomes the observed stress/strain curve shown dotted.[7]

Now it is known [4] that, for a given strain, there are more lamellæ in fewer zones at higher temperatures of deformation. Therefore, the horizontal portions of the curves corresponding to deformation at 250° and — 180° C. are to be drawn longer and shorter respectively than the corresponding lines for room-temperature deformation. The results are, again in the limit, the steeply rising curve of liquid-air deformation (— 180° C.) and the curve, at first steeply rising but rapidly becoming almost horizontal, of deformation at high temperatures (250° C.).[7]

Since, as was shown in Section III, the work-hardening in the absence of recovery is not very different at different temperatures, the sloping parts

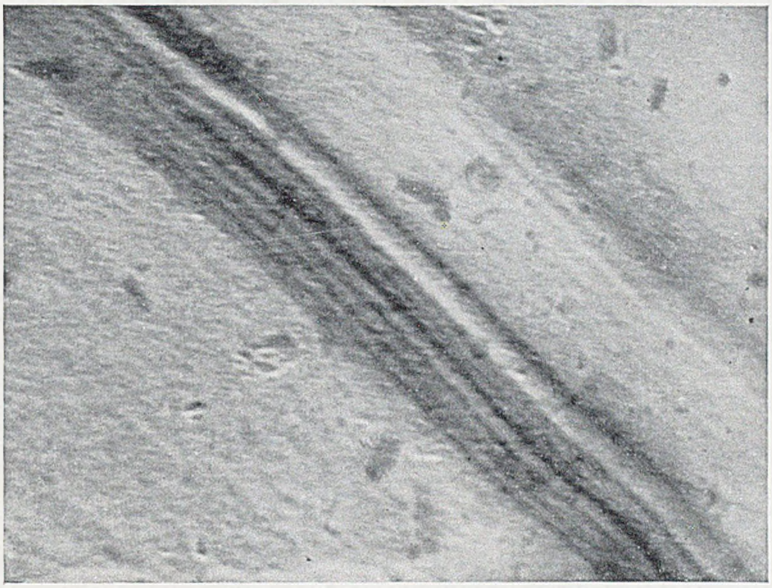

FIG. 4.—Slip Band on (100) Face of a Pure Aluminium Crystal. Approx. 70% shear strain. Aluminium oxide replica shadowed with gold–palladium. × 25,000.

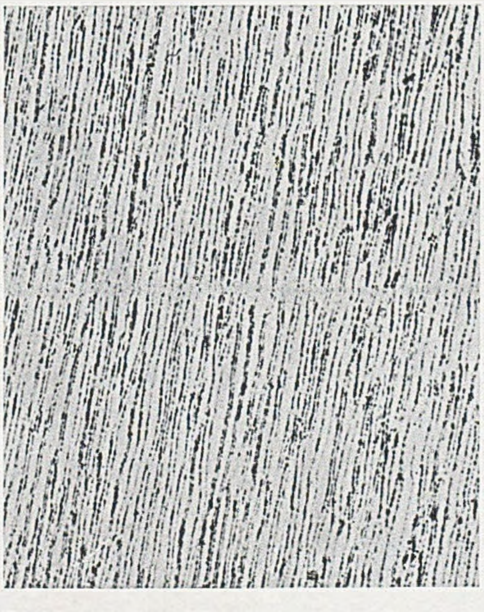

FIG. 5 (*a*).—Coarse Slip Bands on an Aluminium Single Crystal Extended at 450° C. at a Constant Strain Rate of 1% per Day. Total strain about 15%. × 6.

FIG. 5 (*b*).—Part of the Lamellar Structure of One of the Slip Bands Shown in Fig. 5 (*a*). Aluminium oxide replica shadowed with gold–palladium. × 25,000.

of the curves at all temperatures have been drawn of equal gradient; this implies that the differences between measured stress/strain curves at different temperatures are almost entirely due to the mechanism of self-annealing discussed above.

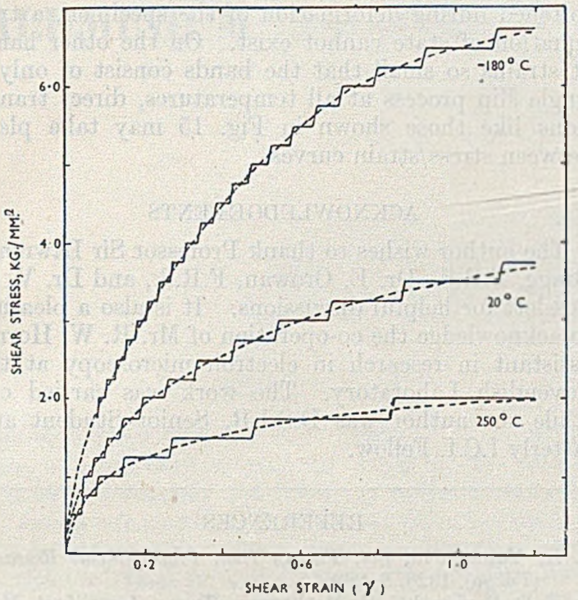

Fig. 14.—Stress/Strain Curves for Aluminium at Different Temperatures. Broken-line curves from Boas and Schmid.[7]

2. Derivation of the Function f from Ordinary Stress/Strain Curves

By constructing stepped stress/strain curves to fit measured curves, it is possible to obtain information about the form of the function f without counting slip bands. Thus in the stepped stress/strain curve, the form of which is given by equation (3), the sloping parts are described by the first term on the right and the horizontal parts by the second. As drawn in Fig. 14, each sloping part represents the formation, not of one new slip band, as was the case in the analysis above, but of some 50 bands. The analysis is unchanged, but γ_0 now represents the strain corresponding to that number of bands per mm. The length $l(r)$ of the r-th horizontal part is, from equation (3) :

$$l(r) = f(r)\,\gamma_0 \quad . \quad . \quad . \quad . \quad (7)$$

and if $f(r)$ is taken to be of the form βr^m as before, this equation can be written, taking logarithms :

$$\log l(r) = m \log r + \log(\beta\gamma_0) \quad . \quad . \quad . \quad (8)$$

Hence the slope of a plot of log r against log $l(r)$ gives m.

This method has been applied to the dotted curves of Fig. 14, which are stress/strain curves for aluminium obtained by Boas and Schmid.[7] The results obtained are shown in Table IV.

The agreement at 20° C. is rather lucky, since the curves obtained by Boas and Schmid are mean values (though with only small scatter) from some ten specimens. Moreover, there is no evidence that their specimens were of purity comparable with Yamaguchi's.

Table IV.—*Determinations of* m *by Equation* (8) *and by Yamaguchi.*

Temp., °C.	m (from equation (8))	m (from Yamaguchi)
−180	1·3	...
20	2·3	2·25
250	3·7	...

VI.—THE EQUATION OF STATE FOR PLASTIC DEFORMATION

The existence of a mechanical equation of state implies that work-hardening is a unique function of the strain (γ), the strain rate ($\dot{\gamma}$), and the temperature (T), i.e.

$$\tau - \tau_0 = f(\gamma, \dot{\gamma}, T)$$

Thus, a metal deformed along the room-temperature curve of Fig. 15 (*a*) to the point A and then cooled to liquid-air temperature would continue to deform from B along the liquid-air-temperature curve and vice versa. Similarly, if an equation of state exists, it should be possible to pass from a high-speed test to a slow-speed test and vice versa by a path such as AB (Fig. 15 (*b*)).

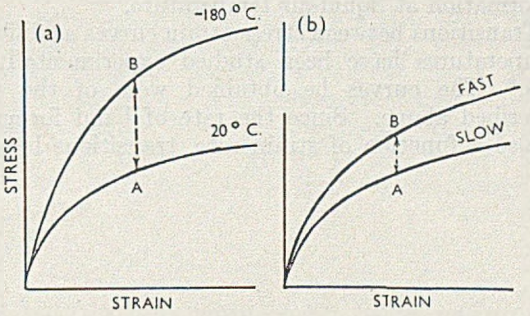

Fig. 15.—Transitions Between Stress/Strain Curves (*a*) at Different Temperatures, and (*b*) at Different Strain Rates, with Equation of State.

To see what, in fact, happens, consider two specimens both strained by the same amount, one at liquid-air temperature and the other at 20° C. The former has many more slip bands than the latter. If it is warmed up to 20° C., the stresses on the slip bands will be sufficient for self-annealing, giving rise to many more planes of easy slip than are available to the specimen deformed at 20° C. At the same time the stress to restart deformation will fall, say to the point B (Fig. 16 (*b*)). Then the metal, having a greater number of soft planes than correspond to its strain and temperature, deforms along an almost horizontal, i.e. non-hardening, curve until the surplus soft planes are used up. Thereafter its deformation follows the 20° C. curve.

Again, if the specimen deformed at 20° C. is cooled to liquid-air temperature, the stress to restart deformation is raised, say to the point *B* (Fig. 16 (*a*)). The metal has then many less slip bands than correspond to its strain and temperature. There will be available a few soft planes resulting from the last stages of deformation at 20° C., but when these are used up they will not be replaced. Since the slip bands are far apart, the easiest sites for slip are then the planes midway between bands, and it is on these that new slip processes will occur predominantly. The result of this extra rapid formation of new bands is the steeply rising stress/strain curve shown in Fig. 16 (*a*). This curve will be followed until it intersects the curve of deformation at liquid-air temperature.

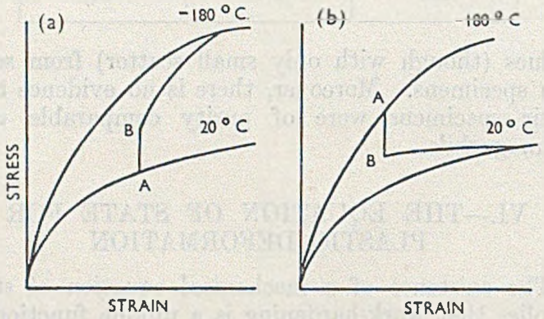

Fig. 16.—Transitions Between Stress/Strain Curves at Different Temperatures, with no Equation of State.
(*a*) Transition from 20° C. curve to −180° C. curve.
(*b*) Transition from −180° C. curve to 20° C. curve.

Transitions between stress/strain curves at different temperatures have been studied experimentally by Łoś.[8] The curves he obtained were of the form described above. Since the rate of band formation is also a function of strain rate, transitions between stress/strain curves corresponding to different rates of strain are of similar form. Thus, when a substantial proportion of plastic strain takes place by slip on existing bands, i.e. by slip on planes which have softened during deformation of the specimen, a true equation of state cannot exist. On the other hand, at strains so small that the bands consist of only a single slip process at all temperatures, direct transitions like those shown in Fig. 15 may take place between stress/strain curves.

ACKNOWLEDGEMENTS

The author wishes to thank Professor Sir Lawrence Bragg, F.R.S., Dr. E. Orowan, F.R.S., and Dr. V. E. Cosslett for helpful discussions. It is also a pleasure to acknowledge the co-operation of Mr. R. W. Horne, assistant in research in electron microscopy at the Cavendish Laboratory. The work was carried out while the author was D.S.I.R. Senior Student and latterly I.C.I. Fellow.

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EXPERIMENTS ON THE REACTION OF ALUMINIUM–MAGNESIUM ALLOYS WITH STEAM *

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

SYNOPSIS

The reaction of aluminium–magnesium alloys containing up to 25% magnesium with pure steam has been investigated over a range of temperatures from 450° to 700° C. A maximum reactivity, dependent on the composition, was found at temperatures between 550° and 625° C. The significance of these results in relation to metal/mould reaction in the aluminium–10% magnesium alloy is discussed briefly.

I.—INTRODUCTION

Calculations indicate that when a metal is cast into a green-sand mould the water in the sand is almost immediately volatilized in sufficient quantity to sweep out practically all the air from the pores in the mould near the metal surface, so that the casting in effect solidifies in an atmosphere of practically 100% steam. The time for which the metal remains above the solidus temperature is likely to be of the order of 10 min. for a casting about 2 in. thick in an aluminium alloy of wide freezing range, and correspondingly longer for a larger casting.[1] When the metal contains a readily oxidized element, the latter may react with the steam during solidification to form oxide and hydrogen. This reaction, commonly known as metal/mould reaction, is prevalent in sand castings of aluminium–10% magnesium alloy, and the work described in the present paper is part of an investigation on metal/mould reaction in this alloy.

It is convenient to distinguish between "internal" hydrogen, which is absorbed by the metal and may give rise to porosity or disruption, and "external" hydrogen, which is released to the atmosphere. For an aluminium–7% magnesium [2] alloy reacting with water vapour at 16 mm. pressure at approximately 600° C., between the solidus and liquidus temperatures, the external hydrogen was much the greater, representing about 98% of the total hydrogen formed in the reaction. Thus the external hydrogen, although not so important in some respects as the internal hydrogen, is a convenient approximate measure of the total extent of the reaction. In these experiments, therefore, various alloys have been allowed to react with pure steam at temperatures between 450° and 700° C., and the external hydrogen collected and measured during the first 15 min.

Alloys with 1–25% magnesium have been studied. Although alloys containing more than 10% magnesium are not normally used in casting, they have been included to investigate the effect of magnesium enrichment of the liquid phase in partially molten alloys. The use of small additions of beryllium to inhibit metal/mould reaction has been suggested, and preliminary experiments on such alloys have been carried out.

II.—EXPERIMENTAL WORK

1. Method

The apparatus used is shown diagrammatically in Fig. 1. For some experiments with the alloys of higher magnesium content, a subsidiary, uncooled, collecting system of 900 c.c. capacity was used. Specimens used in the solid or partially molten state were cylinders 1 cm. dia. and 1 cm. long, with an axial hole of 1·5 mm. dia. drilled for half the length, so that the hot junction of a thermocouple could be placed at their centres. They were machined dry to size and degreased in acetone. Specimens used in the molten state were shaped to fit the boat by filing and were arranged so that the hot junction of the thermocouple was immersed when the specimen melted. The boats used with completely molten specimens were normally of such a size that the exposed surface area was approximately equal to that of the partially molten specimens. For alloys containing more than 20% magnesium the oxidized nickel boat, shown in Fig. 1, was replaced by one of fused alumina.

To ensure rapid response to temperature changes in the specimen, the hot junction of the thermocouple (28-gauge Chromel/Alumel) was insulated only by a thin coat of an aluminous cement. In operation, the apparatus was set up with the specimen *in situ*, and the reaction tube was swept out with purified argon, which was not collected. The temperature was then raised and when it was steady at the desired value the argon flow was stopped and steam substituted by closing clip 1 and opening clip 2 (Fig. 1). Approximately 20 sec. after admitting steam the collection

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of gas in the burette was started, and readings taken at 1-min. intervals up to 15 min. The principal difficulty was in the control of temperature, since much heat is evolved in the reaction. By hand regulation of the furnace current, the temperature of the specimen was maintained constant to ± 8° C. for alloys with less than 15% magnesium, and to ± 15° C. for alloys of higher magnesium content.

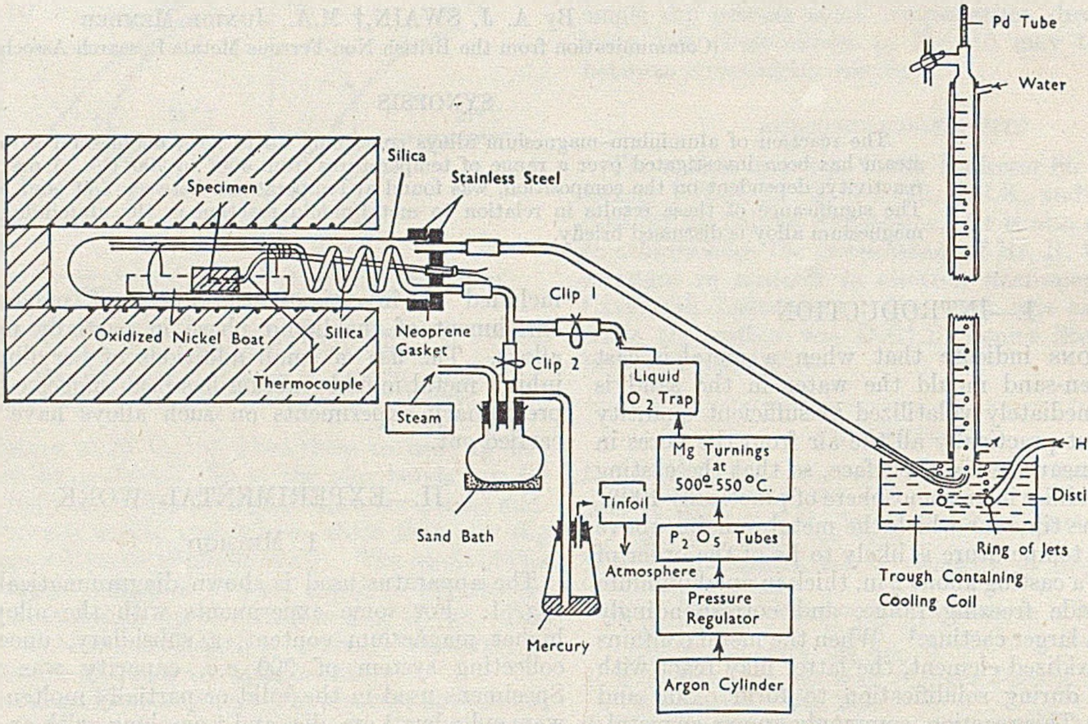

FIG. 1.—Diagram of Apparatus. A stainless steel flange is attached to the reaction tube, and various tubes are led through the end-plate, by compression joints with Neoprene washers.

Preliminary experiments indicated that :

(*a*) Little or no oxidation of the specimen occurred during the heating up.

(*b*) No significant quantity of argon was collected in the burette.

(*c*) The blank, probably owing to the evolution of dissolved gases in the water in the collecting system, was substantially constant at 0·6(7) c.c. at 15° C. in 15 min. and was proportional to the time.

The palladium tube on the burette enabled a check to be made on the identity of the evolved gas. The percentage of gases other than hydrogen varied from about 2% for the largest quantities of hydrogen recorded in Table I to something approaching 10% for the smallest. It seems likely that this gas originated in the water in the burette. In the figures given in the table these gases have been included in the hydrogen figure. The total hydrogen collected during the time at temperature and subsequent cooling was, in many experiments, checked against increase in weight of the specimen, and the oxidation equivalent to the hydrogen collected generally accounted for about 90% of the weight change.

The time lag between the evolution of hydrogen and its appearance in the burette was presumably less than 15 sec., since this time was sufficient to displace completely the argon initially present.

2. MATERIALS

The alloys were made from aluminium and magnesium both of 99·99% purity. The melts were made in salamander crucibles and degassed with chlorine before casting. The alloys containing up to 18·25% magnesium were sand cast (2% boric acid in sand) in the form of D.T.D. bars (6 in. long and 1 in. dia. with a feeder head 2¼ in. long by 2½ in. dia.) ; those of higher magnesium content were cast in a 1-in.-dia. iron chill mould. The bars were cut into four sections longitudinally, and the specimens were machined or filed from these sections.

The compositions given in Table I were obtained by chemical analysis.

III.—RESULTS AND DISCUSSION

Figs. 2, 3, and 4 show the progress of the reaction for a number of the experiments. No allowance has been made for the blank.

Table I gives the quantities of hydrogen collected in 15 min., with due allowance for the blank, expressed as c.c. at N.T.P./100 g., and also as c.c. at N.T.P./unit surface area. The latter values have been calculated on the original surface area. In certain heavily oxidized specimens considerable exudation had occurred.

The table also gives, where it is relevant, the figure obtained by dividing the volume at N.T.P./unit area by the fraction of the alloy molten. This gives some basis for comparison between different alloy compositions at the same temperature. The maxi-

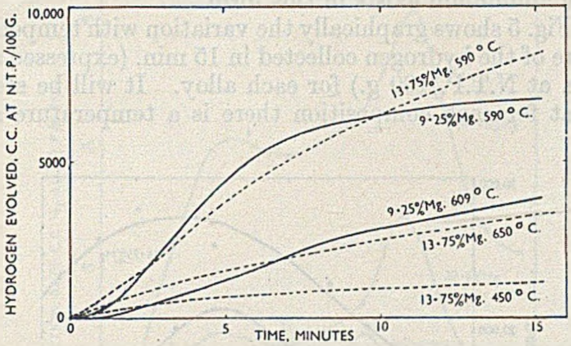

Fig. 2.—Progress of the Reaction for 9·25% and 13·75% Magnesium Alloys.

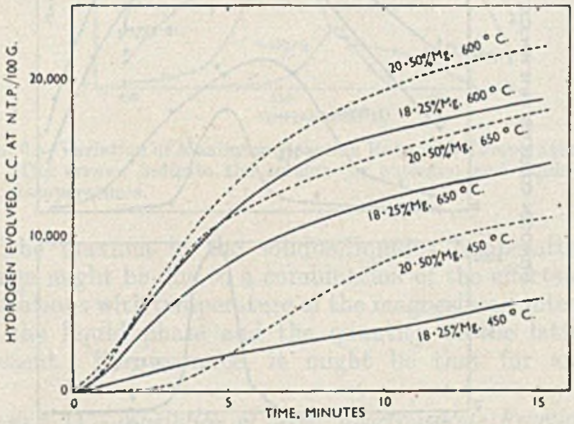

Fig. 3.—Progress of the Reaction for 18·25% and 20·50% Magnesium Alloys.

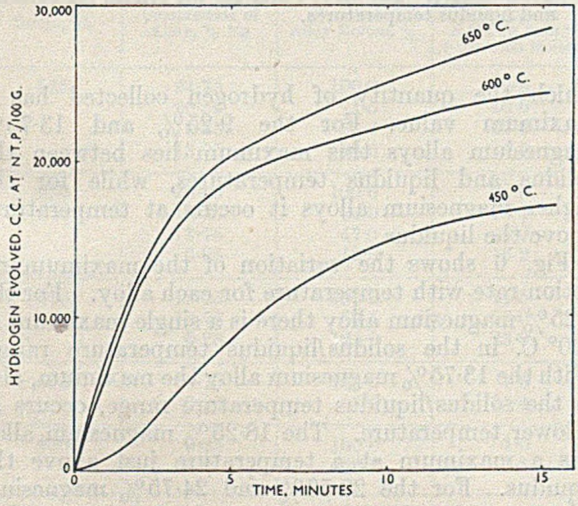

Fig. 4.—Progress of the Reaction for 24·75% Magnesium Alloy.

Table I.—*Summary of Experimental Results.*

Compn., % Mg	Temp., °C.	Fraction of Alloy Molten,* %	Compn. of Liquid, % Mg	Hydrogen Collected in 15 min.			Max. Re-action Rate, c.c. at N.T.P./g./min.
				c.c. at N.T.P./100 g.	c.c. at N.T.P./cm.²	c.c. at N.T.P./cm.² ÷ fraction molten	
0·75	652	50	2·0	93	0·3	0·6	...
				77	0·3	0·6	...
	675	100	...	12	0·2	...	...
				4	0·1	...	...
4·80	550	0	...	0	0	...	...
				0	0	...	...
	570	0	...	143	0·5	...	...
	590	0	...	1,150	3·2	...	1·5
				910	2·4	...	1·1
	609	12	11·2	830	3·2	26·6	0·8 (6)
				870	3·4	28·3	0·8 (6)
	613	18	10·5	620	2·4	13·3	0·7 (0)
				410	1·6	8·9	0·6 (0)
	628	46	7·5	70	0·3	0·65	...
				40	0·2	0·45	...
	650	100	...	5	0·1	...	...
				9	0·1	...	...
	675	100	...	21	0·3	...	...
				28	0·4	...	...
9·25	450	0	...	65	0·25	...	...
				128	0·37	...	...
	500	0	...	179	0·7	...	...
				153	0·6	...	...
	560	16	20	1,100	4·5	28·0	2·5
				1,300	4·7	29·4	2·9
	590	42·5	15·0	7,020	23·8	56·0	11·1
				6,430	22·8	53·3	9·0
	609	73	11·2	3,840	13·2	18·1	3·6
				4,000	15·8	21·6	3·1
	625	100	...	530	7·2	...	0·4 (6)
				460	5·6	...	0·4 (6)
	650	100	...	180	2·5	...	...
				172	2·7	...	...
	675	100	...	60	1·0	...	...
				40	0·7	...	...
	700	100	...	27	0·05	...	...
				22	0·3	...	...
13·75	450	0	...	810	3·07	...	0·8
				1,105	4·07	...	1·4
	500	11·5	28·8	2,410	8·68	75·3	3·7
				2,425	8·13	70·8	2·6
	545 †	38·0	22·2	11,800	41·2	108·5	23·4
				9,100	26·3	69·2	21·8
	590	90	15·0	8,370	65·6	72·8	7·3
				9,520	94·7	105·0	8·7
	625	100	...	6,390	63·3	...	6·2
				6,790	73·2	...	6·8
	650 †	100	...	3,400	60·0	...	...
				3,230	32·0	...	4·4
	675 †	100	...	1,200	19·0	...	...
				770	7·8	...	0·5
	700	100	...	224	2·20	...	0·36
				231	2·20	...	0·37
18·25	450	14·0	35·0	4,820	16·9	121·0	3·1
				5,840	18·2	130·0	5·5
	500	37·5	28·8	11,400	33·5	89·2	17·6
				10,100	29·4	78·2	14·6
	560	86·5	20	17,600	219·0	242·0	17·6
				17,400	220·0	243·0	21·3
	600	100	...	17,890	136·2	...	26·5
				18,750	152·0	...	23·4
	625 †	100	...	15,750	76·5	...	23·2
				16,150	141·0	...	22·7
	650	100	...	14,450	170·0	...	21·7
				15,250	170·5	...	...
	675	100	...	9,790	101·0	...	11·8
				10,250	177·5	...	13·3
	700	100	...	4,500	53·2	...	4·7
				4,210	61·8	...	4·8
20·50	450	26·5	35·0	9,880	31·2	117·5	8·0
				10,950	30·0	113·0	13·7
	500	52·0	28·8	13,820	43·2	82·3	25·2
				14,780	47·5	91·4	34·3
	550 †	93·0	21·4	18,500	119·6	128·0	33·0
				19,550	219·0	235·0	22·5
	600	100	...	22,050	162·5	...	30·5
				21,000	162·0	...	38·5
	625	100	...	20,400	119·0	...	45·0
				20,000	111·5	...	52·5
	650	100	...	17,200	91·5	...	20·7
				17,800	75·5	...	40·0
	675	100	...	13,850	76·7	...	23·5
				13,450	58·7	...	21·8
	700	100	...	7,730	75·7	...	11·4
				6,790	30·5	...	10·8

Continued.

TABLE I.—*continued.*

Compn., % Mg	Temp., °C.	Fraction of Alloy Molten,* %	Compn. of Liquid, % Mg	Hydrogen Collected in 15 min.			Max. Re-action Rate, c.c. at N.T.P./g./min.
				c.c. at N.T.P./100 g.	c.c. at N.T.P./cm.^{2}	c.c. at N.T.P./cm.^{2} ÷ fraction molten	
24·75	450	48·0	35·0	16,250	43·6	90·9	16·0
				16,950	41·5	86·3	21·8
	500	76·0	28·8	20,100	34·8	133·0	50·0
				20,250	28·3	98·2	68·6
	550	100	...	22,350	171·5	...	51·8
				27,000	133·5	...	64·5
	600	100	...	27,300	121·0	...	70·3
				24,100	100·5	...	61·0
	625	100	...	26,450	150·0	...	80·0
				25,850	144·0	...	84·0
	650	100	...	28,250	182·5	...	72·0
				27,950	133·0	...	60·7
	675	100	...	27,000	175·5	...	44·5
				24,850	163·0	...	50·0
	700	100	...	25,700	129·0	...	48·6
				21,200	119·5	...	37·0

* Calculated from Raynor's diagram.[5]

† The specimens used at this temperature had different volumes per unit surface area.

mum reaction rate obtained from the slope of the hydrogen-collection curves, expressed as c.c. at N.T.P./g./min., is also given.

The appearance of the specimens after reaction varied according to the extent of oxidation. Lightly oxidized specimens of low magnesium content retained their metallic appearance; others had a grey powdery surface layer of varying thickness, in some cases comprising the whole specimen. The grey powder was found by X-ray diffraction to consist of magnesium oxide and practically pure aluminium.

The weight per unit surface area for specimens which were heavily oxidized in the liquid state was deliberately varied, and good reproducibility was obtained for the figures for hydrogen collected expressed as c.c. at N.T.P./100 g., but not for the volume of hydrogen per unit surface area. This suggests that for vigorous reactions in the liquid state the oxidation penetrates rapidly into the specimen, so that the reaction is proceeding simultaneously at a number of centres, and consequently the reaction rate is independent of the original surface area with specimens of the sizes used. In fact, it has been observed that, a short time after the start of the reaction, the hydrogen boils off from the interior of the melt. The surface is thus disrupted and a new surface is constantly formed. Presumably this process results in the formation of channels throughout the metal by which further oxidation can proceed.

In some experiments the quantity of hydrogen collected exceeded that expected from complete oxidation of the magnesium present. This discrepancy can be explained by considering the hydrogen resulting from the surface oxidation of the particles of aluminium in the oxidation product, if it is assumed that they are fine and dispersed. For example, in one experiment with the 24·75% magnesium alloy, 28,250 c.c. hydrogen at N.T.P./100 g. were collected, whereas complete oxidation of the magnesium in this alloy would give 22,800 c.c. The excess hydrogen obtained is equivalent to the formation of an alumina film 100 Å. thick on particles of aluminium approximately 2 microns in dia., assuming that the whole of the aluminium exists in this form.

Fig. 5 shows graphically the variation with temperature of the hydrogen collected in 15 min. (expressed as c.c. at N.T.P./100 g.) for each alloy. It will be seen that for each composition there is a temperature at which the quantity of hydrogen collected has a maximum value. For the 9·25% and 13·75% magnesium alloys this maximum lies between the solidus and liquidus temperatures, while for the higher magnesium alloys it occurs at temperatures above the liquidus.

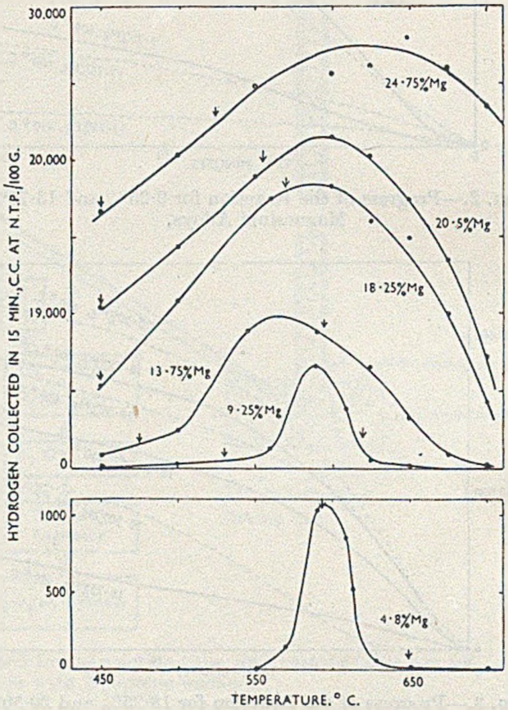

FIG. 5.—Variation of Hydrogen Collected in 15 Min. with Temperature. The arrows indicate the solidus (or eutectic) and liquidus temperatures.

Fig. 6 shows the variation of the maximum reaction rate with temperature for each alloy. For the 9·25% magnesium alloy there is a single maximum at 590° C. in the solidus/liquidus temperature range. With the 13·75% magnesium alloy the maximum, also in the solidus/liquidus temperature range, occurs at a lower temperature. The 18·25% magnesium alloy has a maximum at a temperature just above the liquidus. For the 20·50% and 24·75% magnesium alloys two maxima occur in each case, one within the solidus/liquidus temperature range and the second at approximately 625° C.

The solidus and liquidus temperatures indicated in Figs. 5 and 6 are the equilibrium values. In these experiments, however, the alloys may not have been in equilibrium, and consequently the solidus temperatures quoted do not necessarily apply.

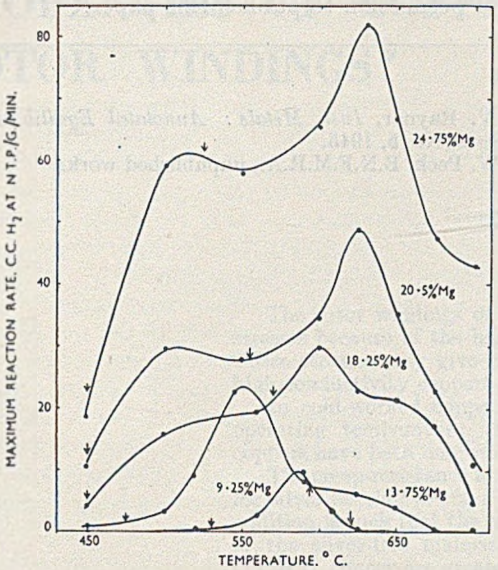

FIG. 6.—Variation of Maximum Reaction Rate with Temperature. The arrows indicate the solidus (or eutectic) and liquidus temperatures.

The maxima in the solidus/liquidus temperature range might be due to a combination of the effects of variations with temperature of the magnesium content of the liquid phase and the quantity of the latter present. For instance, it might be that for any particular liquid composition the initial reaction rate is equal to the rate for a completely liquid specimen multiplied by the fraction of the surface which is molten, and that this fraction is equal to the volume fraction of molten metal. Table II shows, for a number of temperatures, the maximum reaction rate divided by the fraction molten for alloys which are partially liquid at that temperature and have the same magnesium content in the liquid phase (assuming equilibrium between solid and liquid phases). The figures for each temperature vary considerably, and the assumptions made are, therefore, not correct except to a very rough approximation; the lack of agreement is probably due to the true reaction surface at the time of maximum reaction rate being different from the initial surface of the molten metal. However, it is clear from the figures for 100% liquid that the reactivity of the liquid phase in equilibrium with the solid at the temperature of test, does in fact increase as the temperature falls, owing to its increased magnesium content, in agreement with the general explanation of the maximum advanced above.

TABLE II.—*Variation of Max. Reaction Rate/Fraction Molten with Temperature and Composition.*

Temp., °C.	Composition of Alloy, % Mg	Fraction of Alloy Molten, %	Max. Reaction Rate (c.c. at N.T.P./g./min.) ÷ Fraction Molten
525	24·75	100·0	60·5
	20·50	71·5	40·6
	18·25	55·5	49·0
	13·75	25·5	32·5
555	20·50	100·0	28·0
	18·25	80·0	23·4
	13·75	47·0	48·9
	9·25	12·5	16·2
570	18·25	100·0	21·5
	13·75	62·5	28·0
	9·25	24·0	23·0
595	13·75	100·0	7·5
	9·25	48·5	15·4
620	9·25	100·0	1·0
	4·8	26·5	1·1

The occurrence of maxima in the curves of Fig. 6 at temperatures above the liquidus is very surprising, and no simple explanation can be advanced other than a change in the nature of the (rate-controlling) oxide films formed at these temperatures.

It has been found [4] that porosity caused by metal/mould reaction in the 10% magnesium alloy is confined to areas near the reacting surface. From the present results it can be seen that a large proportion of the hydrogen entering a casting of this alloy as a result of metal/mould reaction will do so at a comparatively late stage in the solidification process, namely, while the casting is partially solid, and thus would not be expected to diffuse far into the metal before being rejected from solution.

A few preliminary experiments have been carried out with a 10% magnesium alloy containing 0·025% beryllium, at temperatures between the solidus and liquidus. This alloy was made up from a 5% aluminium-beryllium hardener prepared from super-purity aluminium and commercially pure beryllium.

In all cases the quantity of hydrogen collected was of the same order as the blank value. The specimens after oxidation had a thin oxide film, which showed a bluish-green interference colour, similar to that found on inhibited castings of this alloy. Beryllium additions of 0·025% with the same hardener had been found to inhibit effectively metal/mould reaction in a 2-in.-dia. block casting of 10% magnesium alloy, so that in this instance there is agreement between the reaction rate and foundry experiments.

IV.—CONCLUSIONS

The reaction between aluminium–magnesium alloys and steam, as measured by the external hydrogen formed under the conditions of these experiments, reaches a maximum rate at a temperature dependent on the composition of the alloy, but within the range 550°–625° C. For the 10% magnesium alloy, which is widely used as a casting alloy, this maximum reaction rate occurs at a temperature within the solidus/liquidus range. This probably accounts for the fact that gas porosity caused by metal/mould reaction in sand castings of this alloy occurs mainly in the surface layers of the castings.

No satisfactory explanation of these effects can be advanced at present.

ACKNOWLEDGEMENT

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THE CREEP AND SOFTENING PROPERTIES OF COPPER FOR ALTERNATOR ROTOR WINDINGS*

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

SYNOPSIS

The rotor windings of some large turbo-alternators are subject to longitudinal compressive stresses because of the higher coefficient of expansion of the copper relative to the steel rotor. These stresses may give rise to creep deformation and the resistance to creep of a number of high-conductivity coppers has been investigated in the temperature range 130°–225° C.

In cold-worked coppers resistance to creep may be reduced by softening occurring at the operating temperature. In consequence, the softening characteristics of high-conductivity coppers have been determined over the same temperature range.

The creep-resistance and the resistance to softening have been shown to be very much greater for silver-bearing (0·1% silver) than for silver-free, tough-pitch copper. The effect of the silver addition is such that the creep-resistance of the silver-bearing material at 225° C. is equal to that of the silver-free material at 130° C. The creep-resistance and the resistance to softening of O.F.H.C. copper are greater than those of tough-pitch copper when both are silver-free. Silver-bearing (0·1%) O.F.H.C. and tough-pitch coppers have very similar resistances to creep and softening.

In silver-bearing tough-pitch copper the resistance to creep increases with increase in the degree of cold work up to 10%. Beyond this amount there is no further increase in creep-resistance, and it is clear that maximum resistance to creep coupled with maximum resistance to softening is obtained with cold work of this degree.

I.—INTRODUCTION

The deformation, referred to as copper shortening, occurring in the rotor windings of some large turbo-alternators became a serious problem a few years before the Second World War. Since then, the problem has become more serious owing to the heavy power demands of war-time and post-war production and has been further aggravated by the shift working of alternators.

The generally accepted explanation of copper shortening in alternator rotor windings is given very briefly below. The subject has been discussed very thoroughly by a number of English and American authors.[1–5]

According to usual practice, an alternator is started up with the windings cold, and the excitation current is applied when the machine is running at speed. The effect of the centrifugal force on the windings in the rotor slot is to press them outwards against the retaining wedge and hence increase considerably, by reason of the frictional forces, the resistance to lengthwise movement. The temperature rise in the copper winding produced by the field-excitation current is greater than that in the surrounding parts of the steel rotor, which are heated by the copper. For this reason, and also because the coefficient of expansion of copper is approximately 50% greater than that of steel, the copper will tend to expand lengthwise relative to the steel. This relative expansion is resisted by the frictional forces and as a consequence, the copper is subjected to compressive forces in the direction of its length. If the compressive stress in the copper due to these forces exceeds a certain value, plastic deformation and creep in compression will occur. This creep will be in the nature of relaxation, since, as it occurs, it will relieve the compressive stress and will progressively decrease as the stress is reduced.

The average temperature of the copper under normal running conditions would not exceed 130° C., in accordance with British Standard Specification No. 225, although some parts might approach 160° C., and under exceptional circumstances might even reach 190° C.[6]

The steel of the rotor may, according to Juhlin,[2] be as much as 50° C. lower in temperature. Assuming the copper at a temperature of 130° C., an analysis, based on elastic deformation only, shows that a stress

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as high as 17,500 lb./in.^{2} may be induced in the copper. More recently, Horsley and Coates[5] have expressed the view that temperature differences between the conductors constitute the major factor determining the stresses built up in the copper, but have shown that stresses of the order quoted above may still be approached as the copper progressively work-hardens with each cycle of operation.

When the alternator is stopped, the exciting current, being switched off, will no longer heat the copper on the rotor, and contraction occurs, now with considerably reduced frictional restraint in the absence of centrifugal forces. Thermal contraction of the copper will be greater than that of the steel, and also the copper will have been shortened by any creep which occurred during operation. If on the one hand the ends of the windings are not held rigidly against this contraction, the end coils are pulled towards the rotor, leading in extreme cases to earthing faults after a number of repetitions of heating and cooling cycles. On the other hand, if the end coils are rigidly held against this contraction, tensile stresses will be set up in the copper.

Two different ways of solving this problem have been tried:

(1) By not running the alternator up to speed—and hence not bringing the centrifugal forces into full operation—until the copper windings have been raised to the operating temperature.

(2) By the use of rigid fixing of end coils against contraction. As already mentioned, this leads to tensile stresses in the copper when the rotor cools down, which has the effect of reducing the value of the compressive stress in the copper under full alternator load. This method has been combined with the use of copper having a high resistance to softening at the operating temperature.

The first method is only a palliative and not a complete solution; the second aims at ultimately limiting the compressive stresses to values at which plastic deformation and creep do not occur.

The failure of some earlier attempts to use hard-drawn coppers has been attributed to reduction in mechanical properties brought about by softening at the operating temperatures. It is known that the addition of silver to copper increases its resistance to softening, and hence it appears that the use of silver-bearing copper will reduce, if not eliminate, copper shortening. Clearly, if the copper were capable of withstanding, without plastic deformation, the compressive stresses imposed at the operating temperatures, no shortening would occur and the necessity for rigid end-coil fixing or preheating would not arise.

In view of the possible use of silver-bearing copper, data on the mechanical properties, particularly creep properties, of this material were required for comparison with those of ordinary tough-pitch copper. A programme of investigation was undertaken to determine the effect of elevated temperatures not exceeding 225° C. on these properties in high-conductivity tough-pitch copper, both with and without 0·1% silver. It is considered that, according to the copper–silver equilibrium diagram,[7] this amount of silver was virtually all in solid solution at the test-temperatures. Additionally, these properties were determined for silver-free and silver-bearing O.F.H.C. copper, since some recent American work[8] had suggested that these coppers have higher resistance to creep and to softening than the tough-pitch coppers.

Much of the present work is devoted to stress-free softening tests and creep testing in tension, and it is therefore appropriate at this point to explain the basic reasons underlying the decision to investigate, in particular, these physical properties. As already mentioned, the compressive stresses built up in the alternator rotor windings arise from the specific strain imposed on the copper by thermal expansion. Such stresses may be relieved by creep in compression, the copper increasing its cross-section. It might be thought, therefore, that a compression relaxation test would be the most suitable guide to probable service behaviour. If the copper shortening is to be entirely prevented, relaxation—and hence creep—must not occur. Creep tests in compression are difficult to carry out, and since very short specimens must be used, a high degree of sensitivity of strain measurement cannot be attained. The assumption is generally made that in most materials the mechanical properties in compression are similar to those in tension. This consideration, coupled with the fact that tensile creep-testing equipment was available, led to the use of tensile creep tests as a basis of comparison of the various coppers which have been used or are proposed for use in turbo-alternator windings.

The time to softening of cold-worked copper is known to depend on the degree of cold work and the temperature, and it was clearly important to establish how these variables are inter-related in the coppers used in turbo-alternators. Softening tests were therefore included in the investigation. The absence of softening in such tests does not necessarily mean that the materials will not soften under the same conditions of heat-treatment when simultaneously subjected to stress. Materials which soften in the absence of stress will also do so when stressed, and as the softening test is simple and does not require elaborate apparatus, it affords an easy method of sorting out those materials which will certainly be unsatisfactory under stress. Cook and Richards[9] have recently shown that the softening of cold-worked coppers at relatively low temperatures can be predicted with fair quantitative agreement on the basis of short-time tests at higher temperatures. Hence the softening test afforded a rapid method of assessing the behaviour of the materials as affected by low-temperature heat-treatment.

Although considerable importance is attached to the results of the softening tests, it is realized that the final assessment of suitability of materials must be in terms of creep behaviour, and creep testing therefore constituted the major part of the work.

Table I.—*Analyses of Tough-Pitch and O.F.H.C. Coppers.*

Type of Copper	O_2, %	Ag, %	Fe, %	Pb, %	Ni, %	Sb, %	S, %	Remarks
Tough-pitch	0·03 *	0·002	0·0002	n.d.	0·0003	n.d.	0·002 *	Zn not detected.
Tough-pitch + silver	0·02 *	0·086 *	0·0002	n.d.	0·0001	n.d.	0·001 *	
O.F.H.C.	0·0002 †	0·001	0·0002	0·0003	0·0003	0·0008	0·002 *	As, Bi, Co, P, Si, Mn, Zn, Te, and Sn not detected.
O.F.H.C. + silver	0·0002– 0·0003 †	0·072 *	0·0002	0·0003	0·0003	0·0008	0·002 *	

* Chemical analysis.

† Oxygen content measured by the method described by Baker.[10] Superficial oxidation of the sample between cleaning and insertion in the apparatus may account for about 0·0001% oxygen.

n.d. = not detected. Limits of detection are as follows: Zn, 0·001%; As, 0·002%; Bi, 0·0001%; Co, 0·0001%; P, 0·01%; Si, 0·005%; Mn, 0·001%; Te, 0·004%; Sn, 0·0005%; Sb, 0·0002%; Pb, 0·0003%.

II.—MATERIALS

Primarily, two materials were selected for investigation: silver-free and silver-bearing (nominal 0·1% silver) tough-pitch, high-conductivity coppers. These were prepared by a member firm of the Association from O.R.C.-brand, vertically cast ingots. At a later stage silver-free and silver-bearing O.F.H.C. coppers were also obtained.

The materials were obtained in straight lengths in the form of strip 1¼ × 0·1 in. cross-section, in the annealed condition and also with 10, 25, and 50% cold reduction of area.

All test materials were produced by a combination of hot rolling, annealing, and drawing, the final cold reduction of area being effected by drawing, except in the case of the coppers with 50% reduction. This final temper was produced by drawing (32% reduction) followed by rolling. In the production of copper strip to a given degree of cold work, the same programme was applied to all the coppers.

1. Composition

All materials were analysed spectrographically for minor constituents and chemically for main constituents. Results of these analyses are reported in Table I.

2. Uniformity of Cold Working of the Coppers

The hardness was virtually constant across the width of the strip in the 50% cold-worked materials. In those with 25% cold work the variation from edge to centre was about 5%. The 10% cold-worked materials showed the greatest variation in hardness. Thus, in the tough-pitch coppers, the Vickers hardness ranged from 80 at the centre of a transverse section to 98 on the surface near the edge. In the O.F.H.C. coppers, corresponding figures were 91 to 97 in the silver-free and 81 to 94 in the silver-bearing strips. However, in all the materials the variation in hardness over the centre ½-in. width, which formed the test width of a creep specimen, was negligible.

Hardness tests along the length of the strips indicated that uniformity of cold working in this direction was good in all the materials.

3. Comparison of Hardness and Grain-Size

From Table II, it will be seen that the grain diameter of the O.F.H.C. copper is of the same order as that of the tough-pitch and that the tough-pitch is slightly harder than the O.F.H.C., especially in the annealed condition. In the materials with 50% cold work it was not possible to determine the grain diameter because of the elongated shape of the grains.

Table II.—*Vickers Pyramid Hardness and Grain-Size of O.F.H.C. and Tough-Pitch Coppers.*

Cold Work, %	V.P.N. (Mean of Five Impressions, 20 kg. Load)			
	Tough-Pitch	O.F.H.C.	Tough-Pitch + Silver	O.F.H.C. + Silver
0	48·0	36·8	44·9	37·8
10	94·1	91·0	96·1	93·4
25	105·0	101·3	105·0	102·0
50	114·5	111·0	116·0	115·0

Cold Work, %	Grain-Dia., mm.			
	Tough-Pitch	O.F.H.C.	Tough-Pitch + Silver	O.F.H.C. + Silver
0	0·030	0·040	0·030	0·040
10	0·035	0·040	0·030	0·040
25	0·030	0·035	0·035	0·035
50	...	...	...	...

4. Room-Temperature Tensile Properties

Results of tensile tests on all the coppers used are shown in Table III. It will be seen that there is little difference between the various coppers at any given degree of cold working.

5. Electrical Conductivities at Room and Elevated Temperatures

In any changeover from high-conductivity copper to a material having improved properties, the alternator designer requires to know the extent to which

Table III.—*Comparative Room-Temperature Tensile Properties of Silver-Free and Silver-Bearing O.F.H.C. and Tough-Pitch Coppers.*

Each figure is the mean of two tests.

Cold Work, %	Property	Tough-Pitch	O.F.H.C.	Tough-Pitch + Silver	O.F.H.C. + Silver
0	0·1% P.S., tons/in.^{2}	2·3	3·0	3·0	2·8
	U.T.S., tons/in.^{2}	13·8	14·4	14·5	14·1
	Elongation, %	68	72	65	72
10	0·1% P.S., tons/in.^{2}	13·3	13·5	13·7	13·5
	U.T.S., tons/in.^{2}	16·3	15·7	16·0	15·6
	Elongation, %	41	44	46	52
25	0·1% P.S., tons/in.^{2}	17·3	17·4	17·5	16·8
	U.T.S., tons/in.^{2}	20·2	19·7	20·1	20·0
	Elongation, %	19	23	19	15
50	0·1% P.S., tons/in.^{2}	20·5	21·2	21·3	21·1
	U.T.S., tons/in.^{2}	23·4	23·8	23·8	23·3
	Elongation, %	14	12	13	12

conductivity must be sacrificed. It will be appreciated that any decrease in conductivity in the rotor copper, by virtue of the greater heat generation and higher temperature produced, would increase the tendency to copper shortening. The conductivity at the operating temperature is clearly the true criterion in the comparison, and therefore conductivity measurements have been made on the materials under investigation at 130° and 170° C., as well as at room temperature. The values obtained are listed in Table IV. No figures are quoted for cold-worked, silver-free, tough-pitch coppers at 170° C., as softening of these materials occurs in relatively short periods at this temperature.

Table IV.—*Mass Conductivities of Silver-Bearing and Silver-Free O.F.H.C. and Tough-Pitch Coppers.*

Type of Copper	Cold Work, %	Conductivity *		
		20°C.	130°C.	170°C.
Tough-pitch	0	102·3	70·6	63·7
	10	100·8	69·8	...
	25	100·3	69·2	...
	50	99·7	69·4	...
Tough-pitch + Silver	0	101·9	70·5	63·5
	10	100·5	69·8	63·3
	25	100·4	69·8	63·1
	50	99·5	69·6	62·9
O.F.H.C.	0	102·3	70·9	63·9
	10	101·5	70·4	63·2
	25	100·1	71·1	64·5
	50	100·5	70·1	63·1
O.F.H.C. + Silver	0	102·0	71·0	63·9
	10	101·6	70·7	63·6
	25	101·1	70·3	63·2
	50	97·8	69·0	62·4

* International Annealed Copper Standard Units.

III.—STRESS-FREE SOFTENING TESTS

It was initially intended to study the effect of long-term heat-treatments at temperatures of 130°, 175°, 200°, and 225° C. As testing proceeded, however, it became clear that appreciable softening would occur in reasonable times only in the silver-free tough-pitch coppers and, in consequence, the initial tests were supplemented by tests at 250°, 275°, 300°, and 350° C. on the other coppers.

Heat-treatments of 8 hr. or less were carried out in a salt bath controlled to ± 2° C.; longer treatments were carried out in electric muffles, also controlled to ± 2° C. The surface hardness was measured by taking the average of five Vickers diamond impressions (20 kg. load) equally spaced across the full width of the strip, the two outer impressions being located $\frac{3}{16}$ in. from the edges. Hardness measurements were taken initially and after periods of heat-treatment arranged in geometrical progression. From the softening curves obtained the times to half-softening were determined, and the logarithms of these times were plotted against the reciprocal of the absolute temperatures at which the tests were made. This method of plotting has been shown by Cook and Richards [9] to give a linear relationship, a finding which is substantiated in the present tests.

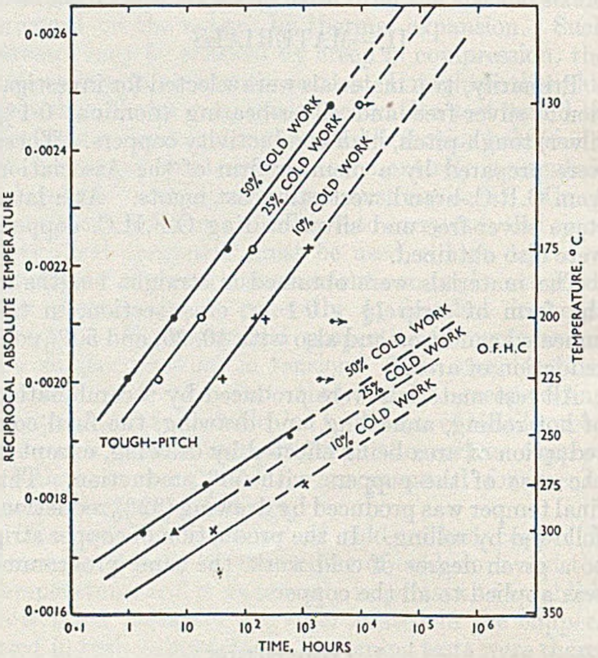

Fig. 1.—Relationship Between Time and Reciprocal Absolute Annealing Temperature to Produce 50% Softening of Cold-Worked Silver-Free O.F.H.C. and Tough-Pitch Coppers.

The results obtained for the tough-pitch coppers are shown in Figs. 1 and 2 (*a*), from which it is seen that the resistance to softening of tough-pitch copper is considerably increased at all three degrees of cold work by the addition of 0·1% silver. Thus, in the materials with 50% cold work, half-softening is reached in the silver-free copper in 10–15 hr. at 190° C., whereas in the silver-bearing copper the corresponding time would be greater than 10^6 hr.

Figs. 1 and 2 (*b*) show the results to date of softening tests made on the O.F.H.C. coppers. In Fig. 1 it is seen that the resistance to softening of silver-free O.F.H.C. copper is considerably greater than that of silver-free tough-pitch copper, probably because any impurities in the O.F.H.C. copper are in solution, whereas any in the tough-pitch copper exist as oxides. There is not such a great difference between the silver-free and the silver-bearing O.F.H.C. coppers as that which characterized the tough-pitch coppers. Comparing Figs. 2 (*a*) and 2 (*b*), it is seen that there is no appreciable difference in the resistance to softening, over the temperature range investigated, of the silver-bearing tough-pitch and O.F.H.C. coppers.

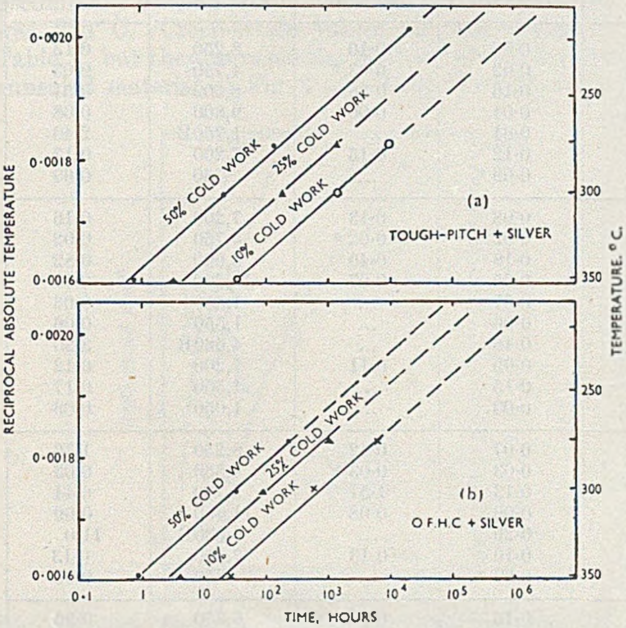

Fig. 2.—Relationship Between Time and Reciprocal Absolute Annealing Temperature to Produce 50% Softening of Cold-Worked Silver-Bearing O.F.H.C. and Tough-Pitch Coppers.

Thus it is clear that unless the presence of stress considerably modifies the softening characteristics—and the creep tests described later failed to show any evidence of this—softening of cold-worked copper used for alternator windings at temperatures up to 225° C. can be completely avoided by the use of silver-bearing copper.

IV.—CREEP TESTS

The specimens, to the dimensions shown in Fig. 3, were milled from the copper strips without surface machining. The effective gauge-length was 5 in. The extensometer was fixed rigidly to the enlarged ends of the specimen and the extension measured by two micrometers reading to 1×10^{-4} in., giving a strain sensitivity of $1{\cdot}7 \times 10^{-5}$. The temperature of the specimen was maintained constant to ± 1° C., while uniformity of temperature along the parallel length was better than ± 1° C. The equipment used is described elsewhere.[11] In starting a test a period of 4–6 hr. was used to bring the specimen to within 2% of the test temperature, this being followed by a further soaking period of 20 ± 4 hr. for close temperature adjustment before loading. The load was applied in increments, and the Young's modulus of the specimen at the test temperature was determined. This enabled the elastic component of the total strain to be calculated and hence the plastic strain determined. Plastic creep strain is used in the tables and graphs where comparison of creep properties is made. Three test-temperatures were used, viz. 130°, 175°, and 225° C.

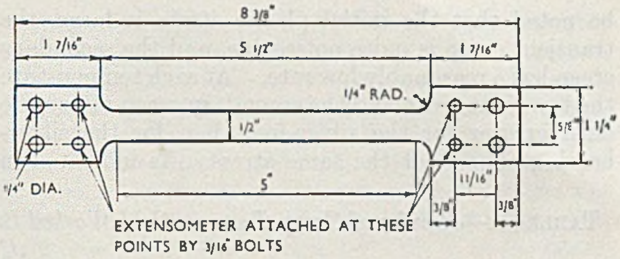

Fig. 3.—Creep-Test Specimen.

Creep tests up to 500 hr. in duration were carried out in triplicate at 130° C. on silver-free tough-pitch copper and at 225° C. on silver-bearing tough-pitch copper, and reproducibility of the results was found to be good.

The programme of tests was most complete for the tough-pitch coppers, the O.F.H.C. coppers being tested only at 130° and 225° C. and at a limited number of stresses. Some of the tests on the tough-pitch coppers were carried on to 10,000 hr. or even longer, but it became clear that comparisons made on a test duration of 5000 hr. or longer were equally valid at a duration of 1000 hr. In some of the longer tests at the lower stresses on the silver-bearing coppers, it was seen that the creep rate was still diminishing.

1. Comparison of Silver-Free and Silver-Bearing Coppers

Some tests were made on the annealed tough-pitch coppers, and the results are shown in Fig. 4. It will

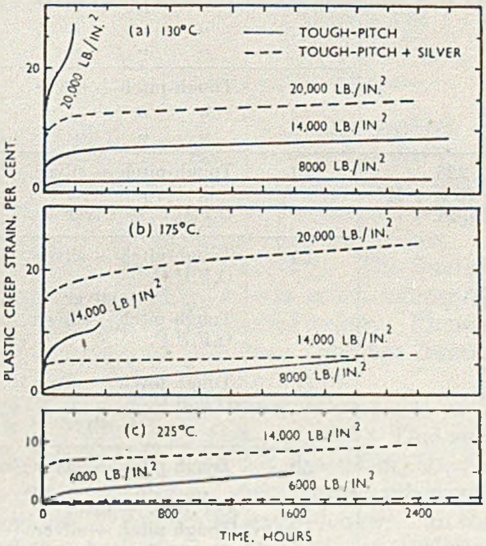

Fig. 4.—Creep Tests on Annealed Tough-Pitch Coppers Showing the Effect of Addition of 0·1% Silver.

be noted that the initial plastic strain is large, the transient creep is quite noticeable, and the secondary creep has a reasonably low rate. At each temperature the transient creep and the secondary creep rates were both greater for the silver-free than for the silver-bearing copper, at the same stress. As indicated in Table II, the materials had the same grain-size. It is clear that the silver addition, which is probably in solid solution at the test temperatures, has increased the resistance to creep.

The results of the tests on the cold-worked materials are given in Table V. The first point to which

Table V.—*Results of Creep Tests on Cold-Worked Silver-Free and Silver-Bearing Tough-Pitch and O.F.H.C. Coppers.*

Test Temp., °C.	Cold Work, %	Type of Copper	Stress, lb./in.^{2}	Plastic Creep Strain, %, at 1000 hr.	Plastic Creep Strain, %, at 5000 hr.	Duration of Test, hr.	Plastic Creep Strain at End of Test, %
130 ($E = 16{\cdot}0 \times 10^6$ lb./in.^{2})	10	Tough-pitch	8,000	0·06	0·10	8,250	0·13
		,, ,, + silver	8,000	0·02	0·03 *	4,750	0·03
		,, ,,	14,000	0·16	0·36	8,600	0·51
		,, ,, + silver	14,000	0·04	0·06	9,800	0·08
		,, ,,	20,000	0·64	...	1,750B	2·40
		,, ,, + silver	20,000	0·12	0·15	7,200	0·17
		O.F.H.C. + silver	20,000	0·09 *	...	750	0·09
	25	Tough-pitch	8,000	0·08	0·13	7,200	0·16
		,, ,, + silver	8,000	0·02	0·02 *	4,750	0·02
		,, ,,	14,000	0·18	0·40	8,600	0·82
		,, ,, + silver	14,000	0·05	0·07	10,200	0·07
		O.F.H.C.	14,000	0·07	...	1,550	0·08
		,, + silver	14,000	0·06	...	1,550	0·06
		Tough-pitch	20,000	0·46	...	4,680B	3·30
		,, ,, + silver	20,000	0·09	0·11	7,200	0·12
		O.F.H.C.	20,000	0·15	...	1,550	0·17
		,, + silver	20,000	0·09	...	1,050	0·09
	50	Tough-pitch	8,000	0·07	0·42	8,250	1·36
		,, ,, + silver	8,000	0·03	0·03 *	4,550	0·03
		,, ,,	14,000	0·13	1·51	8,700	6·44
		,, ,, + silver	14,000	0·06	0·08	11,400	0·09
		,, ,,	20,000	0·29	...	4,030B	11·0
		,, ,, + silver	20,000	0·10	0·13	7,250	0·13
		O.F.H.C. + silver	20,000	0·09 *	...	750	0·09
175 ($E = 15{\cdot}5 \times 10^6$ lb./in.^{2})	10	Tough-pitch	8,000	0·16	0·62	6,850	0·96
		,, ,, + silver	8,000	0·04	0·06 *	4,850	0·06
		,, ,,	14,000	1·12	...	1,100B	2·1
		,, ,, + silver	14,000	0·09	0·12	12,900	0·16
		,, ,, ,,	20,000	0·24	0·38 *	4,900	0·38
	25	Tough-pitch	8,000	2·93	...	1,050	2·96
		,, ,, + silver	8,000	0·04	0·05 *	4,900	0·05
		,, ,, ,,	14,000	0·08	0·10	12,900	0·13
		,, ,, ,,	20,000	0·14	0·21	10,300	0·26
	50	Tough-pitch + silver	8,000	0·04	0·06	6,900	0·07
		,, ,, ,,	14,000	0·09	0·14	12,900	0·16
		,, ,, ,,	20,000	0·17	...	3,700	0·22
225 ($E = 15{\cdot}0 \times 10^6$ lb./in.^{2})	10	Tough-pitch + silver	8,000	0·09	0·15	8,900	0·18
		,, ,, ,,	14,000	0·20	0·39	12,900	0·68
		O.F.H.C. + silver	14,000	0·16	...	1,050	0·16
	25	Tough-pitch + silver	8,000	0·07	0·11	8,900	0·13
		O.F.H.C.	8,000	0·17	...	2,300	0·22
		,, + silver	8,000	0·08	...	3,150	0·11
		Tough-pitch + silver	14,000	0·16	0·27	11,500	0·47
		O.F.H.C.	14,000	0·65	...	2,750	1·49
		,, + silver	14,000	0·21	...	3,100	0·31
		Tough-pitch + silver	20,000	0·35	0·76	9,870B	1·5
		O.F.H.C.	20,000	...	...	275B	2·4
		,, + silver	20,000	0·61	2·43	5,000	2·43
	50	Tough-pitch + silver	8,000	0·09	0·14	8,900	0·18
		,, ,, ,,	14,000	0·20	0·34	12,900	0·62
		O.F.H.C. + silver	14,000	0·14	...	3,250	0·19
		Tough-pitch + silver	20,000	0·42	...	3,000	1·00

* Extrapolated result. B = broken. E = Young's Modulus.

attention should be drawn is that, provided it was retained at the test temperature, cold work materially increased the creep-resistance. Furthermore, the increase in creep-resistance was not essentially different for the three degrees of cold work used. Where the test temperature was such that softening of a cold-worked material occurred either during the preliminary heating period before application of the stress or at an early stage in the test, the subsequent creep-resistance became similar to that of the annealed material. Thus, the silver-free tough-pitch copper, with each degree of cold work, softened during the preliminary heating period when the test temperature was 225° C. Creep-strain values are not given in Table V, but they were similar to those shown for the annealed material in Fig. 4 (*c*). Only the 25% cold-worked silver-free O.F.H.C. copper was tested at 225° C., and this gave no indications that softening was occurring during the creep tests lasting over 2000 hr.

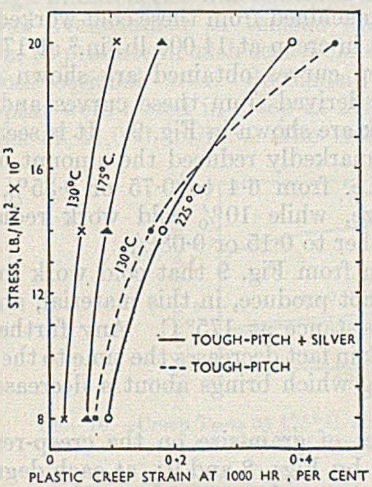

Fig. 5.—Creep Tests on Cold-Worked Tough-Pitch Coppers Showing the Effect of Addition of 0·1% Silver. Mean results for 10, 25, and 50% cold-worked materials.

Young's Modulus at Test Temperatures.

130° C. . . .	$16{\cdot}0 \times 10^6$ lb./in.².
175° C. . . .	$15{\cdot}5 \times 10^6$ lb./in.².
225° C. . . .	$15{\cdot}0 \times 10^6$ lb./in.².

In the tests at 130° C., at each degree of cold work and at each value of stress used, the silver-bearing tough-pitch copper was noticeably more resistant to creep than the silver-free, tough-pitch copper. At this temperature there was little, if any, difference between the two silver-bearing coppers, tough-pitch and O.F.H.C. The silver-free O.F.H.C. had greater creep-resistance than the silver-free tough-pitch, but the resistance was not as high as that of the silver-bearing coppers.

At 175° C. comparison was made only between the two tough-pitch coppers, and the creep-resistance of the silver-bearing copper was much greater than that of the silver-free.

At 225° C. there was again little difference between the creep-resistances of the silver-bearing coppers, tough-pitch and O.F.H.C., while both were noticeably more resistant than the silver-free O.F.H.C. copper.

A more readily appreciated comparison of the creep properties of the silver-free and silver-bearing tough-pitch coppers is shown in Fig. 5. In this, the mean value of the plastic creep strain at 1000 hr. for the three degrees of cold working is plotted against the stress. It is seen that the silver-bearing copper has about the same creep-resistance at 225° C. as the silver-free copper at 130° C.

2. Fractures in Silver-Free and Silver-Bearing Coppers

Some fractures have been obtained in all materials under creep conditions and details are given in Table VI.

At 130° and 175° C. fractures have been obtained only of the silver-free tough-pitch coppers, while in the case of silver-free O.F.H.C. and silver-bearing tough-pitch and O.F.H.C. coppers, fractures have been obtained only at 225° C., because at the lower temperatures the stresses used in the creep tests were not high enough to cause fracture in the testing time. In all cases the fractures were intercrystalline, and intercrystalline cracking was observed in micro-specimens taken at different places along the parallel gauge-length of the creep specimens.

Table VI.—*Details of Creep Fractures of Coppers.*

Type of Copper	Cold Work, %	Test Temp., °C.	Stress, lb./in.²	Life, hr.	Total Elongation at Fracture, % on 5 in.
Tough-Pitch	Annealed	130	20,000	170	29·8
	10			1750	2·4
	25			4680	3·4
	50			4030	11·0 [a]
	10	175	14,000	1100	2·0
	25			365	14·6 [a]
	50			336	31·4 [a]
Tough-pitch + silver	25	225	25,000	800	6·0
	25		20,000	9870	1·5
O.F.H.C.	25		20,000	275	2·4
O.F.H.C. + silver	10		20,000	1368	2·5
	25		25,000	520	2·8

[a] These specimens recrystallized during test.

The specimen of the 25% cold-worked, silver-bearing, tough-pitch copper stressed at 20,000 lb./in.² broke while still in the second stage of creep (minimum creep-rate stage), and gave no warning of its impending fracture by a turning-up of the creep curve. Plastic creep strain at fracture was 1·5%. The fracture was intercrystalline, and there was much intercrystalline cracking throughout the parallel length. Photographs showing cracking at and away from the fracture are given in Figs. 6 and 7 (Plate XXV).

A creep test has also been carried out at 25,000 lb./in.² and 225° C. on this material. This specimen fractured in 800 hr. with 2% extension, the fracture being intercrystalline, with some intercrystalline cracking along the parallel gauge-length. In this case the fracture took place in the final, or tertiary, stage of creep. Creep fractures obtained in the silver-free and silver-bearing O.F.H.C. coppers were intercrystalline and similar to those obtained in the tough-pitch coppers.

The test results indicate that in the silver-free tough-pitch copper at 130° C. and stresses of 20,000 lb./in.^{2} or less, or at 175° C. and stresses of 14,000 lb./in.^{2} or less, any creep failures would be intercrystalline, since at a given temperature the probability that ultimate fracture will be intercrystalline increases with decrease in the rate of creep strain, i.e. with decrease in applied stress.

In the silver-bearing tough-pitch copper and silver-free and silver-bearing O.F.H.C. coppers, the only creep fractures obtained have been at 225° C. and 20,000 and 25,000 lb./in.^{2}. For similar reasons to those given above, creep fractures in these materials at 225 °C. and stresses of 20,000 lb./in.^{2} or less would also be intercrystalline.

3. Comparison of Rolled and Drawn Silver-Bearing Tough-Pitch Copper and the Effect of Thickness

Tests of 2000 and 3000 hr. duration were carried out at 175° C. and a stress of 14,000 lb./in.^{2} on materials prepared by rolling and by drawing and at two different thicknesses, 0·1 and 0·2 in. The results obtained showed that there was no difference between the creep-resistance of material produced by these two methods, nor was there any significant effect of thickness of material.

4. Effect of Cold Work and of Grain-Size in Silver-Bearing Tough-Pitch Copper

A comparison of the creep curves for the silver-bearing tough-pitch coppers under the same test conditions, but with different amounts of cold work from 10 to 50% showed that there was no marked effect of degree of cold work within this range. This interesting result suggests the choice of a low temper for alternator windings, so that benefit may be derived from its greater resistance to softening. It was felt, however, that further information on this was required, and a number of tests was carried out to determine the influence on the creep-resistance of a range of cold work. At the same time the effect of grain-size was investigated.

Lengths of the silver-bearing tough-pitch coppers, originally with 10 and 50% cold work, were bright-annealed for 2 hr. at 570° C., giving annealed materials with two different grain sizes—0·023 and 0·043 mm. dia. Lengths of material of each grain-size were strained in tension at room temperature to give reductions in area of 5, 10, 15, and 20%. Creep specimens machined from these cold-worked materials were tested in creep at 14,000 lb./in.^{2} at 175° C.

The creep curves obtained are shown in Fig. 8, while data derived from these curves and from the earlier work are shown in Fig. 9. It is seen that 5% cold work markedly reduced the amount of creep at 2000 hr., i.e. from 6·4 to 0·75 or 0·35% according to grain-size, while 10% cold work reduced these figures further to 0·15 or 0·08%.

It is seen from Fig. 9 that cold work in excess of 10% does not produce, in this material, any increase in creep-resistance at 175° C. Any further increase in cold work in fact decreases the time to the beginning of softening which brings about a decrease in creep-resistance.

The effect of grain-size on the creep-resistance is also shown by Figs. 8 and 9; at each degree of cold work the creep strain at 2000 hr. of the coarse-grained specimen was approximately one-half that of the fine-grained specimen.

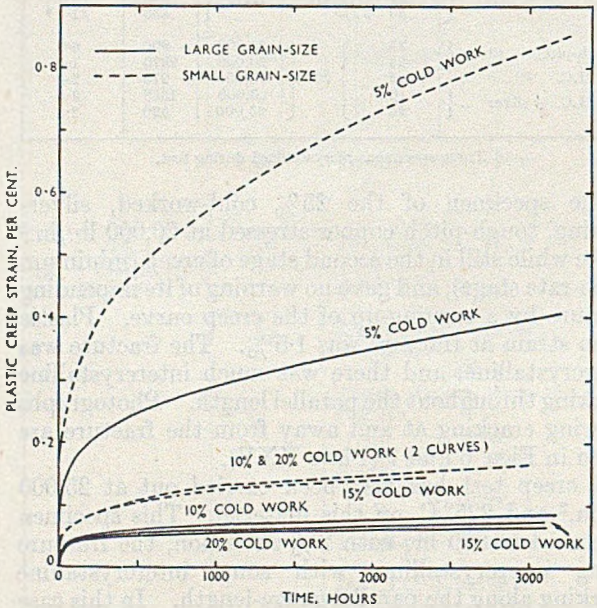

FIG. 8.—Effect of Cold Work and Grain-Size on the Creep of Silver-Bearing Tough-Pitch Coppers Stressed at 14,000 lb./in.^{2} at 175° C.

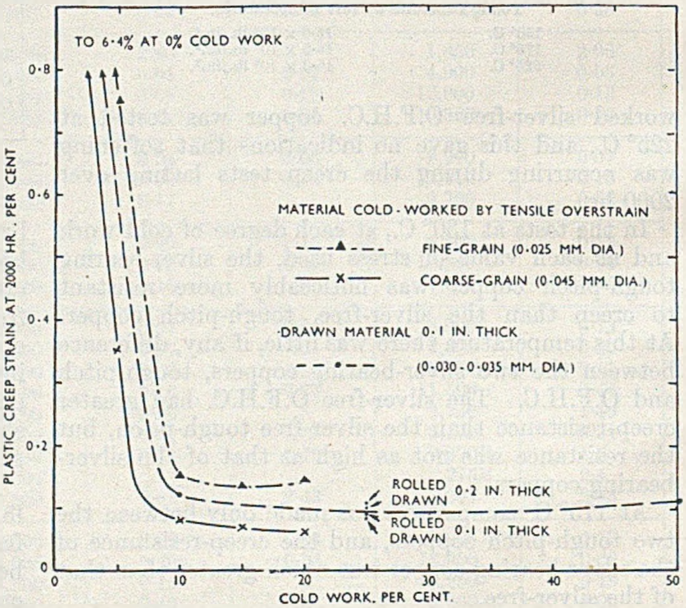

FIG. 9.—Effect of Cold-Work on the Creep of Silver-Bearing Tough-Pitch Coppers Stressed at 14,000 lb./in.^{2} at 175° C.

5. The Effect of Tensile and of Compressive Working on Creep Behaviour

While the cold working of the copper conductors for alternator windings is generally carried out by drawing and hence is of the nature of tensile overstrain, the copper in service in the alternator is subjected to compressive stresses, under the action of which relaxation creep occurs. It is well known that if a specimen of any material is overstrained in tension and then tested in compression, there is a decrease in the stress below which a linear relationship exists between stress and strain. It was considered of interest in the general problem of copper shortening to observe the effect of previous tensile and compressive overstrains on the subsequent creep behaviour under tensile stress. Some 1-in.-dia. bar of 0·1% silver-bearing tough-pitch copper was annealed and part of the bar was overstrained in tension and part in compression, to produce in each case a change of 10% in the cross-sectional area. Specimens machined from each part were tested in creep at 175° C. and a tensile stress of 14,000 lb./in.^{2}. The results of the tests are given in Fig. 10, where a comparison has also been made with the results of the creep test on the drawn silver-bearing strip cold worked 10% and used in the main programme. It will be seen that while the specimen overstrained in tension gave a similar creep curve to that of the drawn material, the specimen overstrained in compression gave a higher initial strain but subsequently the same rate of creep.

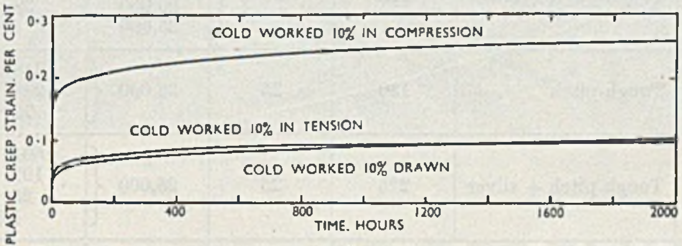

FIG. 10.—Creep Tests at 175° C. and 14,000 lb./in.^{2} on Silver-Bearing Tough-Pitch Coppers Showing the Effect of the Direction of Working on Tensile Creep Behaviour.

These results indicate that drawn material tested in creep under compressive stress will give a greater initial strain and a greater plastic strain at any subsequent time than when tested in creep under tensile stress. It is reasonable to suppose, however, that the greater resistance to plastic deformation of the silver-bearing compared with the silver-free copper shown in the tensile creep tests would also be shown in compressive creep tests.

V.—CREEP RECOVERY

When a tensile creep specimen is unloaded at the test temperature, the strain in the specimen is reduced, first by an immediate contraction corresponding to the elastic recovery and hence related directly to the stress in the creep test, and then, gradually with time, by a further contraction resulting from plastic flow. The contraction/time curve is similar in shape to a normal creep curve, the rate of plastic flow being high immediately the load is removed and decreasing with time. The curve appears to become asymptotic to some value of strain. In creep-recovery tests on coppers continued for 100 hr., two thirds of the total recovery occurred in the first 20 hr.

1. Previous Work

Tapsell [12] carried out creep-recovery tests on a nickel–chromium–molybdenum steel and on a 0·13% carbon steel at 450° C., on a 3% nickel steel at 400° C., and on lead at 60° C. The following conclusions were reported:

(*a*) Recovery over a given period after removal of stress was proportional to the original applied stress.

(*b*) The estimated final amount of creep recovery increased with the duration of the preceding creep test.

(*c*) The rate of recovery increased considerably with increase in temperature.

(*d*) A close approximation to a linear relationship existed in nearly all cases between the logarithm of the sum of the rapid elastic contraction and the creep recovery and the logarithm of the time.

Two theories have been put forward to account for creep recovery. The first supposes that in the material under test the grains, according to their orientation, will have varying resistance to deformation in the direction of stress. After a period of creep testing, some grains will be more heavily stressed than others and hence, on unloading, while there will be an immediate elastic recovery, additional recovery will occur with time as negative creep is produced by the action of the uneven stress distribution. Under some conditions, particularly at high temperatures, a large part of the creep recovery has been attributed to viscous flow in grain-boundary regions under the influence of elastic stresses in adjoining grains.

The second theory suggests that creep recovery may be due to an extra-elastic effect, i.e. a change in lattice strain with time in the direction of the applied stress during the creep test, and a change in the reverse direction during recovery. As creep recovery has not been observed in single-crystal specimens, the first theory is thought to be the more likely.

2. Present Experimental Work

In an attempt to obtain some information on the relative creep recovery of silver-bearing and silver-free coppers the following tests were made.

(a) *Effect of Creep Stress on Creep Recovery*

Four specimens of 25% cold-worked, silver-bearing, tough-pitch copper stressed at 8000, 14,000, 20,000, and 25,000 lb./in.^{2} at 225° C. were unloaded when the plastic creep strain in each case reached approximately 0·15%. The creep recovery was followed at 225° C. and the results obtained are given in Table VII (*a*).

Only the specimen tested at 8000 lb./in.^{2} was in the apparent minimum creep-rate stage before unloading; all the others were still in the primary stage of creep.

Table VII.—*Factors Affecting Creep Recovery in Copper.*

Effect Investigated	Type of Copper	Test Temp., °C.	Cold Work, %	Creep Stress before Unloading, lb./in.^{2}	Creep-Test Time before Unloading, hr.	Creep-Strain before Unloading, %	Creep Recovery at 100 hr., %
(*a*) Effect of creep stress at constant creep strain (0·15%)	Tough-pitch + silver	225	25	8,000	8856	0·130	0·020
				14,000	265	0·148	0·041
				20,000	29	0·138	0·043
				25,000	1·3	0·148	0·031
(*b*) Effect of varying creep strain at constant creep stress (25,000 lb./in.^{2})	Tough-pitch	130	25	25,000	648	1·064	0·099
					296	0·515	0·079
					72	0·329	0·102
	Tough-pitch + silver	225	25	25,000	504	1·102	0·076
					192	0·501	0·072
					25	0·302	0·053
					1·3	0·148	0·031
(*c*) Effect of varying amounts of cold work	O.F.H.C.	225	10	6,000	1128	0·071	0·018*
			25		1104	0·097	0·018
			50		1248	0·090	0·016

* In 48 hr. (thermostat failure).

The results do not indicate any simple relationship between creep recovery and creep-test stress.

(b) *Effect of Creep Strain at Constant Stress*

Specimens stressed at 25,000 lb./in.^{2} were unloaded at plastic creep strains of approximately 0·3, 0·5, and 1·0%, and creep recovery allowed to occur at the creep-test temperature.

The tests were made on 25% cold-worked, silver-bearing, tough-pitch copper at 225° C., and on 25% cold-worked, silver-free, tough-pitch copper at 130° C. The results of these tests are given in Table VII (*b*). The results for the silver-bearing specimen unloaded at 0·148% strain, described in (*a*) above, is also included.

In the silver-bearing materials at 225° C., the greater the creep strain, the greater was the creep recovery; but in the silver-free material at 130° C. there was no systematic relationship between creep strain and creep recovery.

(c) *Effect of Cold Work on Creep Recovery*

Three creep specimens of silver-free O.F.H.C. copper with 10, 25, and 50% cold work tested at 225° C. and 6000 lb./in.^{2} were unloaded at approximately the same test times and creep strains, and the creep recovery followed at 225° C. Results are given in Table VII (*c*) and show that there is no effect of cold work from 10 to 50% on creep recovery. This result would be expected from the creep tests, which show no marked effect of this variation in cold work on the resistance to creep.

(d) *Cumulative Effect of Intermittent Creep and Creep Recovery*

In practice, some large turbo-alternators, forming part of stand-by equipment of power-stations, are run only at times of peak load. In consequence, the rotor windings of such machines are subjected to alternate stressing and unstressing, resulting in creep and creep recovery. It is usual for the running times at full load of such equipment to be short, say of a few hours' duration only, and for the creep recovery to occur under conditions of slow cooling of the windings. Such conditions cannot be reproduced conveniently in creep testing, where the requirements of temperature constancy impose certain limitations. It was considered however, that the approach described below gave information on the effect of creep recovery in conditions not too widely different from those met with in the actual windings of an alternator.

Two specimens of 50% cold-worked, silver-bearing, tough-pitch copper were tested in tensile creep at 175° C. and 20,000 lb./in.^{2}. One specimen was stressed continuously. The other specimen was stressed first for 624 hr., and thereafter it was unloaded for 168 hr. (one week) and reloaded for 168 hr.; this cycle of loading and unloading being repeated up to a total time of 3000 hr. The total deformation (including elastic strain) with time in these tests is plotted in Fig. 11, and it is seen that a steady state of increase in total strain in each cycle has been reached.

CREEP FAILURE IN SILVER-BEARING TOUGH-PITCH COPPER, 25% COLD-WORKED.

Temperature : 225° C. Stress : 20,000 lb./in.^{2}
Life : 9870 hr. Elongation : 1·5%

FIG. 6.—Intercrystalline Cracking in Vicinity of Fracture. Longitudinal section parallel to surface. × 125.

FIG. 7.—Intercrystalline Cracking 3 in. Away from Fracture. Longitudinal section parallel to surface. × 500.

This comparison of the intermittently loaded specimen with the continuously loaded one is interesting, since the effective creep rate of the former approximates closely to half the steady creep rate of the latter, while the time under load in these two tests is in the same ratio as the creep rates. It is clear that the plastic strain recovered during negative creep is lost during the next loading cycle. Hence any differences in the creep-recovery characteristics of two materials are of no consequence, whereas differences in creep-resistance are important.

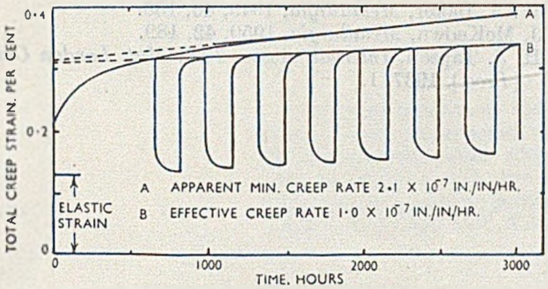

Fig. 11.—Comparisons of Creep Tests using Continuous and Intermittent Stress on 50% Cold-Worked Silver-Bearing Tough-Pitch Coppers at 175° C. and 20,000 lb./in.^{2}.

3. Discussion of Results

In dealing with the subject of creep recovery in relation to the problem of copper shortening, the following considerations are relevant:

(*a*) The fullest extent of creep recovery is obtained only when the material is maintained after unloading at the same temperature as that at which creep had occurred. This condition is not normal in the rotor windings of a turbo-alternator, since when the machine is stopped the exciting current is switched off. Even when the creep-recovery temperature is the same as the creep-test temperature, the amount of recovery is small, 10–30% of the creep strain before unloading. When the recovery temperature is lower than the creep-test temperature, the amount of recovery will be still smaller, and hence is likely to be of negligible importance in turbo-alternator rotor windings.

(*b*) On re-stressing, creep occurs at first more rapidly than if recovery had not taken place, so that the recovery is to a large extent cancelled.

(*c*) Since creep recovery is a negative form of creep, the more resistant a material is to creep, the less creep recovery would be expected from it. Thus, since silver-bearing copper is more resistant to creep than silver-free copper, it must be expected to give less creep recovery, when unloaded from the same stress at the same temperature. It has been noted earlier that the creep-resistance of the silver-bearing tough-pitch copper at 225° C. is similar to that of the silver-free tough-pitch copper at 130° C. The test results given in Table VII (*b*) show that there is a corresponding similarity in the amount of creep recovery in the silver-bearing copper at 225° C. and the silver-free copper at 130° C.

VI.—CONCLUSIONS

The conclusions reached as a result of the investigations described above may be summarized briefly as follows:

(1) Creep tests at 130°–225° C. and of durations up to 10,000 hr. have shown that the addition of silver to cold-worked tough-pitch copper greatly increases its resistance to creep.

(2) The creep-resistance of all the coppers examined increased greatly with cold work up to 10% reduction and thereafter was little affected. In silver-bearing tough-pitch copper, 10% cold work is sufficient to produce the optimum combination of creep-resistance and resistance to softening over the temperature range 130°–225° C.

(3) Cold-worked O.F.H.C. copper containing silver has softening and creep characteristics closely similar to those of cold-worked tough-pitch copper containing silver.

(4) If both are free from silver, and in the absence of stress, the resistance to softening of cold-worked O.F.H.C. copper is much greater than that of cold-worked tough-pitch copper. If silver (0·08%) is added to each of these two coppers, their resistance to softening becomes equal and higher than that of either of the silver-free materials.

(5) In comparative creep tests at 130°–225° C. on cold-worked silver-free materials the creep-resistance of O.F.H.C. copper was much greater than that of tough-pitch copper. Presumably this is due in part to the increased resistance to softening of the O.F.H.C. copper and in part to its inherently greater resistance to creep deformation.

(6) Creep tests on rolled and drawn silver-bearing tough-pitch copper showed no effect of the method of production or of the thickness of the material.

(7) When recrystallization did not occur during a creep test, the fractures obtained were markedly intercrystalline, with intercrystalline cracking throughout the parallel length. Such fractures occurred at small extensions of the order of 2–3%. When recrystallization did occur during a test, considerably larger extensions were obtained, and in such cases, although the ultimate fracture was partially intercrystalline, relatively few intercrystalline cracks were observed in the regions of the specimen away from the fracture.

(8) An investigation of creep recovery in cold-worked silver-free and silver-bearing tough-pitch coppers at temperatures of 130° and 225° C. has shown that creep recovery is greater in the silver-free than in the silver-bearing copper, as would be expected from the greater creep-resistance of the latter.

(9) On reloading a specimen in which creep recovery has occurred, creep is more rapid until the strain reaches about the value before unloading. Thereafter it creeps at the same rate as before unloading. It follows that any creep recovery occurring in a turbo-alternator rotor winding during the shut-down period does not materially reduce copper shortening.

ACKNOWLEDGEMENTS

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THE ALLOYS OF MOLYBDENUM AND TANTALUM *

1334

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SYNOPSIS

An investigation of the binary system molybdenum–tantalum has shown that these metals form a continuous series of solid solutions. This is to be expected on theoretical grounds, as both metals crystallize with a body-centred cubic structure and the atoms are very similar in size. No superlattice was detected and no anomalies occur between the true and X-ray densities. Approximate melting points of the alloys have also been determined. The alloys of the transition metals of Groups IVA, VA, and VIA with each other are discussed briefly.

I.—INTRODUCTION

The binary alloy system molybdenum–tantalum has not been investigated in detail by previous workers. Von Bolten [1] in 1905 stated that molybdenum and tantalum alloyed in all proportions, but gave no further details. More recently Bückle [2] investigated the lattice parameter and hardness of three alloys and from them was able to postulate a continuous series of solid solutions. Myers [3] found that alloys containing up to 20 wt.-% molybdenum were single phase.

The two metals crystallize in the same structure, body-centred cubic, have similar electrochemical properties, and their atoms differ in size by less than 5%; thus it is to be expected that they will form extensive solid solutions with each other.

II.—EXPERIMENTAL METHODS

1. Materials Used

Alloys were prepared both from the metal powders and from the lump material. The metal powders were supplied by Murex, Ltd., who specify a purity of 99·8% for the molybdenum and 99·5% for the tantalum. The lump materials used were in the form of H.S. Spectrographic Standard rods supplied by Johnson, Matthey and Co., Ltd. The manufacturers claim a purity higher than 99·95% for the molybdenum and higher than 99·98% for the tantalum. Spectroscopic analysis indicated traces of copper, magnesium, and calcium in the molybdenum and of niobium, nickel, tungsten, silicon, iron, calcium, and copper in the tantalum.

2. Preparation of the Alloys

An initial attempt was made to prepare the alloys by powder-metallurgical methods. One-g. samples of the powders in the desired proportions were mixed in a small ball mill, described elsewhere,[4] for 24 hr. The mixed powder was then compacted at 40 tons/in.^{2} and the pellets sintered at 1500° C. for 24 hr. in a vacuum of 10^{-4} mm. Hg or better. Even under these conditions some atmospheric contamination occurred, and equilibrium was not attained in the tantalum-rich alloys. It was not possible to attain equilibrium in these alloys in a reasonable time even when a sintering temperature of 1850° C. was used.

To overcome these difficulties further alloys were made by melting. When the metal powders were used, these were first mixed in the required proportions, pressed at 40 tons/in.^{2} to form pellets about 1 cm. × 1 cm. dia. and weighing about 5 g. The pellets were sintered at 1500° C. for 2 hr. in a vacuum of 10^{-5} mm. Hg to remove gases, and were then melted in a purified argon atmosphere in an arc furnace which is described in detail elsewhere.[5] With the spectroscopically pure metal rods no preliminary sintering was carried out, the alloys, weighing about 5 g., being made up directly by melting in the arc furnace. In this furnace the specimen forms the positive pole of a D.C. arc, the negative pole of which is a tungsten rod. The specimen rests on a water-cooled copper hearth and is melted in an atmosphere of argon at about 20 cm. pressure. The argon atmosphere in the sealed furnace vessel is first purified by melting a piece of zirconium, which effectively removes any residual gases. Spectroscopic analyses have shown that there is no pick-up of copper or tungsten during the melting, and other experiments with the furnace have shown that gaseous contamination is also negligible.

Both Myers,[3] with tantalum-rich alloys prepared by sintering *in vacuo* at 2600° C., and Schumb, Radtke, and Bever,[6] with alloys prepared by arc melting, found that there was a loss of molybdenum during

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heating due to evaporation. No such loss was found in the present case, as can be seen from Table I. The analyses in the table were carried out by Johnson, Matthey and Co., Ltd. The greatest deviation from the nominal composition is less than 1·5%, and is not significantly greater for the alloys prepared from the metal powders.

TABLE I.—*Compositions of Alloys Studied.*

Intended Composition, Ta, wt.-%	Analysed Composition		Tantalum, at.-%	Materials Used
	Mo, wt.-%	Ta, wt.-%		
0	100	...	0	Murex metal powders
20	80·66	19·28	11·2	
40	61·21	38·78	25·2	
60	40·28	59·76	44·1	
70	31·24	68·72	53·8	
80	20·32	79·64	67·5	
90	11·20	88·83	80·8	
100	...	100	100	
0	100	...	0	H.S. Spec-pure metals
10	90·24	9·70	5·4	
30	70·97	28·98	17·8	
50	49·90	50·12	34·8	
95	5·75	94·19	89·6	
100	...	100	100	

III.—EXPERIMENTAL RESULTS

1. X-RAY ANALYSIS

Powder photographs were taken in a 9-cm. " Unicam " camera, using copper radiation. This radiation is very suitable for the accurate determination of the lattice parameters of molybdenum and tantalum. For the former, the 400 reflection occurs at 78° and for the latter the 330 reflection at 82°. Nelson and Riley's [7] function was used to obtain the extrapolated values of the parameters at 90°.

The alloys of intermediate composition were brittle and could be powdered readily to pass a 200-mesh sieve; for the other alloys filings of less than 200 mesh were used. The powder samples were annealed at 1200°–1500° C. for 5 hr. in molybdenum boats in evacuated silica tubes. This annealing treatment was required to obtain sharp diffraction photographs with resolved high-angle doublets. All the alloys were single phase with a body-centred cubic structure. The lattice parameters, given in Table II, have been corrected to 20° C., assuming the thermal coefficients of expansion of the alloys to vary linearly between those of the two pure metals. The coefficients of expansion of molybdenum and tantalum are very similar, 4.9×10^{-6} and 6.5×10^{-6} respectively, and with the small temperature corrections that are involved this appears to be justified.

The parameters of the alloys prepared by sintering for 1 hr. at 1850° C. are also included in Table II for comparison. They show that equilibrium had not been obtained; this was most marked in the tantalum-rich alloys, which X-ray photographs showed to consist of two body-centred cubic phases.

TABLE II.—*Lattice Spacings of Molybdenum–Tantalum Alloys.*

Tantalum, at.-%	Lattice Spacing at 20° C., kX	
	Arc-Melted Alloys	Sintered Alloys
0	3·1406	3·1404
0	3·1404	...
11·2	3·1541	3·1478
25·2	3·1747	3·1690
34·8	3·1858	3·1795
44·1	3·2007	3·1923
53·8	3·2150	3·2140
67·5	3·2372	...
80·8	3·2627	...
100	3·2973	...
100	3·2974	...

To determine whether a superlattice was formed, powdered samples of alloys containing 44·1, 53·8, and 67·5% tantalum were annealed at 1500° C. for 5 hr. and then at 400° C. for 100 hr. No superlattice lines were detected in the X-ray powder photographs.

A lattice parameter/composition curve is plotted in Fig. 1. There is a slight negative deviation from Vegard's law, which reaches a maximum at 62·5% tantalum. Data from Bückle [2] and Myers [3] are also included in Fig. 1. The data from the former are for alloys sintered at 2100°–2600° C. and those from the latter for tantalum-rich alloys sintered at 2600° C. The values, particularly those of Myers, lie close to the curve.

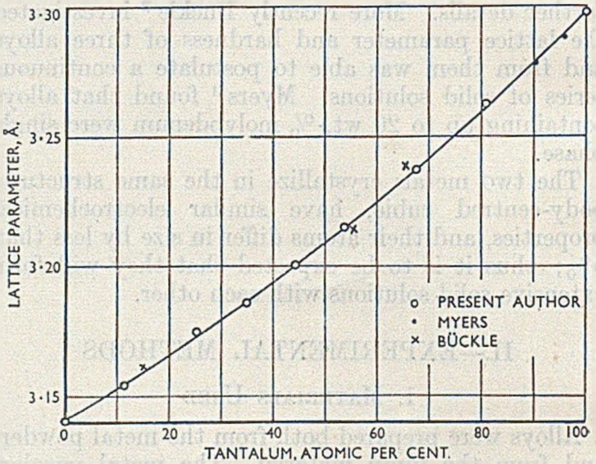

FIG. 1.—Lattice Parameters of Molybdenum–Tantalum Alloys.

2. DENSITY MEASUREMENTS

Before density measurements were made, the alloys were sectioned or broken into lumps to determine whether there was any porosity. No blowholes were found, and the densities of the lump alloys were measured by means of a density bottle. The values are given in Table III, which also includes the densities calculated from the lattice parameters using the formula $\rho = KnM/V^1$, where $n =$ the number of atoms per unit cell, $M =$ the averaged atomic weight, V^1 = the volume of the unit cell in kX^3 using the Siegbahn X-ray wave-lengths, and K a numerical constant = 1·6502, which takes account of Avogadro's number and the conversion of kX units to centimetres.

TABLE III.—*Densities of Molybdenum–Tantalum Alloys.*

Tantalum, at.-%	Density, g./c.c.	
	Measured at 20°C.	Calculated
0	10·17	10·22
5·4	10·65	10·64
17·8	11·65	11·68
34·8	12·87	12·80
44·1	13·50	13·43
53·8	14·13	14·06
67·5	14·95	14·91
80·8	15·70	15·63
100	16·61	16·65

The calculated densities agree with the experimentally determined values to within 0·5%, which is within the experimental error of the density measurements.

3. MELTING POINTS

The approximate melting points of some of the alloys were determined in the arc furnace, using an optical pyrometer. The alloys were loaded into the furnace with samples of pure molybdenum and tantalum, and were melted in turn without altering the conditions in the furnace. The technique used was to maintain a small molten pool of the alloy on the surface of the sample and to measure the temperature of the junction of the molten and solid metal with a disappearing-filament pyrometer. There was considerable adsorption of radiation in the argon atmosphere and the silica window, but from the observed melting points of the pure metals a correction factor can be obtained. To check that the adsorption conditions did not alter, the melting points of the standards were determined both before and after melting the alloys. For the degree of accuracy that can be realized by this method, which measures a temperature between the liquidus and solidus, a linear correction factor is justified; this varies from 400° C. at the melting point of molybdenum to 460° C. at that of tantalum. The observed and corrected melting points given in Table IV are possibly correct to about ± 50° C. Values of 2620° and 3000° C. are assumed for molybdenum and tantalum respectively.

TABLE IV.—*Melting Point of Molybdenum–Tantalum Alloys.*

Tantalum, at.-%	Observed Melting Points, °C.		Corrected Melting Point, °C.
	Expt. No. 1	Expt. No. 2	
0	2240	2200	2620*
11·2	...	2200	2630
25·2	...	2260	2700
44·1	...	2300	2760
53·8	2390	...	2810
67·5	2420	...	2850
80·8	2450	...	2880
100	2560	2520	3000 *

* Assumed values.

The melting points lie on a smooth curve, without a maximum or minimum, which is very slightly curved towards the composition axis.

4. METALLOGRAPHY

Specimens for microscopical examination were prepared by grinding to 0000 grade emery paper and polishing with diamond dust (0–2μ), using white spirit as a lubricant. With these hard alloys this produced a surface substantially free from scratches and suitable for micro-examination.

The alloys as prepared in the arc furnace are virtually in the chill-cast condition and have a heavily cored solid-solution structure. This is best shown by a differential etch: 10 sec. in a 3% HF solution in concentrated HNO_3, followed by 10 sec. in 10% $K_3Fe(CN)_6$, 10% KOH in water. The first etchant brings up the grain boundaries and the second the cored structure of the grains. This structure is shown in Fig. 2 (*a*) (Plate XXVI), for the 34·8% tantalum alloy as melted in the furnace. Subsequent annealing for 12 days at 1200° C. did not remove the coring (Fig. 2 (*b*)), but by annealing for 30 min. in the arc furnace just below the melting point, homogeneous grains of the solid solution could be obtained (Fig. 2 (*c*)). After this arc-annealing treatment all the alloys had a homogeneous equi-axed grain structure. Hardness measurements were made on all alloys in the annealed and slowly cooled condition. The hardness values lie on a smooth curve reaching a flat maximum at about 500 V.P.N. between 50 and 60% tantalum.

IV.—DISCUSSION

The experimental evidence shows that molybdenum and tantalum form a continuous series of body-centred cubic solid solutions with each other. No intermediate phases or superlattices forming at lower temperatures have been detected.

The alloys of the transition elements of Groups IVA, VA, and VIA with each other present an interesting series. All these elements crystallize in the *A*2 (body-centred cubic) structure; in cases where there is more than one allotropic modification this is the high-temperature form. Hafnium is a possible exception; it is not definitely known whether it transforms to a cubic structure at high temperatures, though this is probably the case.

It is to be expected that a number of alloys of these metals would have continuous solid-solubility ranges, and this has proved to be the case for the binary system molybdenum–tantalum. Some information about other binary alloys that have been studied is given in Table V. Only four ternary systems have been reported on, namely Nb–Ta–Mo, Nb–Ta–W, Nb–Mo–W, and Mo–Ta–W, all of which form continuous solid solutions.[2]

It can be seen from Table V that where the atomic radii differ by 14% or more, continuous solid-solubility ranges are not found, but intermediate phases occur. The system titanium–chromium presents a borderline case, for the size difference is about 14·5% and a continuous series of solid solutions occurs over a very restricted temperature range immediately below the solidus. At lower temperatures an intermediate phase Cr_3Ti_2 is formed. In all the other systems continuous solid solubility exists; with small size differences the deviations of the lattice parameters from the additivity rule (Vegard's law) are negative; for larger size differences they are positive. Superlattices have not been detected in any of the systems.

TABLE V.—*Binary Alloy Systems Between Transition Elements of Groups IVA, VA, and VIA.*

System	Remarks	Size Difference, % *	Deviation from Vegard's Law	Ref. No.
Ti–Cr	Continuous solubility below solidus. Cr_3Ti_2 forms from this solid solution.	14·5	Not known	8
Ti–Zr	Continuous solubility in b.c.c. phase.	9	,, ,,	9
Ti–Nb	,,	0	,, ,,	10
Ti–Mo	,,	5	,, ,,	10
Ti–Hf	,,	...	,, ,,	10
Ti–Ta	,,	0	,, ,,	10
Ti–W	,,	4	,, ,,	10
Cr–Zr	Compound Cr_2Zr formed.	25	...	13
Cr–Nb	Compound Cr_3Nb_2 formed.	14·5	...	15
Cr–Mo	Continuous solid solubility.	9	Positive	11
Cr–Ta	Compound Cr_3Ta_2 formed.	14·5	...	15
Cr–W	Continuous solid solubility.	10	Positive	11
Zr–Mo	Compound $ZrMo_3$ formed.	14·5	...	13
Zr–Hf	Continuous solid solubility probable.	...	...	16
Zr–W	Compound ZrW_2 formed.	14	...	14
Nb–Mo	Continuous solid solubility.	5	Negative	7
Nb–Ta	,, ,, ,,	0	,,	7
Nb–W	,, ,, ,,	4	,,	7
Mo–Ta	,, ,, ,,	5	,,	...
Mo–W	,, ,, ,,	0	,,	12
Ta–W	,, ,, ,,	4	,,	7

* Values for interatomic spacing of β-Ti and β-Zr at room temperature taken as 2·85 and 3·10 Å., respectively.

It may be predicted that continuous solid solutions will be formed in alloys of vanadium and titanium, chromium, niobium, molybdenum, tantalum, or tungsten, and in the systems zirconium–niobium and zirconium–tantalum. In the system vanadium–zirconium an intermetallic compound may be expected.

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PHOTOMICROGRAPHS OF 65·2 AT.-% MOLYBDENUM–34·8 AT.-% TANTALUM ALLOYS.

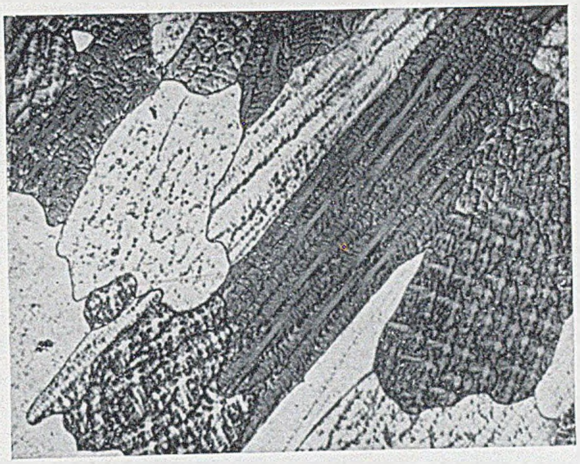

FIG. 2 (*a*).—As Melted. × 30.

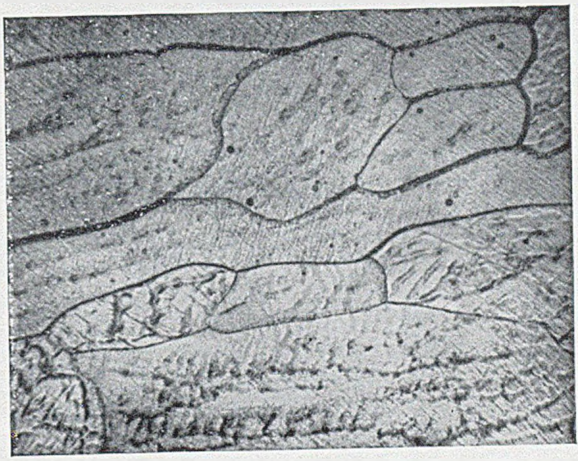

FIG. 2 (*b*).—Annealed 12 Days at 1200° C. × 75.

FIG. 2 (*c*).—Arc-Annealed 30 Min. × 90.

1335

AN INVESTIGATION OF THE STRUCTURAL CHANGES ACCOMPANYING CREEP IN A TIN–ANTIMONY ALLOY *

By W. BETTERIDGE,† Ph.D., F.Inst.P., and A. W. FRANKLIN,† M.Sc., A.I.M., MEMBER

SYNOPSIS

A tin–5% antimony alloy has been used for a microscopical study of the phenomena occurring during creep at room temperature. In addition to general slip within the grains and flow at the grain boundaries, it is shown that localized strain within the grains occurs in directions associated with the boundaries between adjacent grains. This latter plastic flow will be important in any consideration of the quasi-viscous flow of grain boundaries, since it will influence the effective shearing force acting on the flowing boundary. It is also suggested from this work, and from an examination of a pure aluminium sample strained in creep at 250° C., that the " cell structure " observed by X-ray examination is a consequence of the break-up of the grains by the slip bands and by the local strains.

I.—INTRODUCTION

A NUMBER of recent papers have dealt with the mechanism of creep in metals as determined by X-ray and microscopic observation of samples strained at different rates and at different temperatures. These have generally supported the view that, while at lower temperatures and faster rates of straining the creep is largely due to crystallographic slip, at higher temperatures and slower straining rates the deformation is associated with the formation of a sub-structure within the grains. Sub-structures have been observed before, but their importance to the phenomenon of creep has been appreciated only in consequence of the recent X-ray investigations. The sub-structure so revealed has been referred to by Wood and his collaborators [1, 2, 3] as a " cell structure." The relationship between this cell structure and the sub-structure produced by straining and subsequent annealing which is described by the term " polygonization " has been discussed recently.[4, 5]

The much earlier work of Hanson and Wheeler [6] has also shown that at fast strain rates the deformation is largely a consequence of slip-band formation within the grains, but that at slower rates grain-boundary flow is an important feature, although, of necessity, it is accompanied by distortion of the individual grains which occurs without the formation of well-marked slip bands. While grain-boundary flow has clearly occurred to a marked extent in the high-temperature, slow-strain tests of the recent investigators, attention has mainly been focused by them on the mechanism of deformation of the grains, without which appreciable grain-boundary flow cannot take place. Such an outlook is a consequence of the use of X-ray methods which are not sensitive to grain-boundary flow. It is suggested that a better appreciation of the mechanism of creep deformation can be obtained if grain distortion and grain-boundary flow are considered together, although a detailed analysis of the separate processes is, of course, necessary in a complete theory.

Most of the work of earlier investigators has been carried out on pure aluminium. It was thought, however, that a microscopic study of the creep process in a tin–5% antimony alloy might furnish unique evidence which would be of value to a fuller understanding of the mechanism of the deformation process, since this alloy has been shown by Carpenter and Elam [7] to reveal grain-boundary movements on annealing without recourse to repeated etching.

II.—EXPERIMENTAL PROCEDURE

The tin–antimony alloy for this work was prepared by melting under charcoal commercially pure tin and the required amount of antimony to give a 5% antimony alloy. No melting loss was allowed for, and it was assumed that the cast ingot was of satisfactory composition. The metal was cast into a slab 12 × 9 × 1 in. thick, which was machined all over and then cold rolled to approximately $\frac{1}{4}$ in. thick. At this stage, some preliminary experiments were carried out which confirmed that grain-boundary migration could be seen when intermittent annealing of a microsection was carried out.

Flat creep specimens, $\frac{3}{8} \times \frac{3}{16}$ in. with a 2-in. gauge-length, were machined from the rolled strip, rough polished by hand on one flat side, and then annealed for 16 hr. at 190° C. It was established by pre-

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

liminary experiment that this annealing treatment would give a satisfactory grain-size. Specimens were finally prepared by alternate polishing and etching with concentrated ammonium sulphide solution. A simple dead-weight loading system was used for the creep tests, which were carried out at a room temperature of between 20° and 25° C.

III.—EXPERIMENTAL RESULTS

The creep curves obtained by testing at stresses between 712 and 1420 lb./in.^{2} are shown in Fig. 1.

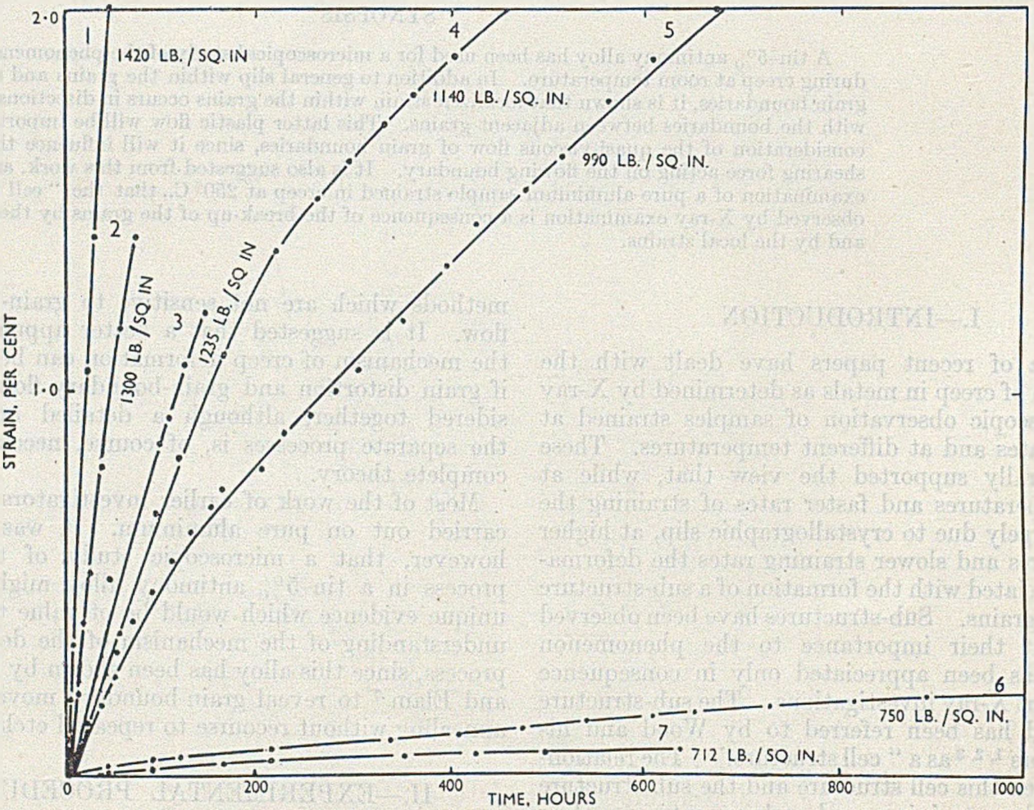

Fig. 1.—Experimental Creep Curves for Tin–5% Antimony Alloy.

The first specimen was etched before the test and strained 3·07% in 66 hr. A typical area after test is illustrated in Fig. 2 (Plate XXVII). Unfortunately, the method of protecting the specimen from dust was not adequate and the detailed structure is obscured, but locally concentrated strain marks and the displacement of scratches at or near grain boundaries could clearly be recognized. A typical example of scratch displacement is shown in Fig. 3 (Plate XXVII).

Further tests were carried out without initial etching of the specimens, it being postulated that if grain-boundary movements occurred then the grains should be revealed without etching.

Although eight specimens in all were tested, particular interest lay in specimens strained at high and low rates. Typical fields of specimens tested at 1300 and 750 lb./in.^{2} (curves (2) and (6) in Fig. 1) are shown in Figs. 4 and 5 (Plate XXVII) respectively, after both had been strained to approximately 0·2% extension, the first in 6 hr. and the second in 1000 hr. It is evident that similar changes have occurred in both specimens, even though there was a 200-fold difference in the rate of extension. The first of these two specimens was further strained to 1·4% in 71 hr. in order to develop the structure, typical examples of which are shown at low and high magnifications in Figs. 6 and 7 (Plate XXVII), respectively, and illustrate the salient characteristics which appeared to greater or less degrees in all specimens. These were:

(i) Rumpling of the surface of grains, varying from relatively sharp parallel lines characteristic of normal slip bands, to irregular markings with only an indication of a general direction. This effect tended to occur most markedly in the smallest grains of a specimen.

(ii) Severely localized strain marks within the grains, usually, but not invariably, running from points of intersection of grain boundaries. (*A* and *B*, Fig. 7, Plate XXVII).

(iii) Thickening of grain boundaries (*C*, Fig. 7).

(iv) Displacement of scratches at grain boundaries (*E*, Fig. 7).

(v) Subsidence of part of a grain (*D*, Fig. 7).

Examination of the first of these features at higher magnifications afforded no further information, as the slopes of the surface were too slight to give adequate contrast, but interesting observations were made on features (ii), (iii), and (iv). The thickening of the grain boundaries and the displacement of scratches were clearly due to grain-boundary flow, the former effect being caused by the component of motion perpendicular to the surface of the specimen and the latter to the component in the plane of the surface. The example shown in Fig. 8 (Plate XXVII) illustrates the apparent widening of grain boundaries by the step formed between the two grains. The escarpments were generally quite steep and, on examination under dark-field illumination, appeared light, as shown in Fig. 9 (Plate XXVII). Under these conditions of illumination the rumples within the grains did not show at all, proving that the slope of these rumples was much less steep than that of the grain-boundary steps.

All specimens were found to show grain-boundary movement to varying degrees, as determined by the amount of displacement of scratches, and the extent of movement differed considerably from boundary to boundary in any particular specimen. It must be remembered that only movements in the plane of the specimen are shown by these scratch displacements. An attempt was made to correlate the amount of boundary movement with the direction of the grain boundaries, as it was expected that the greatest amount of movement would occur on those boundaries lying in the direction of maximum shear stress, but although an indication of a linear relationship between the resolved shear stress and the displacement was obtained, the results were extremely scattered.

Such grain-boundary movements as are demonstrated by the above observations must produce corresponding modifications of the stress in the adjacent grains. Starr [8] has examined the case of a plane crack in an isotropic elastic medium and has shown that, if a uniform shear stress is applied in the plane of the crack, high shear-stress concentrations are produced in the direction of prolongation of the crack. If grain boundaries are regarded as cracks filled with viscous material, then the same considerations will apply to the equilibrium condition when grain-boundary flow has ceased owing to the elastic restraint of the surrounding matrix. If now the matrix is not perfectly elastic and the shear-stress concentrations reach a high enough level, grain slip will occur. When the grain boundary lies in a plane of maximum applied shear stress, the slip in the neighbouring grain will be exactly in the direction of prolongation of the boundary, but even if the boundary lies at an angle to the plane of maximum applied shear stress, the direction of slip in the neighbouring grain will deviate only slightly from the direction of the boundary. These conditions are illustrated diagrammatically in Fig. 15. The initial condition of three grains is shown in (*a*); after flow has occurred between grains *A* and *B* in the direction indicated in (*b*), stress concentration leading to shear along the dotted line in grain *C* would occur. Such shear would, however, not be expected to be observed microscopically except by the displacement of scratches, since there would be no difference in level between the parts of grain *C* on either side of the line of shear. In Fig. 15 (*c*), however, grains *A* and *B* are supposed to move relatively to one another in a direction perpendicular to the diagram, so that *A* is higher than *B*; corresponding displacements of the two parts of grain *C* will occur, so that a visible step will be produced as a prolongation of the boundary between the grains *A* and *B*. This is put forward as an explanation of the many markings of type (ii) which were observed in all specimens. Specimens were closely examined in a search for such a mark which could be shown unambiguously to be due to the mechanism described, and Fig. 10 (*a*) (Plate XXVIII) provides the clearest example. The scratches crossing boundary *Y* are not displaced, those crossing *X* are only slightly displaced, while those crossing *Z* show a very large displacement, and the mark *A* is essentially a prolongation of this boundary. The mark would not, however, be visible, except for the displacement of scratches as at *B*, if the movement were only in the plane of the surface; but it is clear from the difference in focus on either side of the grain boundary *Z* and of the mark *A*, revealed in the high-magnification photograph Fig. 10 (*b*) (Plate XXVIII), that movement has also occurred perpendicular to the surface.

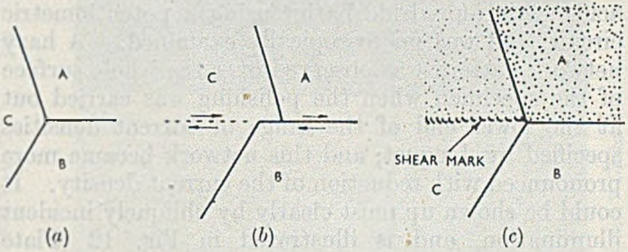

FIG. 15.—Diagrammatic Representation of How Grain-Boundary Movement Between Grains *A* and *B* is Accommodated by Shear in Grain *C*.

(*a*) Before strain.
(*b*) After strain. Grain boundary between *A* and *B* sheared to produce shear in grain *C*.
(*c*) After strain. Same boundary sheared as in (*b*), but in direction normal to figure. Shaded part of *C* rises to produce a shear mark as a prolongation of the grain boundary.

A careful search of all specimens was made for markings within the grains similar to those described as the cell structure by Wilms and Wood (Fig. 2 of their paper [1]). A few areas were found in which such markings were present, and Fig. 11 (Plate XXVIII) illustrates one of these under oblique illumination. In no case, however, was any displacement of scratches observed at the boundaries of such a network, and in view of the more obvious indications of strain of the grains by the presence of the irregular slip bands and the localized shear marks, it was felt that the former markings must play a much less significant part.

In view of these observations on the tin–antimony alloy, it was thought desirable to examine a sample of pure aluminium in order to determine whether the microscopically observable cell structure was more readily developed in that material and whether any relationship between such a structure and other strain markings could be detected. A creep test-bar, ½ in. in dia. of wrought and annealed 99·99% aluminium was prepared with two diametrically opposite flats to facilitate microscopic examination. This bar was electrolytically polished in a standard perchloric acid–acetic anhydride bath,[9] using a potentiometric circuit, and was microscopically examined. A hazy network pattern was observed over the whole surface of the specimen when the polishing was carried out at the lower end of the range of current densities specified by Jacquet, and this network became more pronounced with reduction of the current density. It could be shown up most clearly by obliquely incident illumination, and is illustrated in Fig. 12 (Plate XXVIII). The structure appears very similar to that shown by Wilms and Wood as a typical example of the cell structure produced after slow straining.

The aluminium test-bar was then loaded in a creep-testing machine to a stress of 1000 lb./in.^{2} at a temperature of 250° C., and after 25 hr. it had stretched 9·42%. A further microscopic examination of the surface showed that the structure referred to above was still clearly visible, under oblique illumination (Fig. 13, Plate XXVIII), but only faintly under normal illumination (Fig. 14, Plate XXVIII), when the additional markings due to the creep strain were exactly similar to those in the tin–antimony alloy, viz. rather irregular slip bands within the grains, displaced and thickened grain boundaries, and strain marks within the grains extending from adjacent boundaries. On this aluminium sample it was confirmed by back-reflection X-ray photographs that the initially perfect grains had each broken up into a number of small "cells."

IV.—DISCUSSION

The use of the tin–5% antimony alloy as the experimental material has not resulted in the observation of any unique phenomena, but three different microscopically detectable mechanisms appear to take part in the creep process, viz :

(i) Viscous flow at the grain boundaries.

(ii) General slip within the grains characterized by the formation of slip bands varying from fine straight bands to very irregular coarser bands.

(iii) Localized strain within the grains, often associated with the grain boundaries between neighbouring grains which have flowed viscously.

The first two of these processes are well known and have been described by Hanson and Wheeler,[6] but the importance of the last one does not appear to have been pointed out before. The elastic and plastic properties of the grains which are deforming in this manner will clearly be of importance in controlling the rate of viscous flow of the boundaries. A separate theoretical study of this problem is being made.

The microscopical examination of the unstrained pure aluminium shows that critical etching conditions can develop a structure very similar in appearance to that described as the cell structure by Wilms and Wood,[1] and Lacombe and Beaujard [10] have also shown that such structures can be formed in unstrained metals. True cell formation is convincingly proved by the X-ray examinations of Wilms and Wood, and since Wood and Rachinger [2] have shown that the number of cells into which a grain divides during creep increases with increasing rate of strain and also with decrease in temperature, while at the same time the number of slip bands increases, it is difficult not to associate the division of a grain into cells with its division by the intersecting slip bands, as in fact was suggested by Wood and Rachinger for the lower-temperature tests. Although, if the slip were perfectly regular, no relative rotation of the different portions into which a grain is divided would result, the irregularity in direction of the slip bands, particularly when the creep-rate becomes slower, suggests that some relative rotation must take place. In the creep tests at the slowest rates, after which according to Wood and Scrutton [3] no evidence of crystallographic slip is visible, subdivision of the grains would still occur by the third process mentioned above, viz. strain along the directions of extension of the boundaries between neighbouring grains. This phenomenon is clearly shown in the work of Wood and Scrutton [3] and of Wilms and Wood [1] for samples in which cell formation is claimed without the generation of slip bands.

ACKNOWLEDGEMENTS

The authors wish to thank The Mond Nickel Co., Ltd., for permission to publish this paper. Their thanks are also due to Dr. L. B. Pfeil, O.B.E., F.R.S., for suggesting the investigation.

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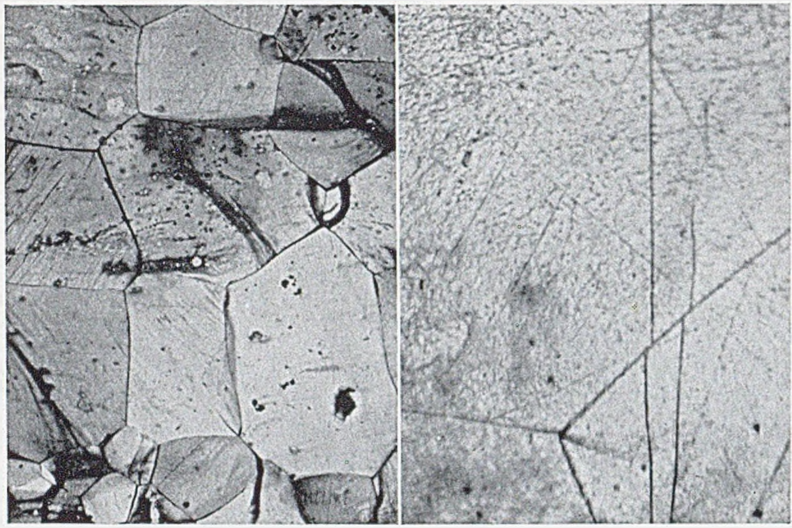
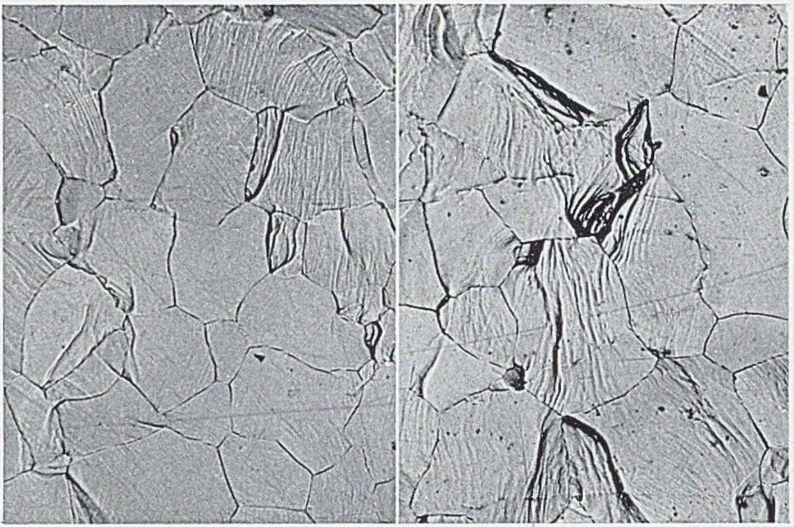
TIN–5% ANTIMONY ALLOY.



Typical Fields of Specimen Strained 3·07% in 66 Hr. at 1420 lb./in.².

FIG. 2.—Etched in Ammonium Sulphide Solution. × 100.

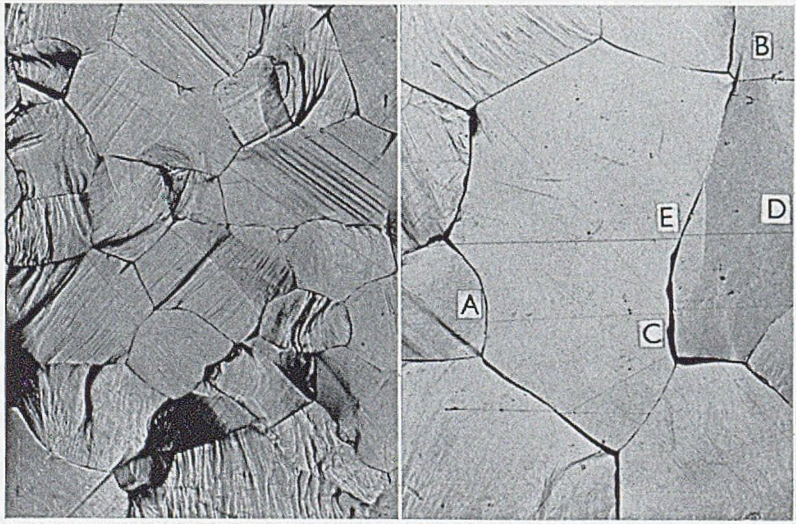
FIG. 3.—Showing Displacement of Scratches at Grain Boundary. × 1000.



Typical Fields of Specimens Strained Approximately 0·2%. × 50.

FIG. 4.—In 6 Hr. at 1300 lb./in.².

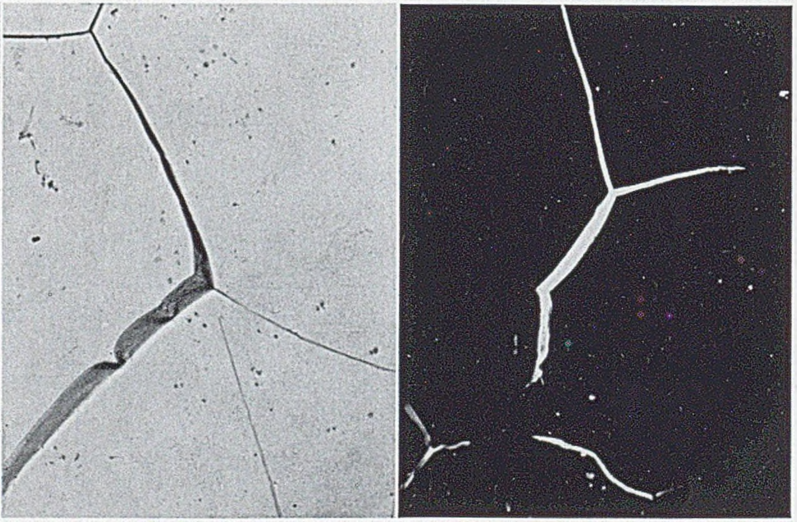
FIG. 5.—In 1000 Hr. at 750 lb./in.².



Specimen in Fig. 4 Further Strained to 1·4% in 71 Hr. at 1300 lb./in.².

FIG. 6.—× 75.

FIG. 7.—× 500.



Appearance of "Thickened" Grain Boundaries Such As Shown at *C* in Fig. 7.

FIG. 8.—Normal Illumination. × 500.

FIG. 9.—Dark-Field Illumination. × 300.

All photomicrographs reduced by $\frac{1}{10}$th linear in reproduction.

PLATE XXVII.

[*To face p.* 150.

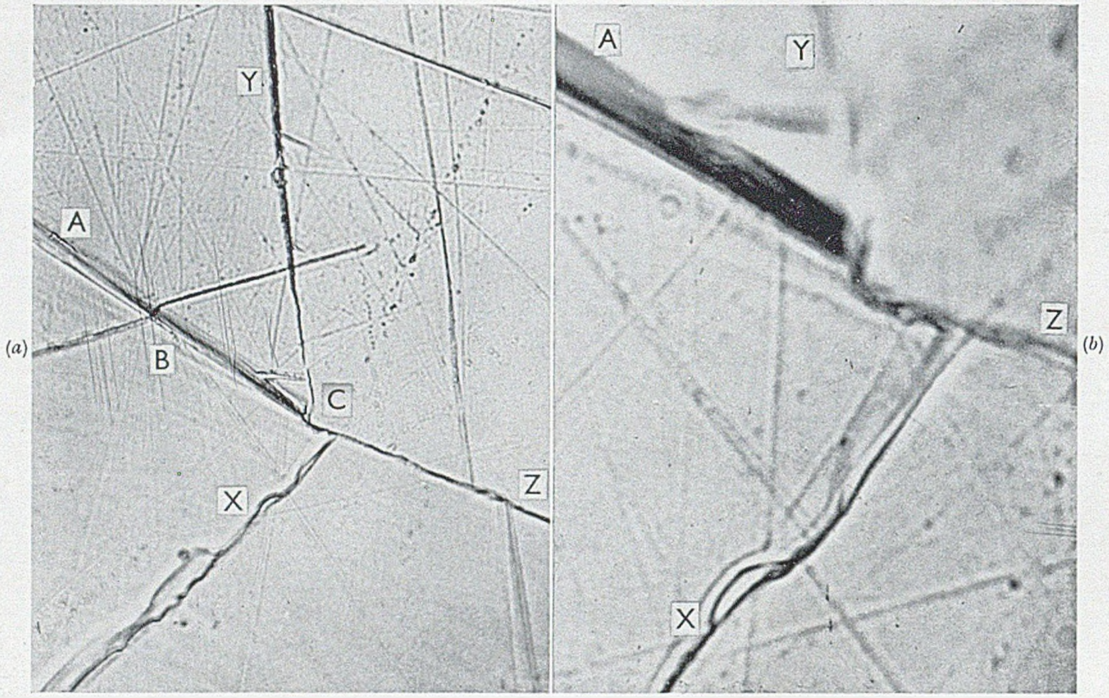


FIG. 10.—Grain-Boundary Movements at Intersection of Three Grain Boundaries at *C* in Tin–5% Antimony Alloy.
(*a*) Shear at *A* arising from movement in grain boundary *Z*. × 1500.
(*b*) Showing that movement perpendicular to the surface has occurred. × 5000.

FIG. 11.

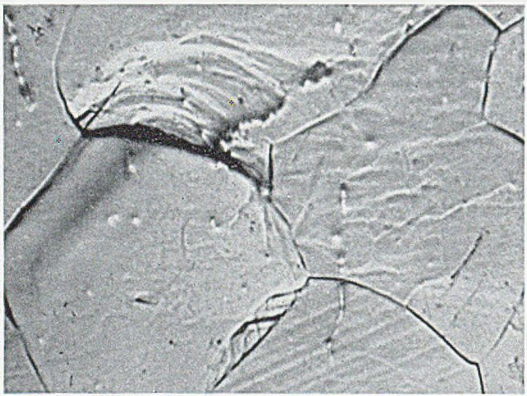


FIG. 12.

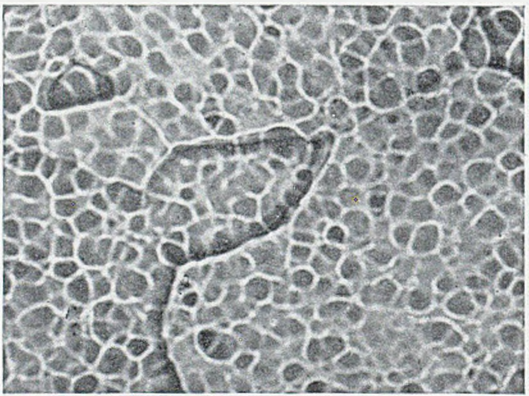


FIG. 13.

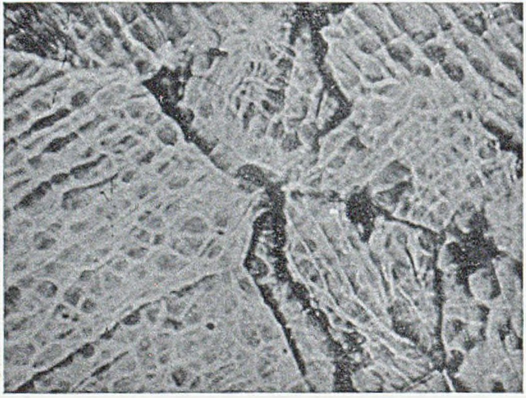


FIG. 14.

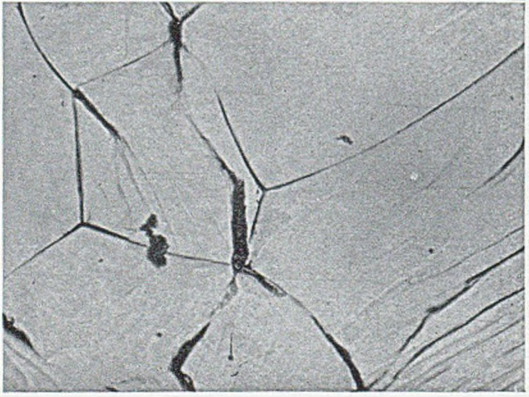


FIG. 11.—Examples of "Cells" in Tin–5% Antimony Alloy Strained 0·42% in 2051 hr. Similar to Those Described by Wilms and Wood in Aluminium. Oblique illumination. × 300.

FIGS. 12–14.—Structure of Electropolished Pure Aluminium (99·99%) Before and After Straining 9·42% at 250° C. in 25 Hr. at 1000 lb./in.^{2}. × 200.

FIG. 12.—Before Creep Test. Oblique illumination.
FIG. 13.—After Creep Test. Oblique illumination.
FIG. 14.—After Creep Test. Normal illumination.

All photomicrographs reduced by $\frac{1}{10}$th linear in reproduction.

METALLURGICAL ABSTRACTS AUTHOR INDEX

METALLURGICAL ABSTRACTS

GENERAL AND NON-FERROUS

VOLUME 19 NOVEMBER 1951 PART 3

1—PROPERTIES OF METALS

***Density Variations in Aluminium.** J. L. Snoek (*Phil. Mag.*, 1950, [vii], **41**, (322), 1188–1192).—The *d* of Al specimens has been measured by a procedure capable of detecting a change of $\sim$1 part in 10^5. Heat-treatment of a sample of high purity except for 0·025 wt.-%. Mg, for 50 hr. at 400° C. decreased the *d* from 2·69885 to 2·69858 g./c.c. Subsequent treatment at 600° C. increased the *d* progressively; after 200 hr. at 600° C. it became 2·69892. No change in the Mg content was found. A second set of treatments of 2–100 min. at 600° C. on the same material without prior treatment at 400° C. showed a correlation of *d* increase with increase of grain-size. Experiments using rapidly chill-cast material of fine-grained structure, with an Mg content of only 0·005 wt.-%, also showed an increase in *d*. Since micrographic examination did not disclose voids in any samples, the possibility is suggested that grain boundaries may be of lower *d* than has previously been supposed. In an addendum, *N. F. Mott* and *F. C. Frank* point out that if these *d* changes are attributed to openness of structure at the grain boundaries, the equivalent void layer in each boundary would have to be 330 Å. thick. The difficulties of explanations based on pptn. of solutes or of gas going into soln. are also pointed out. —P. C. L. P.

[Discussion on a Paper by F. D. Rosi and C. H. Mathewson on :] A Study of the Plastic Behaviour of High-Purity Aluminium Single Crystals at Various Temperatures. —— (*J. Metals*, 1951, **3**, (7), 545–547).—Cf. *Met. Abs.*, 1950–51, **18**, 417.

[Discussion on a Paper by W. R. Opie and N. J. Grant on :] Hydrogen Solubility in Aluminium and Some Aluminium [-Copper and -Silicon] Alloys. —— (*J. Metals*, 1951, **3**, (7), 561).—Cf. *Met. Abs.*, 1950–51, **18**, 481.

***The Preparation and Some Properties of Americium Metal.** Edgar F. Westrum, Jr., and LeRoy Eyring (*U.S. Atomic Energy Commission Publ.*, 1951, (**UCRL-1055**), 13 pp.; and *J. Amer. Chem. Soc.*, 1951, **73**, (7), 3396–3398).—Am metal in the form of 40–200-γ silvery globules obtained by reduction of AmF_3 with Ba in a high-vacuum micro-furnace at 1100° C., was very malleable and ductile as compared with U and Np prepared under similar conditions. The *d* determined for Am (11·7 $\pm$ 0·3 g./c.c.) was low in comparison with the values for the preceding elements, but indicated striking resemblance to the behaviour of its lanthanide analogue, Eu. The heat of soln. of Am in 1·5 *M*-aq. HCl was determined as -160 ± 4 kg.cal./mole. Attempts to determine the m.p. by means of the converging-jaw apparatus described by W. and E. (*ibid.*, p. 3399; *Met. Abs.*, this vol., col. 217) were inconclusive, no sharp convergence of the jaws holding the specimen being noted, but only gradual convergence beginning at $\sim$850° C. and continuing up to 1200° C. without completion.—J. R.

[Discussion on a Paper by E. A. Gulbransen and K. F. Andrew on :] The Kinetics of the Reactions of Beryllium with Oxygen and Nitrogen and the Effect of Oxide and Nitride Films on Its Vapour Pressure. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 257).—Cf. *Met. Abs.*, 1950–51, **18**, 337.—G. T. C.

***Hall Coefficient and Resistivity of Evaporated Bismuth Layers.** W. F. Leverton and A. J. Dekker (*Phys. Rev.*, 1951, [ii], **81**, (1), 156–157).—A letter. The Hall coeff. of films 0·2–1·3 μ thick, evaporated from 99·8% Bi, have been measured at 20° C. under magnetic fields of 1500–12,000 gauss. The temp. coeff. of resistivity deduced from measurements at 77° and at 293° K. was $-3{\cdot}1$ for films 0·2 μ thick, and $-1{\cdot}7$ for 1·3 μ films. If a field was applied $\parallel$ the plane of the sample, at 293° K., the increase in resistivity with magnetic fields up to 12,000 gauss was greater than if the field was $\perp$ the sample. The results suggest that some free electrons are trapped in surface states.—P. C. L. P.

***Reactions of Cæsium in Trace Amounts with Ion-Exchange Resins.** Hoke S. Miller and G. E. Kline (*J. Amer. Chem. Soc.*, 1951, **73**, (6), 2741–2743).—Reactions of Cs salts with ion-exchange resins in concentrations as low as 10^{-8} p.p.m. of active Cs show that phenosulphonic- and sulphonic-type resins are effective for the adsorption of Cs in trace amounts. The phenosulphonic type exhibits two *p*H ranges of strong adsorption, corresponding to the two species of exchange groups present. Cs is selectively adsorbed by phenolic groups in the presence of high concentrations of Na salts in alkaline soln.—J. R.

[Discussion on a Paper by W. J. Kroll, W. F. Hergert, and L. A. Yerkes on :] [Production and Properties of] Ductile Chromium. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 252).—Cf. *Met. Abs.*, 1950–51, **18**, 146.—G. T. C.

***The Sorption of Gases on Metals [Chromium and Steels] at Room Temperature.** A. L. McClellan and Norman Hackerman (*J. Phys. Colloid Chem.*, 1951, **55**, (3), 374–382).—McC. and H. determined isotherms for the sorption of O, Cl, NO, and NO_2 at 30° C. on unreduced and reduced surfaces of S.A.E. 1020 steel, 18 : 8 Cr–Ni steel, and pure Cr powders by the use of a quartz-spiral spring balance. The amount of gas picked up decreased in the following order: S.A.E. 1020, 18 : 8 Cr–Ni steel, Cr. Reduction by H treatment increased the uptake of O and Cl on steel, but made little difference to the other metals. The equivalent number of layers of gas acquired by the metals was calculated from the max. sorption, the measured surface areas, and the molecular dimensions of the gases, assuming an h.c.p. structure for the adhering phase. In only one case (Cl on 18 : 8 Cr–Ni steel) were $>$6 equivalent layers sorbed on the passive metals. A 50 Å. layer formed on unreduced steel in O and a 110 Å. layer on reduced steel. 14 ref.—D. M. D.

***Creep of High-Purity Copper.** William D. Jenkins and Thomas G. Digges (*J. Research Nat. Bur. Stand.*, 1950, **45**, (2), 153–161).—Creep tests in tension were made in air at 110°, 250°, and 300° F. (43°, 120°, and 150° C.) on annealed, O-free, high-conductivity Cu. It was found that the pre-

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

vious thermal and mech. history of the specimen and the rate of loading considerably affected the plastic extension and the creep rate, the general tendency being for both plastic extension and creep rate in the second stage of creep to decrease as the rate of loading increased. A standard rate of loading of 3200 lb./in.^{2}/hr. was used in all tests except those designed to investigate the effect of variations in loading rate. During the second stage of creep the strain rate was found to be not const. but to vary in a cyclic manner, the amplitude of the cycles decreasing with increase in stress or temp., and smaller variations were also observed in the first and third stages. Metallographic examination and hardness measurements were made on specimens before and after testing. In some cases the beginning of the third stage of creep was accompanied by the formation of micro-cracks, but the presence of these cracks was not necessary for tertiary creep to occur. Cracking started near the axis in specimens tested at room temp. and considerable necking took place before rupture, but fracture was predominantly intercryst. in a specimen tested at 250° C. The original grains of Cu were fragmented during creep, and strain markings were observed in all specimens taken to complete rupture.
—J. W. C.

***Relation Between the Energy Levels of the Valence Electrons in Copper and Its Photoelectric Effect: Light Absorption by Thin [Copper] Film.** [—II. 2]. Mitsuru Satô (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], 2, (5), 725–731).—[In English]. Cf. *ibid.*, (3), 424; *Met. Abs.*, this vol., col. 99. In agreement with conclusions found in a previous paper (*loc. cit.*) S. finds that light-absorption max. of thin Cu film confirm values of the energy levels of the valence electrons in Cu. The mechanism of the photoelect. effect at the threshold energies and the selective absorption max. are discussed.
—J. S. G. T.

***Electrical Properties of Thin Germanium Films.** J. M. Dunoyer (*J. Phys. Radium*, 1951, **12**, (5), 602–606).—The elect. resistance of thin films of Ge, $\sim$1 μ thick, prepared by vaporization *in vacuo* at 1200°–1400° C., was studied. The results confirm König's conclusion that thin films of Ge, deposited at ordinary temp., are amorphous, and that when such films are heated at 450°–500° C. they acquire a cryst. structure (*Reichsber. Physik*, 1944). A considerable reduction of the elect. resistance, of the order 10^6 : 1, occurs when the films are heated at $\sim$350° C. On open circuit amorphous, high-resistance Ge films provide a relatively large rectified voltage, but only a small rectified current on closed circuit. Crystallized films, of low ohmic resistance, on the contrary, furnish a very feeble rectified voltage on open circuit, but a relatively large rectified current on short circuit. Composite films comprising a cryst. deposit, of thickness $\sim$1 μ, coated with a very thin amorphous layer, provide a very considerable rectifying effect, comparable with that furnished by a crystal of pure Ge.—J. S. G. T.

***Growth of Germanium Single Crystals Containing *p–n* Junctions.** G. K. Teal, M. Sparks, and E. Buehler (*Phys. Rev.*, 1951, [ii], **81**, (4), 637).—A letter. Single crystals of Ge up to 8 in. long and $1\frac{1}{4}$ in. dia. have been produced by withdrawing a sample from the melt at such a rate that the solidification boundary is at the same level as the surface of the melt. It is simple to make controlled addn. of impurities.—P. C. L. P.

***Theory and Experiment for a Germanium *p–n* Junction.** F. S. Goucher, G. L. Pearson, M. Sparks, G. K. Teal, and W. Shockley (*Phys. Rev.*, 1951, [ii], **81**, (4), 637–638).—A letter. Using the method described by T., S., and Buehler (*ibid.*, p. 637; preceding abstract) a single crystal of Ge has been prepared, part of which was *n*-type and the remainder made *p*-type by a small addn. of Ga. The rectifying properties, photoresponse to light, and other characteristics of this junction were found to agree closely with theoretical expectations.—P. C. L. P.

***Inverse-Voltage Characteristic of a Point Contact on *n*-Type Germanium.** Lloyd P. Hunter (*Phys. Rev.*, 1951, [ii], **81**, (1), 151–152).—A letter. It is pointed out that the curves relating the current through a point contact on *n*-type Ge with the voltage applied across the contact can be constructed semi-quant. if self-heating at the contact is taken into account. Among other assumptions, it is assumed that the real isothermal characteristics are straight lines and that the difference in temp. immediately beneath the contact is directly proportional to the power dissipated.
—P. C. L. P.

***Pulse Measurement of the Inverse-Voltage Characteristic of Germanium Point Contacts.** A. I. Bennett and L. P. Hunter (*Phys. Rev.*, 1951, [ii], **81**, (1), 152).—A letter. In connection with the theory (*ibid.*, p. 151; preceding abstract) it was found that if pulses of $<$0·5 μsec. duration were used to eliminate heating effects, the voltage/current characteristics of a 0·015-in.-thick *n*-type Ge single crystal soldered to a Cu block were approx. straight lines. The pulsed and the D.C. characteristics are compared and discussed theoretically.
—P. C. L. P.

***Significance of Composition of Contact Point in Rectifying Junctions on Germanium.** W. G. Pfann (*Phys. Rev.*, 1951, [ii], **81**, (5), 882).—A letter. A brief survey is given of the effects of the compn. of the metal contact on the rectification characteristics of metal–Ge junctions. New experiments are reported in which the Sb content of the point contact was varied between 0 and 0·01 wt.-%, giving rise to considerable differences in the rectification properties.—P. C. L. P.

***Electron-Bombardment-Induced Conductivity in Germanium Point-Contact Rectifiers.** A. R. Moore and F. Herman (*Phys. Rev.*, 1951, [ii], **81**, (3), 472–473).—A letter. Measurements are reported of the elect. conductivity induced in *n*-type Ge on scanning by a 5–10-kV. electron beam. A kinescope using a synchronously deflected beam, modulated by the induced conductivity, reveals the effects of grain boundaries and other obstacles to the movement of positive holes.—P. C. L. P.

***Measurement of Hole Diffusion in *n*-Type Germanium.** F. S. Goucher (*Phys. Rev.*, 1951, [ii], **81**, (3), 475).—A letter. The potential developed at a fixed probe attached to a single crystal of *n*-type Ge was found to fall exponentially with the distance between the probe and a traversing light spot. The diffusion distance of the positive holes generated by the light was determined from these measurements as 0·028 cm.
—P. C. L. P.

***The Mobility and Life of Injected Holes and Electrons in Germanium.** J. R. Haynes and W. Shockley (*Phys. Rev.*, 1951, [ii], **81**, (5), 835–843).—The mobility of positive holes injected into a single crystal for *n*-type Ge and of electrons into *p*-type Ge has been determined from oscilloscope measurements of the transit times between emitter and collector. Strong elect. fields were also applied to minimize spreading effects due to diffusion. The apparent hole mobility was found to vary with emitter current, but could be readily extrapolated back to zero current. The average mobilities corrected to 300° K. for holes and electrons in Ge were found to be 1700 and 3600 cm.^{2}/V./sec., resp. After making corrections for all the differences that could be readily allowed for, a scatter between the values of $\sim$6% between different samples remained. The mobility of the electrons is approx. 20% higher than that estimated from the Hall effect. This discrepancy may be due to non-spherical electron-energy surfaces.—P. C. L. P.

***Measurement of Semi-Conductor Impurity Content [in Germanium].** Carl N. Klahr and L. P. Hunter (*Phys. Rev.*, 1951, [ii], **81**, (6), 1059–1060).—A letter. A convenient method is given for the calculation of the effective impurity content of Ge from measurements of the conductivity and Hall coeff. at a low and at an intermediate temp. The effective mass of current carriers and the dielect. const. can also be calculated.—P. C. L. P.

***Fundamental Studies of Dissolution of Gold in Cyanide Solutions. IV.—Effects of Water : Solid Ratio, Cyanide Strength, Alkalies, Lead Acetate, and Some Impurities Usually Contained in Foul Cyanide Solutions on Dissolution Rates of Gold and Silver in Cyanidation Tests for Gold Ores.** Mitsuo Kameda (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 787–800).—[In English]. Cf. *ibid.*, 1949, [A], **1**, 435; *Met. Abs.*, 1950–51, 18, 642. Cyanidation tests were made to evaluate the effects of the variables influencing the extraction of Au and Ag from ores derived from 3 Japanese mines, and the results are discussed and compared with results obtained by using pure Au plate. Soln. of Au increases with a rise in the vol. of soln. within the range of water : solid ratio (W/S) 1 : 1—10 : 1, and shows no appreciable increase between $W/S = 10 : 1$ and 20 : 1. In the case of ores contg. much argentite, the rate of dissoln. of Au increases rapidly as the strength of cyanide soln. rises to ~0·03% and then increases slightly as the strength increases to 0·30%. The rate of dissoln. of Au, in the case of one ore, decreases sensibly at higher lime concentration, while it increases markedly in the case of the other two ores. The difference is attributed to the presence of Pb in the first ore, and Fe in the latter two. The rate of dissoln. of Au, in all 3 ores, tends to increase rapidly with increase of Pb acetate; the rate attains a max. value and thereafter falls with concentration of Pb acetate. Most favourable conditions for the ultimate % extraction of Au and Ag with 0·08% cyanide soln. are specified. $K_4Fe(CN)_6$, zincocyanide, thiocyanate, sulphate, and double Cu cyanide have little effect on the rate of Au extraction, whereas xanthate and Na_2S have a retarding effect.—J. S. G. T.

***Adhesion of Indium Under Static and Dynamic Loading.** A. W. Crook and W. Hirst (*Research*, 1950, 3, (9), 432–433).—A letter. The behaviour of cylinders of In when brought into static and into impact contact was studied, a piezoelect. measurement technique being used. The results showed that for equal areas of contact the adhesive force and the welded area were less under dynamic than under static conditions. This result confirms a suggestion that the strength of the metallic weld depends on the time of contact: this in turn is probably related to the variation of the coeff. of friction and of mech. wear with speed. —F. A. F.

***Thermoelectromotive Force of a Supraconductor versus the Same Metal in the Non-Supraconductive State [Lead, Tin, Indium, Thallium].** M. C. Steele (*Phys. Rev.*, 1951, [ii], **81**, (2), 262–267).—Using a D.C. amplifier capable of recording 10^{-9} V., measurements are reported of the thermal e.m.f. from a single supraconducting/normal junction in Pb, Sn, In, and Tl wires. The results are presented as graphs showing the thermal e.m.f. as a function of applied magnetic field and as a function of junction temp. Although the results for In agree approx. with those reported by Keesom and Matthijs (*Physica*, 1938, 5, 437; *Met. Abs.* 1938, 5, 769), who used a different method, those for Sn do not agree. Very marked differences were also observed between the effects for Pb of 99·98 and 98% purities. The significance of the results is discussed. They are in agreement with free-electron theory, provided it is assumed that the Thomson coeff. of the supraconductor is zero and not the value consistent with sp.-heat investigations. 17 ref. —P. C. L. P.

***Ignition Temperature of Magnesium and Magnesium Alloys.** W. Martin Fassell, Jr., Leonard B. Gulbransen, John R. Lewis, and J. Hugh Hamilton (*J. Metals*, 1951, **3**, (7), 522–528).—A report on the development and use of a simple reproducible method of determining the ignition temp. (defined as the "explosion temp."—the temp. at which the rate of oxidation becomes extremely rapid, and a "flame" appears) of Mg and Mg alloys. *Pure Mg.* (i) In O the temp. falls from 644·5° C. at 0·166 atm. to 623° C. at 1 atm., rises steadily to 645° C. at 6 atm. and thereafter remains approx. const. (ii) In O–N mixtures the temp. falls from ~635° C. at 10·5% N to 627° C. at 70% N, and thereafter rises slightly. (iii) In O–SO_2 mixtures the ignition temp. rises from 623° C. at 0% SO_2 to 648° C. at 3% SO_2. *Mg Alloys.* The alloys studied were those in the binary systems with Al, Ag, Bi, Ca, Cd, Co, Cu, In, Mn, Ni, Pb, Sb, Si, Sn, and Zn, up to approx. 60, 20, 20, 5, 90, 5, 30, 30, 5, 25, 60, 30, 5, 25, and 90 wt.-%, resp., as well as a number of ternary and quaternary alloys of sp. compn. In general, the ignition temp. of Mg is lowered by the addn. of alloying elements, and may occur either above or below the liquidus line. It is pointed out that since the ignition reaction of Mg is dependent on the O pressure, any theory of ignition must satisfy this fact. 15 ref.—E. N.

***Low-Temperature Resistance Minimum in Magnesium Measured by a Mutual-Inductance Method.** H. E. Rorschach and Melvin A. Herlin (*Phys. Rev.*, 1951, [ii], **81**, (3), 467).—A letter. A brief description is given of a method of investigating the elect. resistance of a sample using measurements of the mutual inductance of two co-axial coils surrounding it. The method, which eliminates the need to use wire samples and for elect. contacts to the sample, was used to determine the resistance of a Mg sample between 2° and 36° K. It shows a min. at approx. 15° K.—P. C. L. P.

***The Solubility of Mercury in Liquid Phosphorus.** G. J. Rotariu, Eva Schramke, T. S. Gilman, and J. H. Hildebrand (*J. Amer. Chem. Soc.*, 1951, **73**, (6), 2527–2528).—The reversible dissoln. of Hg in liq. P without chem. interaction, the only known case of a metal dissolving by a phys. process in a non-metallic liquid, has been studied. The liq. soln. is perfectly clear, but on rapid crystn. induced by supercooling the Hg separates in numerous very fine globules, giving the solid a grey-black appearance. The solubility of Hg, as measured by the loss in weight after equilibration with liq. P, was 0·29 mg./g. P_4, equivalent to $3{\cdot}7 \times 10^{-5}$ c.c./c.c., which is $\frac{1}{15}$th the value calculated from the regular soln. theory with no allowance for the metallic nature of Hg.—J. R.

***The Melting Point and the Density of Neptunium Metal: A Micro-Melting-Point Apparatus for Metals.** (Westrum and Eyring). See col. 217.

***Supraconductivity of Niobium.** L. C. Jackson and H. Preston-Thomas (*Phil. Mag.*, 1950, [vii], **41**, (323), 1284–1287).—A letter. The supraconducting characteristics of 3 samples of Nb contg. different amounts of impurities have been investigated by: (i) resistance measurements; (ii) an A.C. induction method; and (iii) a ballistic method. Explanations of the differences found between the 3 samples are suggested, and the results are compared with those reported by other workers.—P. C. L. P.

Electronic Properties and Band Structure of Palladium and Platinum. E. P. Wohlfarth (*Proc. Leeds Phil. Soc.*, 1948, 5, (2), 89–101).—The parameters specifying the electronic band structure of metallic Pd and Pt are estimated by a critical analysis on the basis of the collective electron theoretical treatment, and magnetic and thermal properties. The parameters estimated are: (1) q, the number of holes/atom in the d band of the metals; (2) θ_F, the Fermi–Dirac characteristic temp.; and (3) θ', a measure of the interchange interaction energy. The experimental results considered included those on: (*a*) alloys of Pd and Pt with noble metals and of Pd with H; (*b*) low-temp. electronic heats and susceptibilities; (*c*) the variation of susceptibility with temp.; and (*d*) the magnetic properties of alloys with Ni. These results are summarized in tabular form; the values for Pd are found to be in satisfactory agreement, but the values for Pt are less certain. 33 ref.—W. A. M. P.

On the Melting Time of [Silver Wire] Fuses. E. Wintergerst (*Z. angew. Physik*, 1950, 2, 167–174; *Elect. Eng. Abs.*, 1950, 53, 459).—A comprehensive theoretical discussion of the dependence of the melting time, t, on the heat loss from, and phys. properties of, a fuse wire. The various methods of heat removal for a wire in a gas or embedded in a solid are separately considered, and approx. values are derived in the general case for the limiting current (i.e. the current for which $t \to \infty$) and for t. A Ag wire is considered in particular, and the theoretical and experimental values agree well.

***An Interferometric Study of Some Optical Properties of Evaporated Silver Films.** R. C. Faust (*Phil. Mag.*, 1950, [vii], **41**, (323), 1238–1254).—Ag films having reflectivities of 4–95%, were evaporated on to glass opt. flats. Within a few hr. of deposition, measurements were made of the air/Ag and glass/Ag transmission coeff. for different film thicknesses and wave-lengths, and of the relative phase changes. The results are compared with those obtained by other workers using sputtered films. The possibility that the films are not homogeneous layers of well-defined thickness is pointed out, and brief consideration is given to the alternative view that the films really consist of globules whose opt. const. are those of the bulk metal.—P. C. L. P.

***The Mechanism of the Reaction between Silver and Sulphur in Mineral Oil.** R. T. Foley, W. Morrill, and S. J. Winslow (*J. Phys. Colloid Chem.*, 1950, **54**, (9), 1281–1292).—F., M., and W. have investigated the relevant diffusion and adsorption processes, i.e. diffusion of S in the oil, adsorption at the interface, and diffusion in the sulphide layer. Strips of Ag, after suspension in the S-contg. mineral oil at 60° C. were removed at intervals, washed, and weighed, until 95% of the S had reacted. Reasonable agreement was found between the sulphiding rate and the calculated rate of diffusion in the oil. A comparison of the rate of sulphiding in five solvents suggested that diffusion is not the rate-determining factor in solvents of low viscosity ($\eta_{60^\circ\text{ C.}} < 0{\cdot}01$ poises). The effect of temp. was found to conform with the effect expected from the change in viscosity. Inhibitors slowed down the rate of reaction but did not prevent ultimate sulphidation. The observed rate of reaction was slower than that calculated if diffusion through the sulphide layer were the rate-determining factor. Thus, for viscous mineral oil, the diffusion of S through the oil normally controls the reaction. 15 ref.—D. M. D.

***Precipitate-Ion Exchange Reactions Using Radiosilver.** George K. Schweitzer and James W. Nehls (*J. Phys. Colloid Chem.*, 1951, **55**, (2), 215–221).—S. and N. have investigated the exchange of Ag between $AgNO_3$ soln., labelled with radio-Ag, and AgCl, AgBr, AgI, and AgCN precipitates, resp., at 30°, 40°, 50°, and 60° C. The inactive precipitate was introduced into the stirred soln. of labelled $AgNO_3$ and samples were removed through a fritted glass disc at intervals extending over 2 hr. The samples were dried and counted using a Geiger–Müller counter. From the results the time (t) for 60% exchange was obtained and a plot made of log $1/t$ versus $1/T$; the slope of this line gave $-E/2{\cdot}303R$, where E is the activation energy for the movement of Ag ions through the crystal lattice. Values of E (kg.cal./mole) were found to be: AgCl 17·6, AgCN 16·0, AgBr 15·0, AgI 11·8. 11 ref. —D. M. D.

***New Aspects of the Photoelectric Emission from Sodium and Potassium.** Jean Dickey (*Phys. Rev.*, 1951, [ii], **81**, (4), 612–616).—The photoelectric emission from thick evaporated layers of Na and K has been investigated for incident photon energies $h\nu$ up to 6·7 eV. With $h\nu < 4$ eV., the energy distribution of the emitted electrons agrees with that expected of an ideal metal. With $h\nu > 4$ eV. the proportion of low-velocity electrons steadily increased. Three possible explanations of this phenomenon are discussed: (i) loss of energy because of scattering by other electrons; (ii) another electron absorbing the recoil momentum instead of the crystal as a whole, so that the photoelectron would absorb only part of the energy of the incident photon; and (iii) modification of the distribution of the initial energy states to give more low-energy states. The spectral distribution of the photoelectric yield was found to be sensitive to the smoothness of the deposit, which could be varied by altering the rate of deposition.—P. C. L. P.

***Calculations of Surface Energy for a Free-Electron Metal. [Sodium, Copper].** H. B. Huntington (*Phys. Rev.*, 1951, [ii], **81**, (6), 1035–1039).—Previous treatments (e.g. Huang and Wyllie, *Proc. Phys. Soc.*, 1949, [A], **62**, 180; *Met. Abs.*, 1949–50, **17**, 637) of the surface energy of a metal have assumed a barrier of a height determined by the cohesive energy located at the geometrical surface of the metal. The surface tension, calculated from this surface energy, was found to be greater than that observed experimentally. H. points out that the assumptions made do not allow the surface to remain elect. neutral. New values of the surface energy of Na and of Cu are calculated using a sq.-cut barrier placed so that there is no net charge on the surface. For Na, calculations are also made using self-consistent soln. of an image type of barrier. For Cu, the calculated surface tension is 500 dynes/cm., in comparison with the observed value of 1100 dynes/cm. in the liq. state. For Na, the sq.-cut barrier leads to a surface tension of 136 dynes/cm., and the self-consistent barrier to 54 dynes/cm., whereas the observed value for liq. Na is 290 dynes/cm. An attempt is made to explain the discrepancies, but most of the factors that have been ignored in the theory would tend to decrease the calculated surface tensions.—P. C. L. P.

***Periodic Deviations in the Schottky Effect for Polished Tantalum.** G. B. Finn, W. B. LaBerge, and E. A. Coomes (*Phys. Rev.*, 1951, [ii], **81**, (5), 889).—Cf. *ibid.*, 1950, [ii], **80**, 887; *Met. Abs.*, this vol., col. 103. A letter. The periodic deviations with potential in the emission from unpolished Ta wire (0·001 in. in dia.) at 1500° K., agree with the results from polished 0·005-in.-dia. wire, in the range of fields investigated.—P. C. L. P.

***Thermoelectric Properties of Titanium, with Special Reference to the Allotropic Transformation.** H. W. Worner (*Australian J. Sci. Research*, 1951, [A], **4**, (1), 62–83).—The direct e.m.f./temp. relation for Ti/Pt thermocouples, and their thermoelect. power, using a differential arrangement, were determined. A special differential thermocouple and ratio potentiometer were developed to measure the thermoelect. power; data so obtained were used to calculate the e.m.f./temp. relationships, and also the temp. dependency of the Peltier coeff., the difference between the Thomson coeff. of Ti and Pt, and the abs. thermoelect. power of Ti. Abrupt changes in thermoelect. power accompany the α–β transformation, the transition temp. range being 883° ± 1°–887° ± 2° C. Gradual decrease in sharpness of changes is ascribed to O and N absorption, even though a vacuum of 10^{-4} mm. Hg was maintained. Commercial grades of Ti remain appreciably harder (230 D.P.N. compared with 70), and the transformation range is 865°–970° C. After cold work, Ti softens on heating to 190° C. for 1 hr., and recrystn. is nearly complete after 1 hr. at 860° C. Preferred orientation affects e.m.f. values greatly, but it is almost eliminated at 890° C. (just above the α–β transformation temp.). X-ray diffraction confirms this.—S. M.

Vanadium Now Available Commercially. —— (*Steel*, 1950, **127**, (14), 108).—A brief review of the phys. and chem. properties of 99·8% pure V.—E. J.

[Discussion on a Paper by E. A. Gulbransen and K. F. Andrew on:] The Kinetics of the Reactions of Vanadium with Oxygen and Nitrogen. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 257).—*Met. Abs.*, 1950–51, **18**, 344.—G. T. C.

Zirconium Offers Interesting Properties and Possibilities. —— (*Materials and Methods*, 1950, **32**, (4), 63–64).—A brief review of the properties and appn. of ductile Zr.—J. W. C.

[Discussion on a Paper by E. T. Hayes, A. H. Roberson, and R. H. Robertson on:] Heat-Resistance of Zirconium in Several Mediums. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 254).—Cf. *Met. Abs.*, 1950–51, **18**, 247.—G. T. C.

***The Permeability of Zirconium to Hydrogen.** Richard B. Bernstein and Daniel Cubicciotti (*J. Phys. Colloid Chem.*, 1951, **55**, (2), 238–248).—B. and C. have measured the permeability of Zr to H, using a H purification train, a stainless-steel holder for the Zr sheet disc, and a vacuum-type collection burette. Purified H was passed through a liq. N trap and up to the Zr disc which was sealed between two flanged tubes, surrounded by N, in a furnace maintained at temp. up to 920° C. The diffusing H was collected and measured in the evacuated system downstream of the Zr disc. The flow rate at const. temp. and pressure was found to decrease with time and to be sensitive to the surface condition of the specimen. Values of the flow (E) at different temp. were corrected to the same time and $\log_{10}$ flow rates, for this time, were plotted against $1/T$. The lowest values for E appeared to be 65 and 85 kg.cal./mole, resp., for two sets of experiments. It was shown that, at a selected ref. time, the flow rate varied with the sq. root of the pressure. The effect of sample thickness could not be evaluated because of the variation in flow rate observed between samples of similar thickness. B. and C. suggest that the time dependence of the flow rate is due partly to a change of structure in the Zr caused by the soln. of H and partly to the formation of an oxide film. 12 ref.—D. M. D.

***Action of Hot Ionized Gases Upon Zirconium and Copper.** Andrew Dravnieks (*J. Phys. Colloid Chem.*, 1951, **55**, (4), 540–549).—D. found that, in gas-metal reactions, the rate effects caused by ionization of the gas were most pronounced with Zr and Cu. A method of measuring the change in elect. resistance, and thereby the reaction rate, of strips of Zr at 986° C. and of Cu at 750° C. in a flow of low-pressure gas is described. The reaction rates were found for Zr in non-ionized and ionized O, air, H_2O, CO_2, CO, N + 0·5% O, C_2H_6, and technical He. In some cases ionization caused acceleration and in others retardation, the change in rate never varying by more than a fraction of 2. The oxidation of Zr in water vapour was faster than in N + 0·5% O and much faster than in CO_2. Cu was slowly oxidized by ionized CO_2, but was not oxidized in N + 0·5% O or CO, whether ionized or not. 31 ref.—D. M. D.

Why Machine Parts Fail. V.—Bending and Tensile Fractures. VI.—Torsional Fractures. Charles Lipson (*Machine Design*, 1950, **22**, (9), 147–150; (10), 97–100).—Cf. *ibid.*, (8), 157; *Met. Abs.*, 1950–51, **18**, 755. [V.—] This part is intended to show how to recognize failures resulting from bending and axial loads. Fatigue failures in bending are discussed with ref. to the particular modes of formation, e.g. the various types of stress concentration. L. discusses the sub-surface origin of cracks and presents a brief note on tensile and compressive fatigue fractures. [VI.—] L. describes and illustrates the two main types of torsional failure and indicates that fractures of different appearance may be readily explained if the local stress distribution and properties of the material are considered. The characteristic features of the fractures obtained in various types of components are illustrated and discussed. The influence of surface finish on torsional failure is outlined. Splined shafts and hollow shafts produce fractures which give indications of the stress distributions in those parts. Crankshafts are examples of components subjected to combined torsion and bending.—D. M. L.

†**The Fracture Stress of Metals as Affected by Plastic Deformation, the Stress System, and Size.** D. J. McAdam, Jr. (*Cold Working of Metals* (*Amer. Soc. Metals*), **1949**, 321–364).—Cf. McA., Geil, Woodard, and Jenkins, *Trans. Amer. Inst. Min. Met. Eng.*, 1949, **180, 363**; see *Met. Abs.*, 1948–49, **16**, 120. A review of work in this field by McA. and his co-workers and by others, mostly carried out on steels.—N. B. V.

Limiting Conditions for Brittle Failure and Plastic Behaviour of Ductile Metals. C. Torre (*Österr. Ing.-Arch.*, 1950, **4**, (2), 174–189; *Appl. Mechanics Rev.*, 1951, **4**, 164).—T. accepts a paraboloid of rotation expressed by: $(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1) + 2(c - 1)(\sigma_1 + \sigma_2 + \sigma_3) = 2c\sigma_z^2$ as the boundary surface for the initiation of plastic deformation. The ratio c of the stress at uniaxial compression (σ_d) and stress at uniaxial tension (σ_z) at equal strains is supposed to be >1 and <2–3, except at the initial yield stress, where $c = 1$. Hence c is a function of strain and also is different for different metals. This condition, which is applied to several cases of loading such as torsion, axial tension, and compression with balanced transverse compression, plane tension, and triaxial tension in connection with the brittle fracture of welds, is regarded as being in reasonable agreement with experiments.

***A Theory of the Plastic Bulging of a Metal Diaphragm by Lateral Pressure.** R. Hill (*Phil. Mag.*, 1950, [vii], **41**, (322), 1133–1142).—A treatment is given of the behaviour of a metal diaphragm clamped over a die with a circular aperture, when subjected to an increasing pressure from one side which causes it to bulge through the aperture. The predicted effect of previous cold working agrees with experiment. The theory gives a relatively simple expression for the instability strain, when the bulge will go on increasing under falling pressure.—P. C. L. P.

Fatigue in Metals: A Critical Survey of Recent Research and Theories. Paul Feltham (*Iron and Steel*, 1948, **21**, (11), 431–436).—80 ref.—N. B. V.

†**Cold Work and Fatigue.** Alfred M. Freudenthal (*Cold Working of Metals* (*Amer. Soc. Metals*), **1949**, 248–261).—The mechanism of fatigue failure in a polycryst. metal (in which the grain boundaries act as a comparatively rigid skeleton and provide the principal resistance to deformation) is discussed generally. Previously-applied cold work usually increases the endurance limit, but may be expected to reduce the fatigue life at higher stresses. Brief ref. is made to the beneficial effects of cold working the surface of parts subjected to fatigue stressing.—J. C. C.

***Analogy Between the Fracture of Methyl Methacrylate and That of Metals by Fatigue.** Felix Zandman (*Compt. rend.*, 1951, **233**, (1), 20–21).—Z. analyses the origin and propagation of fractures in specimens of methyl methacrylate submitted to simple stresses, and suggests that the mechanism applies equally to simple fractures of metals.—J. H. W.

***Study of the Coefficient of Internal Friction of Materials Used in Civil-Engineering and Architectural Structures.** Daisuke Shimozuru (*Bull. Earthquake Research Inst. Tokyo Univ.*, 1949, **27**, (1/4), 85–89; *Appl. Mechanics Rev.*, 1951, **4**, 167).—Damping in solid cylindrical specimens of steel, Al, ebonite, and wood was measured by decay of free torsional oscillations. Vibrations were initiated by an electromagnet, the specimen, with a disc fixed at each end, being suspended by a fine wire in a vacuum to reduce external losses. The relative motion of the ends of the specimen was opt. recorded. The curves that are given of logarithmic decrement against air pressure have no numerical ordinates, so that the effectiveness of the design in reducing external losses cannot be judged. No information as to stresses, phys. properties, and condition of the specimens is given, so that results cannot be compared with other work.

Fluidity of Metals: A Review of Test Methods and Results. J. E. Worthington (*Iron and Steel*, 1949, **22**, (14), 615–619; 1950, **23**, (1), 9–11).—32 ref.—N. B. V.

***The Flow of Liquid Metals on Solid Metal Surfaces and Its Relation to Soldering, Brazing, and Hot-Dip Coating.** G. L. J. Bailey and H. C. Watkins (*J. Inst. Metals*, 1951–52, **80**, (2), 57–76).—Earlier investigations of soldering performance are reviewed and, after a theoretical discussion of factors expected to control the flow of liquids on solid surfaces, experimental investigations of the behaviour of several liq. metals and alloys in contact with various solid metals, are described. Most of the work was carried out using H as flux, but a few experiments with liq. fluxes are also described. It is found that the contact angle between solid and liq. surfaces of the metals examined is not, in general, zero. Its true value in given circumstances is obscured by the effects of roughness of the solid surface, which cannot be allowed for. Roughness effects could also account for the fact that the contact angle against a surface over which liquid has receded is generally lower than that against a surface over which liquid has advanced. The development of the particularly low contact angles which are formed against Cu by Sn–Pb alloys with compn. in the range preferred for practical soldering, appears to be preceded by the formation of an alloy layer in the surface of the Cu over which the liq. metal spreads. The alloy layer may be formed relatively slowly by diffusion through the surface layers of the Cu from the bulk of the liq. metal, or it may be formed relatively quickly as a result of transfer of metal ions through suitable liq. fluxes. The prodn. of particularly low contact angles between Pb and Fe or Cu and between Sn and Cu, through very small addn. to the Pb or Sn, is described.
—AUTHORS.

On the Latent Heat of Fusion and the Hole Theory. N. R. Mukherjee (*J. Chem. Physics*, 1951, **19**, (4), 502–503).—A letter. Since the temp. of a liquid remains const. during fusion, M. attributes the increase in vol. as compared with the solid state to the formation of additional holes, and postulates that the total energy of this hole formation is equal to the latent heat of fusion. Considerations arising are developed and discussed, and relevant data for a number of substances, including 13 non-ferrous metals, are tabulated.—J. R.

Thermal Expansion at Low Temperatures. S. Visvanathan (*Phys. Rev.*, 1951, [ii], **81**, (4), 626–627).—A letter. A modified Grüneisen relationship between the thermal expansion and the atomic heat of an isotropic solid is derived, taking into account the temp. dependence of the kinetic energy of the conduction electrons. The results imply that the thermal expansion of metals should decrease more slowly at liq.-He temp. than at higher temp. For the transitional metals, which have a large electronic sp. heat, the effects should become noticeable at 15° K.—P. C. L. P.

Active Conditions and Reaction Characteristics in the Solid State. R. Fricke (*IVA*, 1948, **19**, (5), 192–195).—[In German]. Developments in our knowledge of the detailed structural characteristics of solids, consequent upon the discovery of X-ray diffraction, by von Laue, Friedrich, and Knipping, and their significance in the study of reactions between solids, are briefly reviewed. Solids having " active " centres react at much lower temp. than inactive solids; powders characterized by disturbed lattices react more readily than powders having undisturbed lattices, the fineness of powdering being equal in both cases. Surface structural changes as well as surface changes associated with allotropy and atomic disorder produce changes of both phys. and chem. properties of solids. 34 ref.—J. S. G. T.

The Production, Properties, and Uses of Thin [Metallic] Films Condensed *in vacuo*. L. Holland (*Vacuum*, 1951, **1**, (1), 23–35).—The technique of producing thin films by evaporation and condensation *in vacuo* is briefly discussed, special attention being given to: (1) evaporation pressure, (2) pumping speed and ultimate oil vapour pressures, and (3) the use of discharge techniques for cleaning contaminated supports. The colour of Al films, by transmitted light, can change from bluish-grey to neutral, indicating formation of oxide; films with high refractive index and low absorption have been formed on glass by oxidizing condensed metal films and by cathodic sputtering in an atmosphere contg. O. The reflectivity of front-surface reflectors of Al is briefly referred to, and the elect. properties and cryst. structures of thin films, a theory of condensation and four principal factors determining the structures and properties of thin films deposited *in vacuo*, are interestingly reviewed. 12 ref.
—J. S. G. T.

***Vapour Pressure and Rate of Evaporation.** I. N. Stranski and G. Wolff (*Research*, 1951, **4**, (1), 15–24).—The equation for effective evaporation rate is: $\frac{dn}{F\,.\,dt} = \frac{(p_s - p)\,.\,\alpha}{\sqrt{(2\pi mkT)}}$, where dn is the number of molecules evaporating from the area F in time dt, p_s = vapour pressure, p = the actual pressure, α = the condensation coeff., k = Boltzmann's const., m = the molecular weight, and T = the abs. temp. If this equation is applied to the overall evaporation process, it does not sufficiently take into account the kinetic aspects of the process. α may depend on the temp. as well as on the degree of saturation of the vapour phase, and hence the abs. rate of evaporation is not always proportional to the vapour pressure p_s. It may follow that, of two modifications of a substance, that with the higher vapour pressure has the lower evaporation rate. Possible evaporation processes may be represented by two energy-level diagrams, one differing from the other in allowing for a finite excitation energy, ψ, for condensation. Phenomena of impeded evaporation and of impeded condensation are discussed, using As_2O_3 as an example. Enforced phase transition may occur if the initial phase receives a sufficient amount of energy at a level $E \gg \psi$ so that the process can then be represented by the alternative energy level diagram. Because of its high rate of formation, the forcibly obtained phase is usually in a severely distorted state. 21 ref.—F. A. F.

On the Variation of the Electrical Resistance of Very Thin Metallic Deposits with Applied Potential. Boris Vodar and Nicolas Mostovetch (*Compt. rend.*, 1950, **230**, (23), 2008–2010).—A theoretical discussion of results previously reported (*ibid.*, (10), 934; *Met. Abs.*, 1950–51, **18**, 754).—N. B. V.

***Studies of Surface Reactions by Means of Radioactive Tracers.** T. Westermark and L. G. Erwall (*Research*, 1951, **4**, (6), 290–292).—A letter. A study is being made of the sorption of ions on the surfaces of solids; the method consists of bringing a plane surface of the solid into contact with a water soln. of a radioactive substance of high sp. activity, followed by examination of the surface activity by Geiger–Müller counters or by autoradiography. An example of the appn. to a metallic specimen is given, consisting of photomicrographs of a Pb–Sn–Sb–Cu alloy, one before treatment, and the other, an autoradiograph, after treatment for 20 min. with a water soln. of ^{212}Pb. Pb-rich areas of the structure and dark areas within the Sb–Cu crystals sorbed ^{212}Pb. A technique of this kind may be useful in attempting to follow surface reactions of the treated metallic surface.
—F. A. F.

Thermo-Electricity: A Survey of Factors Affecting the Thermo-Electric Power of Metals. D. Hadfield (*Iron and Steel*, 1948, **21**, (12), 478–482).—A review, with 25 ref.—N. B. V.

***Thermionic Constants of Polycrystalline Metal.—I.–II.** Tetsuya Arizumi (*J. Phys. Soc. Japan*, 1948, **3**, 1–7; 186-190; *C. Abs.*, 1949, **43**, 4101; 1950, **44**, 5697).—[In English]. [I.—] Math. discrepancies between experimental and theoretical evaluations of Richardson's const. A in the formula for the thermionic emission current i ($\log i/T^2 = \log A - e\phi/KT$) from a metal are discussed, where ϕ is the work-function of the metal. An explanation based on the variation of the thermionic work-function at different cryst. faces is proposed. Calculations show that in polycryst. metals the average value of A should be lower than for single crystals. Agreement is found in the case of Ba, Th, Ta, Mo, W, Pd, and Pt. Zr and Re, which show higher values of A, are assumed to be impure. The theory is further extended to explain the abnormal emission from thoriated W. [II.—] The detn. of the work-function of metals by the initial-velocity method is considered in detail. The effect of the variation of work-function over the polycryst. surface is discussed. The average work-function is calculated for the case of two different work-functions coexisting in the same sample. Comparison of results obtained by the initial-velocity method with those obtained by Richardson's line method show that the former yields larger results and the two become equal when nearly all the emission is contributed by the region with the higher work-function.

Thermodynamics of Thermomagnetic and Galvanomagnetic Effects. P. Mazur and I. Prigogine (*J. Phys. Radium*, 1951, **12**, (5), 616–620).—A thermodynamic theory of irreversible thermomagnetic and galvanomagnetic phenomena, based upon Onsager's relations (*Phys. Rev.*, 1931, [ii], **38**, 265; **37**, 705) is developed, for the case of an isotropic medium in the absence of a magnetic field. Expressions for 6 transverse and 8 longitudinal effects are derived. These include as transverse effects the isothermal and adiabatic Hall effects, the isothermal and adiabatic Nernst effects, the Ettingshausen, and Righi–Leduc effects. The longitudinal effects include isothermal and adiabatic elect. resistance, isothermal and adiabatic thermal conductivity, the isothermal and adiabatic Ettingshausen–Nernst effects, and the isothermal and adiabatic Nernst effects. Relations between thermomagnetic and galvanomagnetic effects are discussed.
—J. S. G. T.

***Measurement of the Gyromagnetic Constant of Ferromagnetic Elements and of Some Alloys of Equal Electron : Atom Ratio of the Iron Group.** (Meyer). See col. 182.

†Limitations of the Bloch Spin-Wave Treatment of Ferromagnetism. E. P. Wohlfarth (*Proc. Leeds Phil. Soc.*, 1949, **5**, (3), 213–223).—The Bloch spin-wave treatment of ferromagnetism is critically reviewed and discussed. The coeff. of $T^{3/2}$ in Bloch's law of approach to saturation is considered in relation to Fallot's low-temp. experimental data, from which values are derived for J, the exchange integral for neighbouring atoms. These are used to estimate values of the Curie temp. of Ni and Fe, which are found to be higher than the observed values. The derivation of Bloch's relation is summarized, and it is shown that the temp. range of validity of the assumptions underlying Bloch's approx. soln. of the secular equation is much more restricted than is indicated by experimental results, and that Bloch's treatment should be applicable only to ferromagnetics which are completely saturated at abs. zero. An analysis of low-temp. sp.-heat results shows that the spin-wave treatment is incompatible with the experimental results, which are again well correlated by the collective-electron treatment. Values of the numerical coeff. in Bloch's relations are given in an appendix. 30 ref. —W. A. M. P.

***On the Theory of Spin-Waves in Ferromagnetic Media.** Conyers Herring and Charles Kittel (*Phys. Rev.*, 1951, [ii], 81, (5), 869–880).—Arguments are presented which indicate that the spin-wave theory of ferromagnetism is. of more general validity than the model upon which it was founded. The spin-wave theory is developed by treating the ferromagnetic material as a continuous medium in which the *d* of the three spin components are regarded as amplitudes of a quantized vector field. The effect of magnetic fields and of anisotropy is considered, and the theory is applied to insulators and to metals. The $T^{3/2}$ law for the variation of saturation magnetization with temp. is derived in conjunction with the theory of the energy of a domain boundary. Wohlfarth's criticisms of the validity of the spin-wave theory (*Proc. Leeds Phil. Soc.*, 1949, 5, 213; preceding abstract) are thought to be unjustified.—P. C. L. P.

***Thin Ferromagnetic Films.** Martin J. Klein and Robert S. Smith (*Phys. Rev.*, 1951, [ii], 81, (3), 378–380).—The magnetic properties to be expected of thin films of ferromagnetic materials have been calculated from the Bloch spin-wave theory. The spontaneous magnetization decreases rapidly as the number of atomic layers decreases below a critical value, and the temp. dependence changes from a $T^{3/2}$ law with very thick films to a linear relationship for a monolayer film. These conclusions are expected to be valid only provided that the magnetization is near its saturation value. —P. C. L. P.

Supraconductivity. Aaron Wexler (*Research*, 1950, 3, (12), 534–540).—A review is given of the advance of knowledge of the fundamental features of supraconductivity from 1908 to the present day. 57 ref.—F. A. F.

***Influence of Plastic Deformation on the Supraconductivity of Metals.** V. I. Khotkevich and V. R. Golik (*Zhur. Eksper. Teoret. Fiziki*, 1950, 20, (5), 427–437; *Physics Abs.*, 1950, 53, 827).—[In Russian]. The plastic deformation of Sn, In, and Tl at low temp. leads to the formation of a metastable condition, characterized by an increased value of the critical temp. (T_{cr}). The existence of a max. on the curves of T_{cr} against degree of plastic deformation is due to the superposition on this effect of the vol. compression, which reduces T_{cr}. The result leads to the conclusion that possibly the phase transformation $\mathrm{Tl_{II}} \rightarrow \mathrm{Tl_{III}}$, observed by Bridgman, takes place. The effects observed are conditional upon low temp. and do not occur under deformations at higher temp., even though they still might be below recrystn. temp. The plastic deformation of Hg does not give rise to similar effects. The observed lowering of T_{cr} is due to the vol. compression.

Field Variation of Supraconducting Penetration Depth. J. Bardeen (*Phys. Rev.*, 1951, [ii], **81**, (6), 1070–1071).—A letter. It is shown how the apparently contradictory results that: (i) the penetration depth depends only slightly on the applied magnetic field, indicating that the transition unit is $>10^{-4}$ cm.; and (ii) even with thicknesses as small as 5×10^{-6} cm., there is little change in the supraconducting transition temp. with film thickness, both follow from the lattice vibration theory of supraconductivity. The evidence regarding the length of the vibrational waves that interact most strongly with the electrons is discussed briefly. —P. C. L. P.

Perfect Diamagnetism of Free Electrons with Application to Supraconductivity. M. F. M. Osborne (*Phys. Rev.*, 1951, [ii], 81, (1), 147).—A letter. Calculations of the Fermi–Dirac statistics of free charged particles in a uniform magnetic field in a finite box, indicate that almost perfect diamagnetism is to be expected up to a critical field. This behaviour is identified with the almost perfect diamagnetic behaviour of certain supraconductors.—P. C. L. P.

On Heisenberg's Theory of Supraconductivity. Kun Huang (*Nat. Peking Univ. Semi-Cent. Vol., Math. Phys. Biol. Ser.*, **1948**, 89–101; *Physics Abs.*, 1950, **53**, 827).—[In English]. It is pointed out that the wave-packets used by Heisenberg to form the correlation lattice in his recent theory of supraconductivity (*Z. Naturforsch.*, 1947, [A], 2, 185; *Met. Abs.*, 1947–48, **15**, 228) are not mutually orthogonal; the method used for obtaining their *d* is accordingly arbitrary. In particular, this invalidated the result that the total energy of the system can always be reduced by the formation of the correlation lattice. This result is again proved in this note by following essentially the method used by Heisenberg, but using a lattice of orthogonal wave-packets. The crude estimate of the transition temp. that follows from the treatment agrees in order of magnitude with that of the usual supraconductor better than that given by Heisenberg. The note concludes with a brief discussion of the exchange correlation energy, which shows that conclusions reached without taking account of this effect more properly are necessarily somewhat uncertain.

***Relation Between Lattice Vibrations and London Theories of Supraconductivity.** J. Bardeen (*Phys. Rev.*, 1951, [ii], 81, (5), 829–834).—B. extends his theory of supraconductivity based on interaction between electrons and lattice vibrations to obtain the London phenomenological equations of supraconductivity. Perfect diamagnetism of the supraconducting electrons and the phenomenological equations follow, if it is assumed that the effective mass of the supraconducting electrons is $\sim 10^{-4}$ electron masses. The use of this effective mass in the conditions for the validity of the London theory is subsequently justified.—P. C. L. P.

Choice of Gauge in London's Approach to the Theory of Supraconductivity. J. Bardeen (*Phys. Rev.*, 1951, [ii], **81**, (3), 469–470).—A letter. A gauge invariant derivation of the phenomenological equation of supraconductivity from quantum theory is described in math. terms.—P. C. L. P.

Hall Coefficient of Semi-Conductors. H. Jones (*Phys. Rev.*, 1951, [ii], 81, (1), 149).—A letter. The theory of the Hall coeff. of semi-conductors given by Johnson and Lark-Horovitz (*ibid.*, 1950, [ii], 79, 176; *Met. Abs.*, 1950–51, **18**, 420) assumes that the mean free path for thermal scattering is independent of the velocity v of the electrons, and that the mean free path for impurity scattering is $\propto v^4$. In their treatment, J. and L.-H. also assume that the resistivity can be expressed as the sum of the resistivity due to thermal scattering plus that due to impurity scattering. J. points out that the second assumption is inconsistent with the first and derives the Hall coeff. from the first assumption alone. Some consequences of the results of this derivation are discussed.—P. C. L. P.

***A Simplification of the Hartree–Fock Method.** J. C. Slater (*Phys. Rev.*, 1951, [ii], 81, (3), 385–390).—The Hartree–Fock equations for calculating the behaviour of electrons in a periodic field using the self-consistent approximation are analysed, and the significance of the various terms is interpreted fairly simply. An important term includes the exchange charge, which corrects for the fact that the electrons do not act on themselves. A simplification of the method is proposed adopting a weighted mean of the exchange charges so that all the electrons can be considered to move in a fixed average potential field, instead of computing the potential field for each electron separately. The calculation of the mean exchange charge is further simplified for the free-electron case. The accuracy obtainable is not much less than that using the Hartree–Fock method, whereas for heavy atoms the simplification in computation is substantial. —P. C. L. P.

Atom Movements. (——). See col. 248.

2 — PROPERTIES OF ALLOYS

***Dimensional Changes in an Aluminium–Copper Alloy During the Individual Stages of Precipitation [During Ageing].** Johann Christian Lankes and Günter Wassermann (*Z. Metallkunde*, 1950, **41**, (11), 381–391).—A study is reported of the changes in dimensions and hardness on age-hardening, at temp. of 20°–450° C. for periods up to 24 days of specimens of an Al–Cu (4·08%) alloy which had been prepared from pure Al (99·99%) and electrolytic Cu. The accuracy of the measurements was such that a 0·0005% change in length could be determined. The results are shown in tables and graphs. On ageing at temp. below 100° C. a contraction occurs first, which is related to the formation of Guinier–Preston zones; at higher temp. and up to 200° C. this contraction is soon followed by considerable expansion, which is caused by pptn. of the phase θ'. Above 200° C. expansion only occurs, but below 260° C., when such expansion has ceased, a contraction then takes place, due to the transformation $\theta' \rightarrow \theta$. Calculations of the activation energies for the pptn. processes show that they are 33,500 and 24,000 cal./g.-mol. for temp. of 150°–200° and 200°–340° C., resp., while the value for Guinier–Preston zone formation is 7000 cal./g.-mol. All the dimensional changes could be correlated to changes in hardness; increase in hardness takes place during Guinier–Preston zone formation as well as during the pptn. processes, but, owing to the variations of the extent of combinations of these phenomena, the rate of increase of hardness does not necessarily correspond with the increase in temp. Reversion phenomena, which can take place during both the Guinier–Preston zone formation and the pptn. processes, afford an explanation of all the other changes in properties which are observed. The alterations which occur during a second ageing at temp. considerably lower than those used in the initial ageing are also described. It is concluded that the results show age-hardening to be a continuous process, viz. α-solid soln. $\rightarrow$ Guinier–Preston zones $\rightarrow \theta'$ phase $\rightarrow \theta$ phase. 28 ref.—E. N.

***Decomposition of a Solid Solution of Silver in Aluminium.** N. V. Ageev and M. A. Skryabina (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1947, **15**, 80–87; *C. Abs.*, 1950, **44**, 5767).—[In Russian]. Alloys were prepared from pure Ag and Al contg. Fe 0·23 and Si 0·03%. The alloys, which were cast in steel moulds preheated to 150°–200° C., contained 3·83–40·41% Ag. The ingots were annealed for 5 days at 500°–510° C. and quenched in cold H_2O. Up to 20% Ag, the ingots were cold-rolled 45–50% and above 20% Ag they were rolled 14–15%. From these ingots were cut test specimens which were heated for 3–4 hr. at 510° C., quenched, and aged at 100°, 150°, and 200° C. The ageing of these specimens was followed by studying their hardness, elect. resistance, microstructure, and radiographs. The changes in properties with ageing time are presented in curves. Two factors determined the properties of an alloy undergoing ageing: decreasing amount of Ag in the basic solid soln. and sepn. of the excess component. A comparison between the experimental and calculated curves showed that the hardness of aged Al–Ag alloys was determined primarily by the sepn. of the excess component, while the effect of the decrease in the concentration of the basic solid soln. was secondary. In the case of the elect. resistance, the opposite was true. Both the hardness and the elect. resistance declined under the influence of the changes in the concentration of the solid soln. on ageing. With the sepn. of the excess component, the hardness rose sharply, while the elect. resistance decreased. An alloy contg. 8–9% Ag had the greatest practical advantages. This alloy, aged for 50 hr. at 150° C., had a B.H.N. of 71, an increase of 94%, and a U.T.S. of 20·7 kg./mm.², an increase of 96%.

[Discussion on a Paper by J. E. Dorn, P. Pietrokowsky, and T. E. Tietz on :] The Effect of Alloying Elements on the Plastic Properties of Aluminium Alloys. —— (*J. Metals*, 1951, **3**, (7), 556).—Cf. *Met. Abs.*, 1950–51, **18**, 455.

***Effects of Alloying Elements on the Electrical Resistivity of Aluminium Alloys.** A. T. Robinson and J. E. Dorn (*J. Metals*, 1951, **3**, (6), 457–460).—The alloys studied were the same as those previously used for the detn. of the effects of alloying elements on the plastic properties of Al (D., Pietrokowsky, and Tietz, *ibid.*, 1950, **188**, 933; *Met. Abs.*, 1950–51, **18**, 455), and all measurements were made at room temp. (22°–26° C.) and were corrected to 24° C. The resistivities of the various alloys are shown in tables and graphs. The usual linear increase of resistivity of dil. soln. with at.-% of solute atoms was obtained, and the appn. of Linde's rule to the data suggests that each Al atom contributes 2·5 electrons to the metallic bond. 15 ref.—E. N.

***The Distribution of Tensile Properties in Extruded Sections [of a Duralumin-Type Alloy]. II.—Sections Having a Wood-Fibre Fracture.** H. Kostron (*Metall*, 1951, **5**, (3/4), 58–63).—Cf. *ibid.*, 1950, **4**, 451; *Met. Abs.*, this vol., col. 13. A report of further studies on extruded sections of the same Duralumin-type alloy, but on sections which show wood-fibre fractures—elongated surfaces of sepn. caused by the intermetallic compounds segregating around the primary grain boundaries of the cast material and extending in the direction of working so that they form stringers. Micro-tensile and fatigue tests were carried out to determine the influence of these stringers on the mech. properties. Except in a direction ⊥ to that of the stringers, the static tensile properties were found to be only slightly lower than those of material showing a normal type of fracture; the fatigue properties were, however, very inferior. It is concluded, therefore, that cast material with its intermetallic compounds segregated around the grain boundaries is unsuitable for heavy directional forming operations; its prodn. can be avoided, however, by the use of special methods of continuous casting. 5 ref.—E. N.

***Effects of Prior Static and Dynamic Stresses on the Fatigue Strength of Aluminium Alloys.** John A. Bennett and James L. Baker (*J. Research Nat. Bur. Stand.*, 1950, **45**, (6), 449–457).—Fatigue tests were carried out on specimens of Alclad 24S-T sheet after appn. of a static bending load and on bare 24S-T sheet after prestressing at a smaller amplitude than that used in the subsequent test. Tests were made on Krouse sheet-bending machines, and a new type of specimen, having each side formed by a circular arc of 4 in. radius, was found to give better results. The static bending load had a significant effect on the fatigue life of the specimens if the stress amplitude in the fatigue test was small. A large static load in the same direction as the fatigue load increased the fatigue life and in the reverse direction decreased it. The fatigue tests were stopped by means of fracture wires when cracks had formed and by extrapolation of the subsequent rate of growth of cracks the number of cycles to start cracking was determined. Fatigue damage defined as $D = \frac{N_c - N_{cp}}{N_c} \times 100$, where N_c, N_{cp} are the number of cycles to produce a crack in the original material and the prestressed material, was found to be proportional to the cycle ratio within the experimental error for prestress amplitudes of 32,500 and 22,500 lb./in.²; but prestressing at 17,000 lb./in.² resulted in negative damage, i.e. an increase in fatigue life. When the stress amplitude in the subsequent test was 20,000 lb./in.² the increase in life was >400%. 9 ref.—J. W. C.

***Effects of Precompression on the Behaviour of the Aluminium Alloy 24S-T4 During Cyclic Direct Stressing.** S. I. Liu (*J. Metals*, 1951, **3**, (6), 452–456).—A study is reported of the effects of cyclic straining of ¾-in.-dia. rods of the commercial Al alloy, 24S-T4 which had been re-soln. heat-treated at ~495° C. for 45 min., quenched in water, and aged for 4 days at room temp. The results show that: (1) such cyclic straining creates two superimposed and counteracting effects, viz. strain-hardening and deterioration; after small amounts of precompression, strain-hardening prevails during the first few cycles, while after large precompression, deterioration prevails from the first cycle; (2) when acting alone, either of the two factors, precompression or cyclic straining, causes a reduction in fracture stress of the material; on the other hand, when acting together, the combined effect is greater than the sum of the two individual effects; (3) within the range of precompression and the number of cycles studied, precompression accelerates the onset of fracture; (4) cyclic straining permanently damages the material; and (5) a pronounced yield inflection appears after repeated tension cycles, and the flow and fracturing characteristics are slightly lowered; this yield inflection is not present in the virgin unstrained metal. 11 ref.—E. N.

[Discussion on a Paper by J. Marin and B. J. Kotalik on :] Plastic Biaxial Stress-Strain Relations for Alcoa 24S-T [Aluminium Alloy Tubing] Subjected to Variable-Stress Ratios. —— (*J. Appl. Mechanics*, 1951, **18**, (2), 221).—See *Met. Abs.*, 1950–51, **18**, 650.

*__Effect of Pulsating Loads on the Creep Characteristics of Aluminium Alloy 14S-T.__ M. J. Manjoine (*Proc. Amer. Soc. Test. Mat.*, 1949, **49**, 788–798; discussion, 799–803).—A creep-rupture testing machine is described in which oscillating and steady loads may be applied, so that the effect of adding a small oscillating stress to a steady stress in a creep-rupture test may be determined. The results of a series of tests at 400° F. (205° C.) on Al alloy 14S-T under an oscillating stress of 10% of the mean stress are given. The effect of the oscillating stress depends on the magnitude of the mean stress. The min. creep rate for a pulsating load test of 20 c./s. was higher than that for a const. load test for a mean stress above 25,000 lb./in.^{2} and lower for lower stresses. The rupture time was poorer for a mean stress above 22,000 lb./in.^{2} and greater below. A possible explanation of the results is discussed.—P. T. G.

Copper-Bearing High-Strength Wrought Aluminium Alloys. —— (*Materials and Methods*, 1950, **32**, (6), 83).—Materials Engineering File Facts No. 202. Data sheet giving the compn., phys., mech., and fabricating properties, corrosion-resistance, and typical uses.—J. W. C.

*__Recent Developments in Al-Zn-Mg Alloys for Sheets and Rivets.__ H. G. Petri, G. Siebel, and H. Vosskühler (*Metall*, 1951, **5**, (3/4), 47–52).—A comparison of the properties of the Al-base alloy Hy43, contg. Zn 4·5, Mg 3·5, Cu 0·3, Mn 0·15, and V 0·03%, with those of Duralumin shows that although there is little difference in tensile properties or resistance to stress-corrosion after cladding, the formability of Hy43 only approaches that of the Al-Cu-Mg alloy immediately after soln. treatment, and is much inferior after age-hardening. The results of a study of a series of Al-Zn-Mg alloys contg. Zn 3·5–4·8, Mg 1·5–3·5, Cu 0·2–1·5, Mn 0·2–0·6, Cr up to 0·15, V up to 0·1, Fe 0·35–0·5, and Si 0·1–0·2% show that the following alloys have excellent tensile properties and improved formability: (1) Zn 3·5, Mg 2·5, Cu 1·0, Mn 0·2, Cr and V $\sim$0·1% each, and (2) Zn 4·5, Mg 1·5, Cu 1·5, Mn 0·5, Cr and V $\sim$0·1% each. However, by virtue of their somewhat high Cu contents, they must be clad with an alloy which does not exhibit stress-corrosion, suitable materials for this purpose being pure Al and its alloys with MgZn$_2$ 3 + Mn 0·2%, or Mg 3, Mn 0·3, and Cu 0·5%. Another alloy which has excellent forming properties is that contg. Zn 3·7 and Mg 2·8%. An alloy suitable for the manufacture of rivets contains Zn 4, Mg 3, Cu 0·4, Cr 0·15, Fe $<$0·5, and Si $<$0·3%, balance Al; it has a shear strength of 27–29 kg./mm.^{2} in the age-hardened condition. 7 ref.—E. N.

[The Effect of Additions of Magnesium on] The Form of Some Heavy-Metal Aluminides. Adalbert Rühenbeck (*Z. Metallkunde*, 1950, **41**, (11), 394–395).—The addn. of 2% Mg to binary Al-base alloys contg. Fe 2, Co 1·5, or Ni 1% was found to suppress the formation of harmful inclusions of the brittle intermetallic compounds Al$_3$Fe, Al$_4$Co, or Al$_3$Ni, which tend to separate out (on solidification) around the grain boundaries or in the form of large plates or needles, and thereby impair the mech. properties of the metal. The Mg addn. result in the formation of ternary compounds which are innocuous.—E. N.

*__Effect of Tin and Antimony on Aluminium.__ D. P. Chatterjee (*Quart. J. Geol., Min. Met. Soc. India*, 1947, **19**, (1), 15–23).—It was found that addn. of Sb and Sn separately and together in various proportions did not lead to any improvement in properties of Al, nor were the alloys so formed amenable to ageing treatment. The working properties of an alloy contg. 1·6% Sn, 0·15% Sb, remainder Al, were studied in some detail.—N. B. V.

*__Cadmium and Its Influence on the Mechanical Properties of Some Non-Ferrous Metals and Engineering Alloys.__ D. P. Chatterjee (*Quart. J. Geol., Min. Met. Soc. India*, 1948, **20**, (1), 1–16).—C. studied the effect of adding up to 3% Cd to Na, Pb, Zn, gun-metal, P bronze, Al bronze, Pb bronze, Duralumin, and Mazak alloy. Addn. of Cd strengthens and hardens Pb and Zn, but has a deleterious effect on the alloys. The corrosion-resistance of Cd, Zn, and Sn when submerged in tap-water was also investigated; Sn proved most and Zn least resistant.—N. B. V.

*__Sigma Phase in Chromium-Molybdenum Alloys with Iron or Nickel.__ John W. Putman, N. J. Grant, and D. S. Bloom (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 61–68; discussion, 69–70).—Cr- and Mo-base alloys have been considered for use at temp. $>$1500° F. (815° C.). A promising alloy (I) contg. Cr 60, Mo 25, Fe 15% was, however, very brittle; this was thought to be due in part to formation of sigma phase. A study was therefore made of sigma formation in the Cr-Mo-Fe system. The extent of the phase was determined and some of its properties described. Mo increases the high-temp. stability of sigma in Fe-Cr alloys. The sigma phase in (I) is stable at 1100° C., and the tendency for the formation of sigma is greater at lower temp., provided sufficient time is allowed. The sigma phase has also been found to occur in Cr-Mo-Ni alloys, though its occurrence has not been reported in any of the 3 constituent binary systems. —P. T. G.

*__Magnetism in Copper Alloys: The Effect of Iron as Impurity.__ Allison Butts and Paul L. Reiber, Jr. (*Proc. Amer. Soc. Test. Mat.*, 1949, **49**, 857–883; discussion, 884–886).—The magnetic susceptibilities and residual magnetic moments of 10 wrought and 10 cast Cu alloys contg. varying % of Fe as impurity were measured after different thermal and mech. treatments. From the data obtained it is possible to specify the max. Fe content in alloys used in the construction of instruments in which only weakly magnetic materials are permissible. Three of the alloys were found to be unsuitable for this purpose under any conditions. The remaining alloys would all be suitable if the Fe content were $\not>$0·01% and some may contain more Fe. The magnetic properties of most of the alloys were lowered by a soln. treatment. Of the cast alloys three common brasses and bronzes were rated as good without heat-treatment; with the others the magnetic properties began to rise rapidly at Fe contents $<$0·1%. Al bronze was the best wrought alloy at low Fe contents and nickel silver (18% Ni, 15% Zn) at higher Fe contents. No other wrought alloys were rated good unless heat-treated. After soln. treatment 4 wrought and 6 cast alloys were good. Annealing caused increases in magnetic properties which were small at very low Fe contents but increased rapidly with Fe content. Magnetism produced by small amounts of unalloyed Fe was compared with that caused by the same amount of Fe alloyed in the specimens. The correlation of magnetic properties with compn. is briefly discussed. 10 ref.—P. T. G.

*__High-Temperature Strength of Several Binary Copper Alloys.__ M. V. Zakharov (*Izvest. Akad. Nauk S.S.S.R.*, **1949**, [Tekhn.], (1), 124–130; *C. Abs.*, 1950, **44**, 5783).—[In Russian]. The Brinell hardness of cast binary Cu alloys contg. up to Zn 63, Sn 38·4, Al 20·5, Be 5, Mn 25, and Ni 45% was determined at 325°–500° C. in the conditions: (*a*) as cast; (*b*) stabilized for 100 hr. at the testing temp.; and (*c*) annealed at 800° C. for 3–4 hr. The Bochvar method of hardness testing for various times up to 1 hr. was used to determine the high-temp. strength. The condition of the alloy had little effect, but at high temp. the two-phase alloys tended to be softer than the α solid soln. alloys, although the reverse was true at 20° C. In the Cu-Ni system the 1-hr. hardness rose continuously at 500° C. from

24 kg./mm.^{2} at 0% Ni to ~53 kg./mm.^{2} at 45% Ni. The 30-sec. hardness values of some intermetallic compounds at 20° and at 500° C. were : β (CuZn) 120, 2·5 ; $CuAl_2$ 300, 20 ; Cu_3Sn 300-350, 27 ; $Cu_{31}Sn_8$ 350–400, 35 ; γ (Cu_5Zn_8) 450–500, 25 ; Cu_5Al_3 500–550, 109. These data give further support to the " dissolving phase " mechanism of softening at high temp. of two-phase alloys. However, in some systems this may be offset by the relatively high hardness retained by the second phase.

***High-Tensile Bronze [Brass] Casting.** D. P. Chatterjee (*Quart. J. Geol., Min. Met. Soc. India*, 1946, **18**, (4), 129–132).—Five alloys were made up by adding Fe (1·0–1·6%), Mn (~2%), Al (~1·7%), and Ni (~1·3%) to 60 : 40 brass. Their tensile properties were determined and their microstructures examined. Values obtained were : 0·2% P.S. = 18–19 tons/in.^{2}, U.T.S. = 36–38·5%, elongation = ~14–17%, B.H.N. = 114–125.—N. B. V.

The Nickel–Tin Bronzes [—II.]. James S. Vanick (*Foundry*, 1951, **79**, (4), 130–133, 278–279).—Cf. *ibid.*, (2), 102 ; *Met. Abs.*, 1950–51, **18**, 651. V. discusses the metallography and strength properties of the " as-cast " and soln.-treated and aged Ni–Sn bronzes in detail. The effects of different heat-treatments are shown in numerous curves, and the influence of compn. is illustrated in a ternary diagram on which lines of equal tensile strength are plotted for alloys soln.-treated at 760° C. and aged 5 hr. at 343° C. V. suggests that the increase in strength results from pptn.-hardening and draws an analogy with the hardening and tempering of steel. The ternary diagram shows that with a given total of alloying elements the highest strengths are attained when the proportions of Ni and Sn are about equal ; alloys contg. 6–8% of each element possess the max. strength (~50 tons/in.^{2}). It is suggested therefore that the hardening constituent contains Ni and Sn in equal proportions. Considerable data are given on the mech. properties of the alloys. —R. W. R.

Nickel-Alloyed Manganese Bronze. James S. Vanick (*Foundry*, 1951, **79**, (6), 122–125, 226, 228).—Cf. *ibid.*, (4), 130 ; preceding abstract. V. describes the advantages of adding up to 3% Ni to high-tensile brasses. The main effect of the Ni addn. is an increase in the resistance to corrosion. Results of corrosion, cavitation-erosion, and other tests carried out by various workers are reviewed. The appn. of Ni-contg. high-tensile brass castings are discussed. 7 ref.—R. W. R.

Wrought Tin and Aluminium Brasses. —— (*Materials and Methods*, 1950, **32**, (3), 85).—Materials Engineering File Facts No. 197. Data sheet giving the compn., phys., mech., and fabricating properties, corrosion-resistance, and typical uses.—J. W. C.

***Identification and Mode of Formation and Re-Solution of Sigma Phase in Austenitic Chromium–Nickel Steels.** E. J. Dulis and G. V. Smith (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 3–29 ; discussion, 30–37).—Cf. *Met. Abs.*, 1950–51, **18**, 654.

***Sigma Phase in Several Cast Austenitic Steels.** V. T. Malcolm and S. Low (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 38–44 ; discussion, 45–47).—Cf. *Met. Abs.*, 1950–51, **18**, 654.

***The Formation of Sigma Phase in 17% Chromium Steel.** J. J. Heger (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 75–78 ; discussion, 79–81).—Cf. *Met. Abs.*, 1950–51, **18**, 654.

***Sigma Phase and Other Effects of Prolonged Heating at Elevated Temperatures on 25% Chromium–20% Nickel Steel.** G. N. Emmanuel (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 82–99 ; discussion, 120–127).—Cf. *Met. Abs.*, 1950–51, **18**, 654.

***The Occurrence of the Sigma Phase and Its Effects on Certain Properties of Cast Fe–Ni–Cr Alloys.** J. H. Jackson (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 100–119 ; discussion, 120–127).—Much of the prodn. of heat-resistant alloy castings is from types of alloy that can contain sigma phase. Although not always detrimental, this phase often needs to be avoided, especially for appn. at certain temp. such as 1600° F. (870° C.). A common alloy contains 26% Cr and 12% Ni, in which sigma-phase formation has been exhaustively studied. Alloys of this type may have lower creep properties and be embrittled if sigma is present, the detrimental effects increasing with the amount of sigma phase produced. Means of avoiding sigma formation by compn. control have been developed. A " thermal-fatigue " test for determining hot-cracking and distortion tendencies at high temp. has been developed which shows the harmful effects of sigma phase. An alloy contg. 26% Cr and 20% Ni is being increasingly used and is less susceptible to sigma formation than the 12% Ni alloy, but comprehensive data on the effect of sigma phase on the properties of the alloy are lacking. Of the other alloys currently used for heat-resistant castings most contain at least 33% Ni and ≯21% Cr and are thought not to be susceptible to sigma formation. A few alloys relatively high in Cr and low in Ni are produced, in which sigma phase is very prevalent. However, these are usually restricted to service appn. where the sigma formation does not matter. —P. T. G.

***A Contribution to the Metallurgy of the Parkes Process. [Equilibria in the Pb Corner of the System Pb–Ag–Zn].** Friedrich Johannsen and Kurt-Heinz Lange-Eichholz (*Z. Erzberg. u. Metallhüttenwesen*, 1950, **3**, (12), 397–402).—After a review of previously published work (Henglein and Köster, *Z. Metallkunde*, 1948, **39**, 391 ; *Met. Abs.*, 1949–50, **17**, 164) on the solubility of Ag and Zn in molten Pb in the Pb corner of the ternary system, a report is presented of a new study of this region, the experimental methods used being similar to those of Jolivet (*Compt. rend.*, 1947, **224**, 1822 ; *Met. Abs.*, 1947–48, **15**, 183). A partial equilibrium diagram is given which shows the phase boundaries for the pptn. of : (1) pure Pb, (2) Ag-rich solid soln., and (3) Zn-rich, Ag-poor solid soln. The coincident boundary of (2) and (3) corresponds to the sepn. of doubly saturated Ag and Zn solid soln., and terminates in a ternary eutectic at Zn 0·5% and Ag 5 g./ton. The results are applied to current German and American practice in the desilverization of Pb by the Parkes process. 5 ref.—E. N.

***Solid Solutions of Lithium in Lead.** S. A. Pogodin and E. S. Shpichinetsky (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1947, **15**, 88–95 ; *C. Abs.*, 1950, **44**, 5783).—[In Russian]. The objects of this study were Pb–Li alloys contg. 0·03–1·1 wt.-% Li. To determine the boundaries of the α-phase, specimens were vacuum-sealed in glass tubes and heated in three batches for 18 days at 200° C., for 25 days at 170° C., and for 20 days at 120° C., resp. They were then quenched in a mixture of solid CO_2 and alcohol (−80° C.). A fourth batch was cooled from 120° to 20° C. during 7 days. The boundaries of the solid soln. of Li in Pb were found to be 2·1 at.-% (0·07 wt.-%) at 200° C., 1·8 at.-% (0·06 wt.-%) at 170° C., and 1·2 at.-% (0·04 wt.-%) at 120° C. The solubility of Li in Pb conformed to the Schroeder–Le Chatelier equation. By extrapolation, the solubility of Li in Pb at the eutectic point (235·15° C.) was 2·6 and at 20° C. 0·35 at.-%. The hardness of Pb–Li alloys increased for the first 2 hr. after casting and then gradually dropped to 10·5–11 B.H.N. at 0·03–0·09 wt.-% Li.

***Ternary Alloys of Lead with Sodium and Lithium.** S. A. Pogodin and E. S. Shpichinetsky (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1947, **15**, 96–111 ; *C. Abs.*, 1950, **44**, 5783).—[In Russian]. Cf. preceding abstract and Kumakov *et al.*, *ibid.*, 1940, **13**, 233 ; *Met. Abs.*, 1941, **8**, 297. Pb–Na–Li alloys with up to 50 at.-% Na and up to 50 at.-% Li were subjected to thermal analysis, and their hardness and microstructure were studied. In the phase diagram the surface of the liquidus comprised 4 areas of crystn. : α, β, NaPb, and LiPb. The system contains 3 eutectics : (1) m.p. 234° C. ; Li 16, Na 5 at.-% ; (2) m.p. 265° C. ; Li 12·3, Na 27·5 at.-% ; and (3) a eutectic formed by NaPb and LiPb (m.p. 310° C.) and contg. Li 15 and Na 36 at.-%. The phase diagram clearly showed the existence of a Berthollide-type β-phase. The boundaries of the ternary solid soln. at 200° and 20° C. were also approx. determined. The alloys with greatest hardness contained Na 0·6–0·8 and Li 0·03–0·06%.

***Influence of [Additions of] Grey Cast Iron on the Grain-Size and Mechanical Properties of Magnesium Alloys.** Walter Mannchen (*Z. Metallkunde*, 1950, **41**, (11), 391–393).—As it was known that finely divided C acts as a nucleating agent during the solidification of Mg–Al melts, experiments were carried out on the Mg-base alloys AZ31, AZF, AZ855, and A9 (Al 3–8·3, Zn 0·4–2·7, and Mn 0·1–0·3%) in order to elucidate the effects of such C being added in the form of filings of grey cast Fe. The results show that considerable grain refinement and corresponding increases in mech. properties take place when the grey Fe addn. (0·4%) are stirred into the molten alloys at temp. above 800°–820° C. and the melt is then allowed to stand for various periods—5–40 min., according to the weight of the charge—before pouring. Such grain refinement is not: (1) observed when filings of pure Fe are used, and (2) influenced by the Si or P in the grey cast Fe. Hence it is concluded that the grain refinement can result only from the action of the graphite particles present in the grey Fe, and that grain refining by this method is more simple and easier to control than when it is carried out by the customary method of addn. of organic compounds.—E. N.

***Ignition Temperature of Magnesium and Magnesium Alloys.** (Fassell, Gulbransen, Lewis, and Hamilton). See col. 165.

[Discussion on a Paper by P. D. Frost, J. G. Kura, and L. W. Eastwood on :] Ageing Characteristics of Magnesium–Lithium-Base Alloys. —— (*J. Metals*, 1951, **3**, (7), 560–561).—Cf. *Met. Abs.*, 1950–51, **18**, 507.

[Discussion on a Paper by R. S. Busk, D. L. Leman, and J. J. Casey on :] The Properties of Some Magnesium–Lithium Alloys Containing Aluminium and Zinc. —— (*J. Metals*, 1951, **3**, (7), 556–557).—Cf. *Met. Abs.*, 1950–51, **18**, 458.

[Discussion on a Paper by P. D. Frost, J. H. Jackson, A. C. Loonam, and C. H. Lorig on :] The Effect of Sodium Contamination on Magnesium–Lithium-Base Alloys. —— (*J. Metals*, 1951, **3**, (7), 557–558).—Cf. *Met. Abs.*, 1950–51, **18**, 422.

Thermodynamic Investigations on Ternary Amalgams. Carl Wagner (*J. Chem. Physics*, 1951, **19**, (5), 626–631).—Under certain conditions interaction between positively charged metal ions in an alloy may be considered of minor importance in comparison with interaction between conduction electrons and metal ions. The activity of a solute metal *B* dissolved in metal *A* will be increased by a third component *C* if metals *B* and *C* change the valence-electron : atom ratio in the same direction, and conversely the activity of a solute metal *B* will be decreased by a third component if metals *B* and *C* change the valence-electron : atom ratio in opposite directions. Ref. is made to the changes in (*a*) the solubility of H in Cu by Zn, Sb, Al, Pt, and Ni as alloying elements, (*b*) the vapour pressure of Zn dissolved in Cu by Al and Ni, and (*c*) the activity of alkali metals dissolved in Hg by Tl. A derived equation permits calculation of the change in the activity of metal *B* dissolved in metal *A* owing to the addn. of a solute metal *C* if the changes in the activity coeff. with compn. for the binary systems *A*–*B* and *A*–*C* are known and interaction between positively charged metal ions can be disregarded; this relation has been confirmed for the systems Hg–Li–Tl, Hg–Na–Tl, Hg–K–Tl, and Hg–Na–K. 40 ref.—J. R.

***Magnetic Susceptibility of Some Amalgams.** D. Krishna Iyer (*J. Annamalai Univ.*, 1941, **11**, 77–96; *C. Abs.*, 1950, **44**, 5661).—Cf. Rao and Aravamuthachari, *Proc. Indian Acad. Sci.*, 1939, [A], 9, 181; *Met. Abs.*, 1939, 6, 203. Measurements on dil. amalgams (all below 2%) of Zn, Cd, Sn, Pb, Se, and Bi, made by the Gouy method, are tabulated and shown graphically. Values for the pure metals, including pure Hg, are also tabulated and compared with values obtained by other investigators. The results are applied to an explanation of the phys. nature of the amalgams. 29 ref.

***The Measurement of Young's Modulus of Metals and Alloys by an Interferometric Method. I.—Young's Modulus of Nickel–Copper Alloys.** Tadao Fukuroi and Yoshio Shibuya (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 748–757).—[In English]. An interferometric method of observing the flexural strain in a rod, subjected to pure bending stress, is described and applied to a study of the variation of Young's modulus of 12 Ni–Cu alloys contg. 1–100% Ni. The results agree qualitatively with Yamamoto's results, obtained by a dynamical method (*Nippon Kinzoku Gakkai-Si*, 1942, **6**, 249).—J. S. G. T.

***The Equilibrium Diagram of the System Nickel–Manganese.** B. R. Coles and W. Hume-Rothery (*J. Inst. Metals*, 1951–52, **80**, (2), 85–92).—The equilibrium diagram of the Ni–Mn system above 800° C. has been determined by a combination of thermal, microscopical, and X-ray methods; some information about the diagram at lower temp. was also obtained. The liquidus and solidus pass through a smooth min. at 62 at.-% Mn and 1018° C. Ni is freely soluble in γ-Mn and increases the axial ratio of the face-centred tetragonal structure until it becomes cubic and unites with the solid soln. of Mn in Ni. Immediately below the solidus there is thus a complete range of solid soln. extending across the diagram. At lower temp. (~910° C.) this solid soln. in the equi-atomic region is transformed into a β phase with a b.c.c. structure which cannot be retained by quenching. At a still lower temp. the β phase is transformed into a phase with a face-centred tetragonal superlattice of the CuAu type. The temp. of the $\delta \rightleftharpoons \gamma$ transformation of Mn is raised by the addn. of Ni, whilst those of the $\gamma \rightleftharpoons \beta$ and $\beta \rightleftharpoons \alpha$ transformation of Mn are lowered. The diagram in some respects resembles that of the system Ni–Zn, and this suggests that Mn acts as a divalent element in some of the alloys.—Authors.

***Alloys Between Ferromagnetic and Diamagnetic Metals : Effect of the Structure and the Ferromagnetic Metal [Nickel] on the Mean Moment.** André J. P. Meyer and Pierre Taglang (*J. Phys. Radium*, 1951, **12**, (6), 54S–55S).—Detn. of the ferromagnetic moments of binary alloys of ferromagnetic metals with diamagnetic metals have been made mostly with Ni alloys, and when values of these moments for these Ni alloys are plotted as ordinates against the corresponding electronic compn., the very approx. linear graphs have a slope of −1 Bohr magneton/electron ($-\mu_B/e$). These results show that the free electrons of the metal alloyed with Ni are incorporated in the atmosphere of electrons, 4*s*, of Ni, in conformity with the hypothesis of dynamic equilibrium proposed by Néel ("Le magnétisme", II, Strasbourg, **1940**, p. 105). They gave a value of the *d* of free electrons equal to 0·6 electron/atom of alloy, and are const. so long as the alloy is ferromagnetic. M. and T., from results deduced from a large series of alloys of ferromagnetic-diamagnetic metals, show that these results cannot be generalized to apply to alloys of different cryst. structures: the slopes of the various graphs corresponding to metals with different crystal structures vary with the structure, and the value of the magnetic moment for a given electron concentration, and a given lattice, depends upon the nature of the carrier of the moment.—J. S. G. T.

***Measurement of the Gyromagnetic Constant of Ferromagnetic Elements and of Some Alloys of Equal Electron : Atom Ratio of the Iron Group.** André J. P. Meyer (*Ann. Physique*, 1951, [xii], **6**, (Jan./Feb.), 171–210).—Precision measurements by a magnetic-resonance method were made of the Landé (gyromagnetic) ratio *g*, for pure Fe and Ni, and for 2 pairs of isomorphous solid soln. each with the same electron : atom ratio, viz. FeNi and (FeNi)–Co 73·5 at.-%, and FeCo and $\mathrm{Fe_3Ni}$. For all these substances $g = 2{\cdot}00$, which shows that the electronic spin alone and not the orbital moment contributes to the atomic magnetic moment, contrary to some previously published evidence. In view of the fact that $g = 2$ for the "isoelectronic" alloy pairs, it is concluded that the difference in their magnetic moments cannot be due to different orbital moments, but must be attributed to a variable excitation of unpaired electrons into the 3*d* orbit (where their spins contribute to the atomic moment). The apparatus and its use are described in detail, and certain insurmountable difficulties are pointed out, which preclude measurements on "magnetically hard" substances such as Co. 30 ref.—R. W. C.

***Precipitation and the Domain Structure of Alnico V.** A. H. Geisler (*Phys. Rev.*, 1951, [ii], **81**, (3), 478–479).—A

letter. A microstructure of Alnico V, soln. heat-treated and aged for 300 hr. at 800° C. without any external magnetic field, is reproduced and discussed. This reveals the importance of spontaneous internal fields on the pptn. processes in the alloy. The discussion is extended to account for the differences found between Alnico II and Alnico V.
—P. C. L. P.

***Phase Diagram of the Ni–O_2 System and the Physico-Chemical Nature of the Solid Phases in This System.** D. P. Bogatsky (*Zhur. Obshch. Khim.*, 1951, **21**, (1), 3–9).—[In Russian]. Investigations have revealed the existence of only 3 oxides: NiO_2, Ni_2O_3, and NiO. Their physico-chem. and thermodynamic properties have also been studied.
—N. B. V.

***The Structure of the Carbide System TiC–TaC–WC.** H. Nowotny, R. Kieffer, and O. Knotek (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1951, **96**, (1), 6–8).—A microscopic and X-ray study is reported of this system over the compn. range WC 31·4–91·5, TiC 2·0–48·0, and TaC 6·5–76·4 at.-%. The specimens were prepared by powder-metallurgy methods, sintering being carried out at 2200° ± 50° C.; some of the specimens were then annealed for 1 hr. at 1450° C. *in vacuo* or in H; others were ground, sieved, mixed with ~8% Co to act as a binder, extruded through graphite dies at 1500° C. and 150 kg./cm.^{2} pressure, and then annealed for 40 hr. Equilibrium diagrams show the phase fields existing at 1450° and 2200° C., viz. an extensive field of homogeneous (Ti, Ta, W)C solid soln. and a heterogeneous region consisting of this solid soln. and practically pure WC. The homogeneous region becomes more extensive with increase in temp., i.e. increase in solubility of WC in the continuous solid-soln. system TiC–TaC, until at 2200° C. ~30 mole-% WC is in soln. Specimens which had been annealed at 1450° C. showed: (1) less porosity than those which had been freshly sintered, (2) a coarse crystal formation, and (3) the presence of small amounts of a third phase—of a blue colour, and brittle—at the grain boundaries; this is probably a Co-contg. phase or a double carbide. 9 ref.
—E. N.

Cast Zinc Alloys. —— (*Materials and Methods*, 1950, **32**, (4), 91).—Materials Engineering File Facts No. 199. Data sheet giving the compn., phys., mech., and fabricating properties, corrosion-resistance, and typical uses.—J. W. C.

Wrought Zinc Alloys. —— (*Materials and Methods*, 1950, **32**, (4), 93).—Materials Engineering File Facts No. 200. Data sheet. See also preceding abstract.—J. W. C.

***Zinc-Base Alloys for Die-Casting. I.—Effect of the Quality of Zinc on the Decay Phenomena of Zamak.** Takeo Hikage (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 719–724).—[In English]. Seven samples of Zamak, made from Zn of purity 98·4—>99·99%, were tested for: (1) change of length, (2) Charpy impact value, and (3) hardness, after exposure for 10 days in steam at 95° C., which was considered to be equivalent to an exposure for ~5 years to the atmosphere. Re-electrolysed Zn, of purity $\nless$99·99%, used in making Zamak, shows no deterioration. Zn, of purity 99·98%, may also be used for this purpose, provided that a suitable proportion (0·03–0·1%) of Mg is added to the alloy. Zamak, free from Cu, showed the least decrease of impact value and very little expansion, under test, and it is considered that electrolytic Zn of somewhat lower purity can be used in this alloy, provided a proper proportion of Mg is added. The decrease of impact value of Zamak contg. 3% Cu is very large, even when highest-purity Zn is used. Addn. of Cd affects unfavourably the decay phenomena of Zamak, especially in the range 0·3–0·5% Cd. Addn. of some Cd to the ternary eutectic Zn–Al–Cu alloy, made with electrolytic Zn, improves the properties of the alloy.
—J. S. G. T.

***Phase Relationships in the Binary Systems of Nitrides and Carbides of Zirconium, Niobium, Titanium, and Vanadium.** Pol Duwez and Francis Odell (*J. Electrochem. Soc.*, 1950, **97**, (10), 299–304).—It was found by X-ray-diffraction methods that ZrN, NbN, TiN, and VN, all of which have the same NaCl-type structure as the corresponding carbides, form continuous series of solid soln. with each other, with the exception of the pair ZrN–VN (the solubility of ZrN in VN <1% and that of VN in ZrN approx. 5%). The following carbide–nitride systems form continuous series of solid soln.: TiC with VN, TiN, NbN, and ZrN; NbC with VN, TiN, and NbN; and VC with VN, TiN, and NbN. The system VC–ZrN showed very limited solubility. In the system NbN–ZrN no reaction took place at a firing temp. of 2450° C. Carbide–nitride systems involving ZrC were not investigated because ZrC decomposes when fired in a N atmosphere at high temp. In general, the solubility relationships in the above systems are governed by the atomic-size-factor rule, the only exception being in the case of the system NbC–ZrN, in which no solubility could be detected.—N. B. V.

***Dynamic Creep and Rupture Properties of Temperature-Resistant Materials Under Tensile Fatigue Stress.** B. J. Lazan (*Proc. Amer. Soc. Test. Mat.*, 1949, **49**, 757–787; discussion, 799–803).—The limitations of static testing and the importance of dynamic creep and rupture in designing for high-temp. service are discussed, and newly developed dynamic testing machines and measuring equipment for determining creep and rupture properties are described. Data on four temp.-resistant materials are presented: (i) Fe 32, Cr 22, Ni, 19, Co 19, W 1·9, Mn 1·7, and Nb 0·8%; (ii) Co 44, Cr 20, Ni 20, W 3·9, Nb 3·8, Mo 3·7, Fe 2·6, and Mn 1·0%; (iii) Co 61·8, Cr 26·7, Mo 5·9, Ni 2·7, and Fe 1·3%; (iv) Fe 67, Cr 19, Ni 9, Mo 1·4, W 1·1, and Mn 1·1%. The results are presented by plotting a family of curves within alternating-stress/mean-stress co-ordinates. The relationships between testing temp. and dynamic stress and their influence on creep and time to rupture are shown. The increased creep and rupture resistance during some of the dynamic tests is discussed in terms of possible met. changes caused by cyclic stress. Data presented show the greatly decreased ductility caused by the superposition of cyclic stress on tensile load. 11 ref.—P. T. G.

Heating-Element Alloys. A. Schulze (*Elektrotechnik (Berlin)*, 1950, **4**, 183–188; *Elect. Eng. Abs.*, 1950, **53**, 471).—A comprehensive review, with tables and curves illustrating the phys. and chem. properties of the alloys.

†The Effect of Plastic Deformation on Solid Reactions. I.—Diffusion Reactions. B. L. Averbach (*Cold Working of Metals* (*Amer. Soc. Metals*), **1949**, 262–289).—Modern views on the mechanism of pptn.-hardening are summarized, with special ref. to the free-energy changes involved, and the effects of cold work on the age-hardening reaction are "evaluated qual.". When a soln.-treated alloy is cold worked, the speed of ageing in almost invariably increased, presumably because nucleation occurs more rapidly along slip planes. In some systems, such as Be–Cu and Be–Ni, this treatment also increases the max. hardness obtainable. In others, such as Al–Ag, recrystn. may occur during ageing, and the hardening on ageing may be decreased by previously cold working the solid soln. Even in alloys where no recrystn. occurs, as Al–Cu, cold-working may destroy the coherency between the precipitate and matrix and thus decrease the max. hardness. 33 ref.—J. C. C.

***The Effect of Plastic Deformation on Solid Reactions. II.—The Effect of Applied Stress and Strain on the Martensite Reaction.** B. L. Averbach, S. A. Kulin, and Morris Cohen (*Cold Working of Metals* (*Amer. Soc. Metals*), **1949**, 290–319).—Existing data are reviewed and some new experiments on austenitic steels are reported.—N. B. V.

Theories of the Age-Hardening of Metallic Alloys. F. Rohner (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1951, **96**, (2), 21–25).—A brief review, with 40 ref.
—E. N.

***Segregation in Regular Ternary Solutions.—I.** J. L. Meijering (*Philips Research Rep.*, 1950, **5**, (5), 333–356).—The segregation in ternary phases is examined from a thermodynamic point of view. Consideration is given to regular soln. defined by the equation for the Gibbs free-enthalpy surface: $G = axy + bxz + cyz + RT(x\ln x + y\ln y + z\ln z)$, where x, y, and z are the mole fractions of the three components of the system and a, b, and c are the energy parameters for the interaction between different atoms. The parameters a, b, and c are varied independently, and their different combinations are divided into eight characteristic categories. Taking the sequence $a \gg b \gg c$, closed miscibility gaps appear above $T = a/2R$ if $b - c > a > 0$, and above $T = 0$ if $\sqrt{-c} > \sqrt{-b} + \sqrt{-a}$. The ternary critical point in which such a gap vanishes on heating is calculated. Complicated spinodal curves are found when $b + c > a > 0$.—D. K. W.

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3 — STRUCTURE

(Metallography; Macrography; Crystal Structure.)

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

***Metallographic Techniques for High-Purity (99·995%) Aluminium.** Italo S. Servi and Nicholas J. Grant (*J. Metals*, 1951, **3**, (6), 473).—The following polishing and etching techniques are recommended: (1) *Electrolytic Polishing.*—Electrolyte $HClO_4$ (72%) 5 ml., glacial CH_3COOH 95 ml.; 25–60 V. gives a good polish in ~2 min. at room temp. after prepn. on 000 emery. (2) *Macro-etching* of coarse-grained samples. After polishing, the sample is retained in the above-mentioned bath for a few sec. with the voltage lowered to 10 V. (3) *Galvanic Etching.*—The technique described by S. (*Metal Progress*, 1950, **58**, 732) is recommended. (4) *Production of Etch Pits.*—The technique suggested by Lacombe and Beaujard (*J. Inst. Metals*, 1947, **74**, 1; *Met. Abs.*, 1948–49, **16**, 707) is improved if 70% HNO_3 is substituted for fuming HNO_3, the reagent then having the compn.: HNO_3 (70%) 70, HCl (37·5%) 50, and HF (48%) 3 ml. It is pointed out that none of these techniques is suitable for the examination of commercially pure Al. 3 ref.—E. N.

[Discussion on a Paper by B. R. Banerjee on :] A Study of Strain Markings in Aluminium. —— (*J. Metals*, 1951, **3**, (7), 547–548).—Cf. *Met. Abs.*, 1950–51, **18**, 425.

***Inhomogeneities in the Plastic Deformation of Metal Crystals. I.—Occurrence of X-Ray Asterisms [from Cadmium and Aluminium]. II.—X-Ray and Optical Micrography of Aluminium.** R. W. K. Honeycombe (*J. Inst. Metals*, 1951–52, **80**, (2), 45–49; 49–56).—[I.—] A study has been made of the occurrence of X-ray asterisms from deformed single crystals of a typical hexagonal metal (Cd) and a typical cubic metal (Al). Cd can be extended by >100% elongation in tension without the appearance of asterisms in X-ray Laue photographs; asterisms occur only when crystals are macroscopically bent or kinked. On the other hand, they are present in photographs from Al crystals after <4% elongation in tension. These asterisms contain intensity max. which are shown to be a direct consequence of the deformation and not of a secondary phenomenon such as polygonization. On annealing, extended Cd crystals give sharp Laue spots, and the bent crystals polygonize very uniformly at 300° C. The appearance of asterisms from lightly extended Al crystals is unaltered by annealing at 250° C., at which temp. partial recovery of the yield stress occurs; but when the temp. is raised to 475° C. the asterisms become further fragmented as a result of polygonization. These experiments suggest that tensile deformation causes Al, but not Cd, crystals to break down into a series of slightly disoriented blocks connected by regions of distortion or lattice curvature. [II.—] A technique of X-ray microscopy is used in conjunction with opt. micrography to investigate the structure of large Al crystals deformed by small amounts in tension. Two types of microscopic inhomogeneity which cause local variations in orientation are described. First, there are narrow regions of curvature or *kink bands* separating slightly disoriented lamellæ of the crystal; these bands occur initially on (110) planes, the normal to which coincides with the active slip direction. The second type of inhomogeneity comprises *bands of secondary slip*, which are regions nearly devoid of primary slip traces, in which slip on another system occurs preferentially. These bands are almost ∥ the primary slip planes in the early stages of deformation. The effects of crystal orientation, temp. and speed of deformation, and purity of metal on the occurrence of kink bands are investigated, and it is found that these are usually present after deformation, except when this occurs from the beginning by conjugate slip. Bands of secondary slip also generally occur, and in crystals deforming by conjugate slip, they are often of macroscopic size. The inhomogeneities play an important role in the deformation of Al and account for many of the differences between this metal and Cd, such as the occurrence of X-ray asterisms, the marked differences in strain-hardening, and the extent to which recovery can occur on annealing.
—AUTHOR.

***Study of the Internal Disorientations During Plastic Deformation of a Metallic [Aluminium–Magnesium] Single Crystal.** Jean Hérenguel and Pierre Lelong (*Compt. rend.*, 1951, **233**, (1), 53–55).—See also: H. and Segond, *Rev. Mét.*, 1949, **46**, 377; H., *ibid.*, 1948, **45**, 505; H. and Lacombe, *Compt. rend.*, 1949, **228**, 846; and H. and Lelong, *ibid.*, 1951, **232**, 2218; *Met. Abs.*, 1949–50, **17**, 14; 1948–49, **16**, 486; 786. It has previously been shown that the rate of anodic oxidation of the Al–Mg alloys depends on: (1) the local compn. of the solid soln., and (2) the orientation of the crystal face submitted to the oxidation treatment. If a polcryst. specimen of the Al–Mg alloy, perfectly homogenized but cold-worked, is submitted to this treatment, striations will appear on the surface of certain grains. H. and L. used an attack by Lacombe's reagent on this structure as a means of analysing the texture of the solidification twins and the recrystn. of Al. A single crystal of the perfectly homogenized Al–3% Mg alloy was cold-rolled from 5 to 3 mm., electrolytically polished, and etched. At low magnification (×5) alternate strongly and feebly attacked parallel zones in 4 directions were observed. At higher magnifications (e.g. ×600) it was found that the dark zones contained numerous and approx. identical figures; in the light zones, the figures were different and much less numerous. The single crystal had thus been fragmented into zones of small extent in two different systems. Microscopic examination indicated that the two angles of attack are dependent on two fundamental systems virtually in the position of twins. Further transitions between these two orientations were observed.—J. H. W.

***[Grain-Growth and] Creep Deformation of Metals [Aluminium–Silver Alloy].** L. Rotherham and L. W. Larke (*Research*, 1950, **3**, (9), 434–436).—A letter. Grain-growth effects are reported in Al–0·5% Ag alloy tested under creep conditions at 300° C.; the effect on the creep rate was not large. Large crystals formed in this way showed a structure resembling that of polygonized material; the sub-grains showed a considerable difference in orientation, and also evidence of considerable internal strain. R. and L. conclude that the breakdown to sub-grains is not the primary cause of creep, but a secondary effect resulting from creep; breakdown to sub-grains occurs only in heavily deformed regions, and creep occurs then on new planes and leads to further sub-grain formation till the whole crystal is completely broken down to sub-grains.—F. A. F.

***Secondary Recrystallization in Copper Wire.** Guido Bassi (*J. Metals*, 1951, **3**, (7), 533–534).—X-ray-diffraction patterns were taken of electrolytic tough-pitch Cu wire which had been subjected to 99·8% reduction in cross-section, both before and after annealing at various temp., with the following results: (1) in hard-drawn wire the texture at the centre is a mixture of [111] and [100] directions; (2) after annealing for a few min. at 200° C. there is noticeable recrystn., with preferred orientation in the [100] direction at the expense of the [111] deformation texture; with longer times of annealing the final structure is mainly [100] with a small amount of [111] and random orientation; (3) after annealing at 400° C., the recrystn. structure first shows a [100] texture; with longer annealing times the fine-grained structure transforms into coarse grains with a preferred [112] orientation; and (4) the [112] texture can be obtained also after annealing the hard-drawn wire for 6 hr. at 200° C., and then giving a 9·5% reduction in cross-section and annealing again at 200° C. The observations indicate that: (i) the occurrence of secondary recrystn. at 400° C. is affected by other factors of the previous treatment, in addn. to the final deformation; this may be the penultimate deformation and annealing, and (ii) the results obtained under (4) are more logically called secondary recrystn. than primary recrystn. after critical deformation as reported by Kronberg and Wilson (*Trans. Amer. Inst. Min. Met. Eng.*, 1949, **185**, 501; see *Met. Abs.*, 1949–50, **17**, 167). 7 ref.—E. N.

***Influence of Impurities and of Recrystallization on the Crystalline Orientation of Hot- and Cold-Drawn Copper.** Paul Bastien and Jean Pokorny (*Compt. rend.*, 1951, **232**, (26), 2447–2449).—B. and P. undertook an investigation on a range of hot- and cold-drawn Cu bars contg. increasing addn. of the impurities Sb, Bi, As, and Ag, by means of diagrams obtained with Cu$K\alpha$ radiation in a horizontal cylindrical camera aligned on the pencil of incident X-rays and an integrating apparatus using a large number of crystal reflections. The orientations obtained in the axial regions of bars cold drawn with max. deformation (73%) in the range studied were:

	Cu	Cu + 0·45% Sb	Cu + 0·009% Bi	Cu + 0·47% Ag	Cu + 1% As
Randomly orientated crystals, %	25	45	27	42	30
Crystals with orientation [111], %	50	40	48	39	52
Crystals with orientation [100], %	25	15	21	19	12

The textures corresponding to the orientation [111] in the direction of drawing were in all the specimens more perfect than those corresponding to the orientation [100]. B. and P. also studied the modification of the texture by recrystn. after heating between 300° and 1000° C. Between 300° and 500° C., the orientations weaken rapidly for [111] and slowly for [100]; the % of grains randomly orientated is max. at ~500° C. (75–85%), particularly for Cu contg. Sb and As, which form solid soln. Between 500° and 1000° C., X-rays reveal a decided strengthening of the orientation [111], particularly for pure Cu. This is explained from a micrographic study of corrosion figures by a preferential coarsening of certain grains at the expense of the randomly orientated crystals and increasing the vol.-% occupied by the [111] oriented grains in the metal.—J. H. W.

Origin of the Cube Texture in Face-Centred-Cubic Metals [Copper and Aluminium]. Paul A. Beck (*J. Metals*, 1951, **3**, (6), 474–475).—The occurrence of the "cube" texture upon annealing cold-rolled Cu and Al is discussed. It is considered that the oriented growth theory of annealing texture readily explains, at least qual., the various conditions and their inter-relations, thus far described in the literature, that affect the formation of this type of texture. The effect of alloying elements in suppressing its formation appears to be associated with a change in the rolling texture. 11 ref.—E. N.

[Discussion on a Paper by B. F. Decker and D. Harker on :] Activation Energy for Recrystallization in Rolled Copper. —— (*J. Metals*, 1951, **3**, (7), 548–549).—Cf. *Met. Abs.*, 1950–51, **18**, 361.

[Discussion on a Paper by H. F. Walton on :] The Anode Layer in the Electrolytic Polishing of Copper. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 252).—Cf. *Met. Abs.*, 1950–51, **18**, 175.—G. T. C.

***Study of the Detection and Practical Importance of the Chemical Nature of Electrolytically Polished Metallic Surfaces [Copper, Zinc, and Magnesium].** P. A. Jacquet and M. Jean (*Rev. Mét.*, 1951, **48**, (7), 536–545; discussion, 545–546).—J. and J. review the factors involved in the electrolytic polishing of metallic surfaces and, in particular, consider Raether's suggestion that adsorption films occur on such surfaces with certain metals. They tested this hypothesis experimentally for Cu, Zn, and Mg electrolytically polished in a conc. soln. of H_3PO_4 and for Armco Fe in a phosphoric-chromic acid bath. Adaptations of well-tried colorimetric methods were used to detect the presence of P in the Cu, Zn, and Mg, and P and Cr together in Fe. The results showed that a certain amount of P was retained by the Cu, Zn, and Mg surfaces to an extent that could be evaluated numerically, in spite of the relative fewness of the number of specimens and the difficulties of the analysis. No trace of either P or Cr was found on the Fe surface. These results explain Raether's observations on the interpretation of electron-diffraction diagrams of electrolytically polished Zn and of the behaviour of Cu, similarly treated, towards air compared with the same metal electrodeposited *in vacuo*. It appears that, in the case of Cu, the compound contg. P is fixed by adsorption on the surface and can, therefore, be removed by sufficient washing. For Zn and, in particular, for Mg, a true chem. compound with the metal is possible and its removal is difficult. The possible presence of P in such surfaces must be taken into account when making studies of the phys. and physico-chem. properties of the surfaces. The passive state of Fe or steel is not due to a phosphate or chromate on the surface, but undoubtedly to a continuous and very thin film of oxide, which accounts for the friction and wear characteristics of these surfaces. 15 ref.—J. H. W.

***Method of Micro-Examination of Metallic and Petrographic Specimens at Elevated Temperatures by Means of a Heating Plate with Facilities for Gaseous Attack: Examples of Application [to Copper].** Jean R. Maréchal and Marcel Doucet (*Rev. Mét.*, 1951, **48**, (7), 561–566).—M. and D. have devised an apparatus for the micrographic examination of pure metals and alloys *in vacuo* at various temp. An essential part of the apparatus is a circular brass plate supporting a small elect. furnace and surrounded by a water-jacket to protect the objective against heat. The whole is so arranged that the specimen can be heated in a vacuum or in any atmosphere. The specimen is examined from below through a glass or Pyrex window. M. and D. give examples of the micrographic examination with this apparatus of American and French O.F.H.C. Cu, electrolytic Cu, Cu-Te and Cu-Ni-Fe alloys, and micaceous Cu at temp. up to 800° C. They detected a striped (acicular) structure on O.F.H.C. Cu at 600°–800° C. on heating and at 400°–460° C. on cooling, and also the appearance of parallel lines, similar to slip lines, on electrolytic Cu annealed *in vacuo* at 800° C. for 20 min. after electrolytic polishing and attack by persulphate.—J. H. W.

[Discussion on a Paper by R. H. Fillnow and D. J. Mack on :] The Isothermal Transformation of a Eutectoid Beryllium Bronze. —— (*J. Metals*, 1951, **3**, (7), 553–556).—Cf. *Met. Abs.*, 1950–51, **18**, 515.

***The Crystal Structure of Gallium Sulphide, Selenide, and Telluride.** Harry Kahn and Wilhelm Klingler (*Z. anorg. Chem.*, 1949, **259**, 135–142; *C. Abs.*, 1950, **44**, 5675).—Ga_2S_3 crystallizes in two modifications: (i) an α form with a zinc-blende-type lattice ($a = 5{\cdot}171$ Å.), which is stable at low temp., and (ii) a high-temp. modification, β, with a wurtzite structure ($a = 3{\cdot}678$, $c = 6{\cdot}016$ Å.); the transition temp. lies between 550° and 600° C. Ga_2Se_3 and Ga_2Te_3 crystallize only in the zinc-blende-type lattice. For Ga_2Se_3, $a = 5{\cdot}418$ Å.; for Ga_2Te_3, $a = 5{\cdot}874$ Å. A summary is given of the known compounds of the wurtzite and zinc-blende structures.

***Influence of the Support on the Crystallization of Very Thin Films of Gold.** Antoine Colombani and Gaston Ranc (*Compt. rend.*, 1951, **233**, (1), 46–48).—C. and R. extended their earlier researches into the influence of the support on the crystn. of very thin films of Au (*ibid.*, 1951, **232**, 1344;

Met. Abs., this vol., col. 4) to a large number of both cryst. and amorphous supports : KCl, KBr, $NaNO_3$, NaCl, Rhodoid, Plexiglas, and glass, using the same procedure as before. The new results confirmed the earlier ones: (1) *Plexiglas, glass, and Rhodoid.* Below 50 Å. the change in elect. resistivity is an increasing function of the time ; above this thickness the function diminishes ; in both cases it reaches a stable value. (2) *Crystalline supports.* The change in direction of the variation of the resistivity is at ~30 Å. C. and R. record a number of observations linking the resistivity of films of equal thickness with the nature of the support and illustrate these with three electron photomicrographs. All these results agree well with those of Appleyard and Lowell on alkaline deposits and with those of Andrade.—J. H. W.

***Electron-Diffraction Studies of Etched Gold Leaf Films.** W. V. Stauder and A. H. Weber (*J. Franklin Inst.*, 1951, **251**, (3), 351–357).—S. and W. studied a sample of etched Au leaf by electron diffraction and found extra lines, which they attribute to contamination by long-chain hydrocarbon compounds, and abnormal line intensities for the true Au lines, which they attribute to preferred orientation in the Au leaf. The experimental observations are reported in fair detail for the benefit of experimenters who wish to use Au leaf as a supporting medium in transmission electron-diffraction work.—H. J. A.

***The Tetragonality of the Sigma Phase in the Iron-Chromium System.** L. Menezes, J. K. Roros, and T. A. Read (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 71–74; discussion, 74).—Single-crystal specimens of sigma phase were prepared by isothermal transformation of coarse-grained 50 : 50 Fe–Cr ferrite at 775° C. In some cases single grains of ferrite transformed to single crystals of sigma phase which were then isolated by careful fragmentation of the brittle sample. Back-reflection Laue diffraction patterns were prepared; these revealed a four-fold symmetry axis and two two-fold axes orthogonal to it and to each other, showing that the sigma phase belongs to the tetragonal crystal system. In all samples crystal imperfections existed which led to diffuseness and multiplicity of the Laue spots.—P. T. G.

***The Crystal Structure of a Sigma Phase, FeCr.** David P. Shoemaker and Bror Gunnar Bergman (*J. Amer. Chem. Soc.*, 1950, **72**, (12), 5793).—A letter. Preliminary findings from an X-ray study of two single crystals ~0·1 mm. in size, isolated from σ-FeCr (46·5 at.-% Cr), are presented.—J. R.

***The Space Group of the σ-FeCr Crystal Structure.** B. Gunnar Bergman and David P. Shoemaker (*J. Chem. Physics*, 1951, **19**, (4), 515).—A letter. Among the indications from an extension of a previous investigation of the general arrangement of atoms, without regard to kind, in a sigma phase (*J. Amer. Chem. Soc.*, 1950, **72**, (12), 5793 ; preceding abstract) is that σ-CoCr and other σ alloys have the same holohedral structure as that reported by S. and B. for σ-FeCr.—J. R.

***X-Ray Study of the Sigma Phase in Various Alloy Systems.** Pol Duwez and Spencer R. Baen (*A.S.T.M. Symposium on the Nature, Occurrence, and Effects of Sigma Phase*, **1950**, 48–54 ; discussion, 55–60).—Cf. *Met. Abs.*, 1950–51, **18**, 172.

***Observations on Cleavage and Polygonization of Molybdenum Single Crystals.** Neng-Kuan Chen and Robert Maddin (*J. Metals*, 1951, **3**, (7), 531–532).—Laue back-reflection patterns were taken of the fracture surface of a single crystal of Mo which had been fractured by bending at room temp. They indicate that the cleavage plane is (100). On bending a single crystal through an angle of ~15°, at 2400° C., Laue photographs reveal that polygonization takes place. The rotation of the crystallites is about a [112] axis, which is in the slip plane (110) and 90° from the slip direction [111]. This checks with the microscopic observations of slip, and also confirms the general type of rotation in the plastic deformation of single crystals. 6 ref.—E. N.

***Magnetic Domain Patterns on Nickel Crystals.** Mikio Yamamoto and Takao Iwata (*Phys. Rev.*, 1951, [ii], **81**, (5), 887–888).—A letter. Complex magnetic powder patterns obtained from single crystals of Ni having a magnetic anisotropy const. <0 are reproduced and explained. After electropolishing, different patterns were found from those given by demagnetization in an alternating magnetic field. —P. C. L. P.

***The Preliminary Recrystallization of Metals [Iron and Nickel].** Hideki Hirata and Masaki Yanagisawa (*X-Sen* (*X-Rays*), 1949, **5**, 79 ; *C. Abs.*, 1950, **44**, 5774).—[In Japanese]. The structural changes brought about by annealing were examined by X-rays, specimens of Fe and Ni foils produced by rolling annealed sheets of electrolytic Fe and Ni 0·5 mm. thick being used. Laue patterns showed that the common axis of the fibrous structures in Fe foils, which had initially been situated parallel to the direction [211], changed over to coincide with the direction [110] on heating at ~420° C. ; while in Ni foils, its direction changed from [111] to [110] at 470° C., which is far below the recrystn. temp. of Ni.

***Secondary Recrystallization of Face-Centred Ni–Fe Alloys.** G. W. Rathenau and J. F. H. Custers (*Philips Research Rep.*, 1949, **4**, (4), 241–260).—The orientation of grains produced during secondary recrystn. of severely rolled Ni–Fe alloys has been shown to be different from that of the non-heat-treated alloy (cubic orientation). Several new orientations have been found. Experimental evidence is put forward to support the theory that normal secondary recrystn. is the grain growth of primary crystals with a high temp. of primary crystn. 14 ref.—D. K. W.

***Comparison of the Perfection of the Crystals of Primary and Secondary Recrystallization [of Ni–Fe Alloys].** A. Guinier and J. Tennevin (*Philips Research Rep.*, 1949, **4**, (4), 316–318).—The Ni–Fe alloys employed for the work by Rathenau and Custers (*ibid.*, p. 241 ; preceding abstract), have been examined using an X-ray-diffraction technique. The following results were obtained. Crystals formed by the secondary recrystn. of severely rolled Ni–Fe alloys were perfect. The experiments were unable to confirm a difference of perfection of lattice planes in primary and secondary crystals in Ni–Fe.—D. K. W.

[Discussion on a Paper by W. E. Seymour and D. Harker on :] Recrystallization Reaction Kinetics and Texture Studies of a 50 : 50 Nickel–Iron Alloy. —— (*J. Metals*, 1951, **3**, (7), 549).—Cf. *Met. Abs.*, 1950–51, **18**, 462.

***X-Ray Studies of Silver Bearings.** Shigeru Nishigaki (*X-Sen* (*X-Rays*), 1949, **5**, 80–84 ; *C. Abs.*, 1950, **44**, 5776).—[In Japanese]. Laue photographs of Ag bearings have been used for the detn. of the method of manufacture, i.e. whether electroplated, cast, centrifugally cast, die-cast, or pressed. The annealing temp. of Ag bearings was also determined by Laue and back-reflection photographs.

[Discussion on a Paper by H. T. Clark, Jr., on :] The Textures of Cold-Rolled and Annealed Titanium. —— (*J. Metals*, 1951, **3**, (7), 549–550).—Cf. *Met. Abs.*, 1950–51, **18**, 428.

[Discussion on a Paper by P. Duwez and J. L. Taylor on :] The Structure of Intermediate Phases in Alloys of Titanium with Iron, Cobalt, and Nickel. —— (*J. Metals*, 1951, **3**, (7), 551).—Cf. *Met. Abs.*, 1950–51, **18**, 462.

***The Crystal Structure of Uranium Silicide.** Georg Brauer and Heinz Haag (*Z. anorg. Chem.* 1949, **259**, 197–200 ; *C. Abs.*, 1950, **44**, 5675).—USi_2 has a cubic crystal lattice, the elementary cell having $a = 4{\cdot}053$ Å. and a mass content of 1·07 U atoms and 2·14 Si atoms. After considering many different atomic distributions, it appears that the only one consistent with the observed X-ray reflection intensities is that in which the U atoms occupy the corners of the elementary cube (0,0,0) and 2·14 Si atoms are statistically distributed in the face centres $\frac{1}{2}\,\frac{1}{2}\,0$, $\frac{1}{2}\,0\,\frac{1}{2}$, $0\,\frac{1}{2}\,\frac{1}{2}$ next to 1·07 U atoms.

***The Crystal Structure of V_3Co.** Pol Duwez (*J. Metals*, 1951, **3**, (7), 564).—Samples of V_3Co, prepared from powders of high-purity V, and Co contg. Ni 0·5, Cr 0·1%, and traces of Si and Fe, were homogenized *in vacuo* for 10 days at 800° C. Powder-diffraction patterns were then obtained using Cu$K\alpha$ radiation. All the results appear to show that V_3Co has a β-W structure : space group O_h^3—$Pm3n$; 2 Co in (a) : $000, \frac{1}{2}\frac{1}{2}\frac{1}{2}$; 6 V in (c) : $\frac{1}{4}0\frac{1}{2}, \frac{1}{2}\frac{1}{4}0, 0\frac{1}{2}\frac{1}{4}, \frac{3}{4}0\frac{1}{2}, \frac{1}{2}\frac{3}{4}0, 0\frac{1}{2}\frac{3}{4}$; (khl) reflection present only if $l = 2n$. Each Co atom is surrounded by 12 V atoms at 2·61 Å. ; each V atom is surrounded by 2 V atoms at 2·34 Å., 4 Co atoms at 2·61 Å., and 8 V atoms at 2·86 Å. 3 ref.—E. N.

[Discussion on a Paper by D. C. Jillson on :] Production and Examination of Zinc Single Crystals. —— (*J. Metals*, 1951, **3**, (7), 550–551).—Cf. *Met. Abs.*, 1950–51, **18**, 462.

[Discussion on a Paper by D. C. Jillson on :] An Experimental Survey of Deformation and Annealing Processes in Zinc [Single Crystals]. —— (*J. Metals*, 1951, **3**, (7), 541–544).—Cf. *Met. Abs.*, 1950–51, **18**, 463.

***The Nucleation of Cast Metals at the Mould Face.** J. A. Reynolds and C. R. Tottle (*J. Inst. Metals*, 1951–52, **80**, (2), 93–98).—The nucleation of some cast metals and alloys effected by applying metal powders to the face of a standard mould, has been investigated by comparing the grain-size of ingot surfaces produced with and without the coating. In general, powdered metals having the same crystallographic form as the metal being cast are effective as nucleating agents, provided that the degree of lattice " misfit " does not exceed 5%, or in some cases 10%. This simple mechanism is found to apply to the solidification of Sn, Pb, Zn, Sb, Al, Mg, and Cu. Where the cast metal readily formed an oxide film, reduction of this film to the metal was found to occur when some coatings were applied, giving rise to simple homogeneous nucleation by the particles produced. Homogeneous and heterogeneous nucleation of alloys can take place under the same conditions as with pure metals, the crystal structure of the solid first formed from the melt being the ruling factor. Reduction of oxide films also occurred, and in the case of an 18 : 8 austenitic steel resulted in isomorphous nucleation of the δ solid soln. by Cr particles reduced from the film.—Authors.

[Discussion on a Paper by B. H. Alexander and F. N. Rhines on :] Dendritic Crystallization of Alloys. —— (*J. Metals*, 1951, **3**, (7), 561–563).—Cf. *Met. Abs.*, 1950–51, **18**, 518.

Progress of the Physics of Solids. Frederick Seitz (*J. Franklin Inst.*, 1951, **251**, (1), 156–166).—A review. The study of cryst. solids has passed through three phases corresponding to interest in : (i) macroscopic symmetry properties, (ii) lattice theory, and (iii) quantum-mech. description of a perfect crystal. A fourth and highly fruitful phase which appears to be reaching a climax at the present time deals with the theory of imperfections in real crystals. S. lists six primary imperfections, viz : (*a*) phonons or quantized waves of lattice vibration ; (*b*) vacant lattice sites and interstitial atoms ; (*c*) free electrons and holes ; (*d*) excitons or systems of electron + hole generated in insulators by incident radiation ; (*e*) impurity atoms ; (*f*) dislocations : Three secondary or transient imperfections are suggested as (*g*) light quanta ; (*h*) charged particles ; and (*i*) neutral particles. S. stresses the importance of considering the interaction of, and the inter-relation between, all these imperfections.—H. J. A.

***Deformation Textures in Face-Centred Cubic Metals.** E. A. Calnan and C. J. B. Clews (*Phil. Mag.*, 1950, [vii], **41**, (322), 1085–1100).—In this treatment of the deformation of polycryst. f.c.c. metals, the constraints applied by neighbouring grains are used to explain how some slip planes can withstand a resolved stress greater than the critical until other planes reach their critical stresses. A tensile-strength/elongation curve in agreement with experiment is obtained for polycryst. materials similar to that obtained by Taylor (*J. Inst. Metals*, 1938, **62**, 307 ; *Met. Abs.*, 1938, **5**, 425). The theory explains qual. the textures obtained by tension, compression, rolling, and wire-drawing. The differences found between Cu, Ag, and Au on the one hand and Al on the other, may be related to the lower m.p. of the latter.—P. C. L. P.

Theory of Annealing Textures. Paul A. Beck (*J. Metals*, 1951, **3**, (6), 475–476).—A discussion of the cause of the relative orientation, corresponding to a rotation of 30°–40° around a [111] axis, between a matrix with a strong single-orientation texture and grains growing in the matrix, which is found to occur both in recrystn. and in coarsening " secondary recrystn." It is considered that the oriented growth theory, rather than that of oriented nucleation, affords the more promising explanation of the phenomenon. The elementary processes involved in atomic movements necessary for grain-boundary migration seem likely to be the same as those involved in grain-boundary self-diffusion, the rate of which appears to be a function of the relative orientation of the grains meeting at the boundary. 12 ref. —E. N.

***Anomalous Diffraction Effects in X-Ray Patterns of Ageing Polycrystalline Alloys.** A. M. Elistratov, S. D. Finkel'shtein, and A. I. Pashilov (*Doklady Akad. Nauk S.S.S.R.*, 1949, **68**, (6), 1017–1020 ; *C. Abs.*, 1950, **44**, 5698).—[In Russian]. A " coarse-grained " (0·02 mm.) X-ray method is described to replace the single-crystal techniques generally used. The Laue method was used with flat or cylindrical films, and the wire specimens examined had a dia. of 0·3–1·0 mm. Two-dimensional diffraction effects appeared in a Cu–2·02% Be alloy after ageing for 1 hr. at 180° C. and disappeared after ageing for 1 hr. at 350° C. Characteristic extra streaks were obtained if the incident-beam direction was parallel to the plane of a Guinier–Preston zone. In Cu–Be alloy they were never observed near a (111) spot. In the immediate vicinity (0·3°–0·5°) of the Laue spot, the intensity of the extra spots could be greater than the spot itself if a low voltage and a long-wave characteristic radiation were used. Exposures of 1–2 hr. were used in this case. Much weaker " white streaks " were also obtained if long exposure times were used at high voltages. Two-dimensional effects were also found in a Ag–7·5% Cu alloy aged for 3 min. at 230° C. and in Fe–Ni–Al magnetic alloy aged for 10 min. at 750° C. In Al–Ag and Cu–Be alloys very weak one-dimensional diffraction effects were observed in the form of small, thin circles, or thin, short, slanting lines coming from extra two-dimensional diffraction spots, or as long, thin lines going through extra spots. On changing the wave-length they either became displaced relative to the extra spots, remained almost parallel to their previous position, or disappeared. " Two-crystal " diffraction effects, weak or white streaks, were obtained from Cu–Be alloy aged for 1 hr. at 220° C. by 12–14 hours' exposure. With change in wave-length they either disappeared or sharply changed direction. This effect was not observed with grains >0·3–0·4 mm., and was observed best with a non-uniform grain-size.

An X-Ray Technique for the Study of Substructures in Materials. P. Gay and P. B. Hirsch (*Acta Cryst.*, 1951, **4**, (3), 284–285).—[In English]. A letter. The use of a spot-counting technique for determining the size of substructures within crystal grains in a metal or alloy, is recommended. The angular range of orientation can also be determined. The method has been applied to substructures in annealed α/β brass, in an Ag–Al alloy subjected to fatigue testing, and in a martensitic steel. Quant. data are quoted. From the data for the steel (which gave the amount of material of any one orientation) it was deduced that 8 out of a possible 24 equivalent habit planes of martensite were occupied on the average in any one grain.—R. W. C.

Study of Imperfect Crystals by Powder Photographs. A. J. C. Wilson (*Research*, 1950, **3**, (9), 394–399).—The breadth of an X-ray powder-photograph line is made up of three components—an intrinsic breadth, β_i, which depends only on the texture ; a trigonometrical factor $2 \tan\theta$; and a broadening due to imperfections in the experimental arrangements. Broadening of a line due to small particle size is considered, and it is shown that the breadth as measured on the film, β_f is related to β_i by $\beta_f/\beta_i = 2 \tan\theta$ for a crystal of const. thickness T. For crystals of varying thickness an apparent particle size ε may be defined by : $\varepsilon = \lambda/\beta_f \cos\theta = V^{-1}\int_v T\,dV$, where V is the crystal vol. A discussion is also given of line broadening due to " mistakes " in the sequence of unit cells within the crystal, the unit cells having different scattering power, and an expression for film breadth β_f is derived. The size of the unit cell may also vary in strained crystals, and broadening may result from the variation. A bent crystal lamella is considered, and it is shown that the line broadening is proportional to tanθ; a sufficiently thin lamella, however, does not give distortion broadening. 72 ref.—F. A. F.

Orientation of Crystals of Cylindrical Form. A. J. Goss (*Research*, 1951, **4**, (6), 292–293).—A letter. For a crystal of cylindrical form, crystallographic orientation is usually defined solely by the angle ψ between one particular crystallographic axis and the cylindrical axis. However, G. shows that the probability of crystals between ψ_1 and ψ_2 is proportional to $\sin\left(\frac{(\psi_1+\psi_2)}{2}\right)$, and there will therefore be an apparent preference for orientations of higher values of ψ, in spite of a uniform distribution. Workers in the field of cylindrical crystals who have relied on the use only of ψ as descriptive of the orientation may find their results in need of further examination.—F. A. F.

***Optical Methods in X-Ray Analysis. I.—The Study of Imperfect Structures.** C. A. Taylor, R. M. Hinde, and H. Lipson (*Acta Cryst.*, 1951, **4**, (3), 261–266).—[In English]. An extension to two dimensions of the opt.-model method of constructing a projection of the reciprocal lattice of an imperfect structure, for comparison with experimental results. A hole grating is made in cardboard to represent a projection of the imperfect structure and used in conjunction with a simple opt. train. Points of practical technique are covered. The method has been applied to mixed cubic and h.c.p. structures, and comparison is made with experimental results for Co, with good agreement. The method is applied also to the problem of order in $AuCu_3$. It is possible, by comparing the resulting reciprocal lattice projections with experimental ones, to deduce that order developed by a gradual increase of short-range order throughout the crystal, rather than by the growth of domains of perfect long-range order. Limitations of the method and some general principles of its appn. are given.—R. W. C.

***Theory of Displacements in One-Dimensional Atom Rows. I.—Periodically Ordered Displacements.** Albert Kochendörfer and Alfred Seeger (*Z. Physik*, 1950, **127**, (5), 533–550).—A theoretical paper designed to study the state of plastically deformed crystals. A single row of atoms is assumed, each atom being elastically coupled to its neighbours. Along this row there is a periodic displacement. All displacements can be expressed in terms of atom number p and time t. The type of soln. depends on the initial conditions. These displacements give rise to contractions of a relativistic character, in which the velocity of sound plays the part of a limiting velocity. The stability conditions in finite rows of atoms are discussed.—A. C.

Statistical Theory of Superlattices of the Type *AB* in a Quadratic and a Simple Cubic Lattice. Y.-C. Hsu (*Chinese J. Physics*, 1949, **7**, 207–229; *Physics Abs.*, 1950, **53**, 869).—[In English]. Wang's generalized theory of superlattices (*Phys. Rev.*, 1945, [ii], **67**, 98; *Met. Abs.*, 1945, **12**, 249) is applied to binary alloys. Nearest-neighbour interactions for 2 shells surrounding each atom are considered, and the number of A and B atoms are not required to be equal. In the special case of equal numbers, results close to those of Bethe's second approximation are obtained; for the simple cubic lattice the discontinuity in C_v at the critical temp. is 2·07 k/atom instead of 1·94 k.

***Theory of Co-Operative Phenomena.** Ryoichi Kikuchi (*Phys. Rev.*, 1951, [ii], **81**, (6), 988–1003).—A new math. method is developed for the calculation of the entropy of order–disorder systems. The method is applied to simple cubic and to f.c.c. lattices, and in both cases gives a lower transition temp. and a larger sp. heat near the transition temp. than previous approximations.—P. C. L. P.

A Note on the Eigenvalue Problem in Crystal Statistics. Y. Nambu (*Progress Theoret. Physics*, 1950, **5**, (1), 1–13; *Physics Abs.*, 1950, **53**, 869).—Onsager's problem is formulated in terms of the operators of second quantization. A Fourier expansion permits the derivation of known results for the quadratic Ising lattice, a honeycomb, and the Kramers–Wannier screw-constructed lattices.

***Statistics of Solid Solutions and Ferromagnetics.** Eleonore Trefftz (*Z. Physik*, 1950, **127**, (4), 371–380).—The Bethe–Ising model for solid soln. and ferromagnetics is used to obtain series for the partition function for simple, f.c.c. and b.c.c. structures at low and high temp. From this function the energy and sp. heat are derived and compared with measurements of the sp. heat of Ni, Fe, and β-brass. An explanation is given of differences between observed and theoretical values of the sp. heats in the anomalous region. —A. C.

The Determination of the Energy of a Ferromagnetic Domain Boundary. Harvey Kaplan (*Phys. Rev.*, 1951, [ii], **81**, (5), 885–886).—A letter. Two corrections to the theoretical energy of a domain boundary are considered: (i) the effect of an additional term in the crystal anisotropy energy; and (ii) the effect of non-uniform deformation of the wall around the crystal circumference. Both effects are shown to be small, of the order of 3%.—P. C. L. P.

***Interaction Between the *d*-Shells in the Transition Metals.** C. Zener (*Phys. Rev.*, 1951, [ii], **81**, (3), 440–444).—Z. adopts the principles that: (i) at least for elements of Group V or beyond, the spin correlation between electrons in an incomplete d-shell remains the same in the solid as in the gaseous state; (ii) in contrast to alternative theories, the exchange integral between the d-shells of adjacent atoms always has the same sign as in the H_2 molecule irrespective of the separation of the atoms; and (iii) the spin of the electrons of an incomplete d-shell is strongly coupled to the spin of the conduction electrons. These principles explain why elements of Group VI and neighbouring Groups, which have very large net d-shell spins, crystallize in lattices allowing antiferromagnetism. The b.c.c. lattice is found which can satisfy the antiferromagnetism requirement that neighbouring atoms have antiparallel spins, whereas this is not possible for the f.c.c. and h.c.p. lattices. These considerations are extended to explain the complex modification of W and structurally analogous alloys. The first and third effects both tend towards the alignment of spins and therefore to ferromagnetism. The conditions governing the occurrence of ferromagnetism at the expense of antiferromagnetism may be reduced to a relatively simple criterion. The theory is extended to alloys, and it is shown why the Heusler alloys become ferromagnetic. Ferromagnetism does not arise in ionic salts, such as $MnCl_2$, because there are no conduction electrons.—P. C. L. P.

Métallographie: Les diagrammes d'équilibre par questions et réponses. (Glazunov). See col. 249.

Metallographie: Grundlagen und Anwendungen. (Mies). See col. 250.

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5 — POWDER METALLURGY

***Rate of Sintering of Copper [Powder] Under a Dead Load.** F. N. Rhines and H. S. Cannon (*J. Metals*, 1951, **3**, (7), 529–530).—Cu powder (0·1, 31·0, and 68·9% of mesh no. +150, −150 to +325, and −325, resp.) was compacted under pressures of 12,500 lb./in.² into cylindrical compacts ½ in. dia. × ½ in. high. These were then sintered in H at 1000° C. under dead loads of 0–165 lb./in.², and the progress of sintering was observed by d measurements. The appn. of the static load was found to accelerate the rate of sintering in such a way that a given load induces the same proportional increase in rate for all times of sintering. Thus, with low loads at least, loading does not introduce any new mechanism into the sintering process, and the applied force is merely added to the forces (surface tension) normally present in sintering. This interpretation is fortified by the fact that metallographic examination of the sintered specimen shows no appreciable change in shape of the pores as a result of loading, i.e. no flattening, and by the finding of radial shrinkage in all samples sintered with loads of $<$10·5 lb./in.^{2}. Therefore, sintering under a load is markedly similar to creep under a fixed load, in that the stress required to accomplish a given degree of densification is a linear function of the logarithm of the sintering time.—E. N.

***Sintering of Metallic [Copper] Powders.** V. I. Likhtman (*Doklady Akad. Nauk S.S.S.R.*, 1950, **71**, (2), 323–325; *C. Abs.*, 1950, **44**, 5773).—[In Russian]. Micrography of a powder of oxidized electrolytic Cu, pressed at 300° C. under graphite, showed tiny bridges of metal connecting grains where the metal had not been brought into immediate contact by the compression. At 400° C. the bridges are broader, with recrystn. beginning to appear within the grains. The formation of the bridges is due to the particularly high mobility of the Cu atoms produced in the reduction of the oxide, at a temp. (300° C.) at which atoms of the metal itself do not possess appreciable mobility. The highly mobile atoms migrate preferably to points where the metal grains are closest to each other. Evaporation of metal, which is possible and which was shown to take place at highly active points at as low a temp. as 300°–400° C., and its condensation at less active points, may also play a part in the formation of the bridges.

***The Consolidation of Metal [Powders by Small Electric Currents].** E. Schwarz-Bergkampf (*Berg- u. hüttenmänn. Monatsh. Montan. Hochschule Leoben*, 1951, **96**, (5), 116–118).—Fe powder of average particle size 0·1 mm. dia. was moulded under a pressure of 4 tons/cm.^{2} into bars 100 × 10 × 7 mm., some of which were then subjected to longitudinal currents of 1–3·5 amp. for 2 min., only the highest of such currents being sufficient to produce evidence of slight warmth within the bar. The bars underwent transverse tests, in which the load was applied at a distance of 80 mm. from one end, and average breaking loads (in kg.) of 1·54 (± 13·5%), 1·84 (± 9·6%), and 1·61 (± 2·6%) were obtained from bars which had been subjected to 0, 1, and 3·5 amp., resp. In spite of the somewhat erratic effect produced by the current of 3·5 amp.—which is attributed to the current being high enough to cause arcing at places of poor contact within the material—it is concluded that improved cohesion of the powder particles results from the appn. of small elect. currents. 6 ref.—E. N.

***The Sintering Velocity of Powdered Metals. I.—A Mixture of Nickel and Copper Powders.** Shiro Ogawa, Joji Mizuno, Tokutaro Hirone, and Shigeo Ogi (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 780–786).—[In English]. The process of sintering mixtures of Ni and Cu powders (passing a 300-mesh sieve) was studied by the X-ray back-reflection method and by thermomagnetic analysis, samples being sintered for 10–800 min. at 600°, 700°, 800°, and 900° C., under pressures of 9–28 tons/in.^{2}. Sintering velocity increases initially with increase of pressure, and thereafter attains and maintains a const. value at 20 tons/in.^{2}. 800° C. appears to be a favourable sintering temp. The activation energy of intergranular diffusion of Cu and Ni, calculated from the results, is 25,500 cal./mole, which differs from the value 35,500 cal./mole derived from measurements of the diffusion coeff.—J. S. G. T.

***The Structure of the Carbide System TiC–TaC–WC.** (Nowotny, Kieffer, and Knotek). See col. 183.

***Particle Agglomeration in Tungsten Metal Powder.** Bernard Kopelman and C. C. Gregg (*J. Phys. Colloid Chem.*, 1951, **55**, (4), 557–563).—K. and G. used the Photelometer to determine the apparent particle size of W powder, prepared by reducing WO$_3$ with H at 850° C., in eleven liquids of different dielectric const. (D); the powder had an average dia. of 4·2 μ as measured by an air-permeability method. It was found that, over a wide range of D, the apparent particle size is inversely proportional to $D^{1/4}$. Agglomeration is complete when $D < 2$ and zero when $D > 25$. A metal powder of average dia. 4·2 μ would be expected to behave in air as if having an average particle dia. of $\sim$10 μ. 5 ref. —D. M. D.

Powder Metallurgy. J. S. Campbell (*Eng. J.*, 1950, **33**, (1), 10–12).—A brief review.—N. B. V.

Design in Powder Metallurgy. [—I.–II.] H. W. Greenwood (*Metal Ind.*, 1951, **79**, (1), 3–5; (2), 31–33).—G. gives a critical and detailed discussion of the basic principles of design of the parts, components, dies, and tools used in powder metallurgy and outlines some recently developed techniques. —J. H. W.

6 — CORROSION AND RELATED PHENOMENA

***Corrosion of Aluminium and Aluminium Alloys by Citric Acid and Citric Acid–Salt Solutions.** J. M. Bryan (*J. Sci. Food Agric.*, 1950, **1**, (3), 84–85; *C. Abs.*, 1950, **44**, 5787).—The corrosion rate of 4 grades of Al (99·25–99·993%), Al–Mn (1·78%) and Al–Mg–Mn (1·78–0·45%) alloys by 0·1 *N*-citric acid (I) under static conditions is approx. doubled per 10° C. rise in temp. Addn. of 0·1 *N*-NaCl (II) produces pronounced rate increases at 25° C., but has little effect between 60° and 95° C. At 95° C. the NaCl has little effect. The corrosion rates of 99·25% Al and of the two alloys in (I) are only slightly increased by agitation. In (II), agitation shows a pronounced effect.

***The Role of Coating Films Produced in a Chemical Manner on the Corrosion of Iron and Aluminium.** Gy. Bácskai (*Acta Univ. Szeged., Chim. et Phys.*, 1948, **2**, 139–144; *C. Abs.*, 1950, **44**, 5785).—[In English]. Al plates contg. Al 99·5 and Si, Zn, and Cu varying from 0·5 to 0·05% were treated: (1) by the method of Miyata with Na; (2) by Jirotka's method; (3) by the MBV process (waterless Na and K$_2$Cr$_2$O$_7$ at 90° C. for 15 min. then washing and immersing in a soln. contg. 4 g. KMnO$_4$/l.); (4) similar to (3) but omitting the immersion in KMnO$_4$; and (5) without pretreatment. Untreated Al plates showed a relatively rapid increase in corrosion, then a slow decline. Photomicrographs showed more or less uniform corrosion. Plates treated as under (1) showed at the beginning two max. then a horizontal line; under the microscope uniform corrosion could be observed. Such plates were well protected against the action of dil. acids. Plates treated as under (2) showed gradual increase to rather higher corrosion reaction velocity than the former. Photomicrographs showed marked but relatively uniform corrosion. Plates treated as under (3) showed a very rapid increase of the corrosion curve, which changed into a slight decrease. It was striking to note that Al plates as under (4) behaved quite differently to those under (3), since they showed a very low max. in the beginning and then the curve became almost horizontal at a low value. The microscopic picture of the plate showed the formation of a very uniform layer which consisted of a mixture of Al$_2$O$_3$ and Cr$_2$O$_3$, the latter compound filling up the small cavities not covered by the former. This layer seemed to protect the Al very well against further corrosive action. 10 ref.

An American Long-Time Test of the Corrosion-Resistance of Steel and Light-Alloy Coal Hoppers [and Wagons]. Jean Reinhold (*Rev. Aluminium*, 1949, (156), 201–205).—Tests were begun in America in 1932 to compare the behaviour of steel and light-alloy coal hoppers. Four fixed hoppers were constructed of 3S, 4S, and 17S Al alloys, and a fifth was built partly of various steels (Cor-Ten, Man-Ten, and Cu-bearing steel) and partly of Al alloy 52S. The hoppers were mainly riveted, but there were also some welds. The hoppers were filled with different grades of coal and left in the open air for 12 years. At the end of that time it was found that the light-alloy hoppers had stood up well and required no repair, whereas the steel hopper (which had been exposed only 9 years) had been badly attacked, particularly at the bottom. Tests on American railways showed that light-alloy panels of coal wagons last four times as long as steel ones, at the same time affording a saving in weight of 60% of the steel panels. The increase in payload thereby effected soon offsets the initially greater cost of the light alloy.—N. B. V.

***The Kinetics of the Oxidation of Cobalt.** Earl A. Gulbransen and Kenneth F. Andrew (*J. Electrochem. Soc.*, 1951, 98, (6), 241–251).—The problem is complex, since two or more oxides are formed and the metal itself has a phase transition from h.c.p. to f.c.c. between 300° and 500° C. The vacuum-microbalance method was used to study the effect of time, temp., pressure, pretreatment, &c., on the rate of the reaction below 700° C., and thermodynamic calculations were made on: (*a*) the equilibrium between the metal, the several oxides, and O; (*b*) the equilibrium between the several oxides and the metal; and (*c*) the reduction of CoO and Co_3O_4 with H_2. The reaction of cold-worked Co (h.c.p. form) and annealed Co (largely f.c.c. form) was studied over the range 200°–750° C., and it was found that coloured oxides were formed at all temp. except 200° C. In the range 600°–700° C. the oxide films cracked on cooling, whereas below 600° C. they appeared adherent. The cold-worked samples oxidized far more readily than the annealed ones. This is attributed to the crystal structure of the Co. The rate of reaction was found to increase abruptly between 600° and 700° C. The rate data are interpreted in terms of the transition-state theory, and the energy and entropy of activation of the reaction were calculated. It is shown that Co oxidizes more readily than Fe, Ti, or Be and that it cannot be classed as an oxidation-resistant material. A modified form of the parabolic rate law fits the data. Activation energies of 23,400 and 22,200 cal./mole were calculated for the cold-worked and annealed samples, resp. The effect of contamination of the atmosphere was found to be important only at low temp. The solid-phase reactions occurring in the oxide were studied using the high-temp. electron-diffraction camera. The results show that the reaction: $Co_3O_4 + Co \rightarrow 4CoO$ occurs readily at 450° C. The effect of length of oxidation and temp. on the solid phase reaction was determined. A mechanism of the oxidation of Co and of the solid phase reactions occurring with the oxide is suggested, based on the diffusion of cations and electrons through the oxide from the metal to the gas–oxide surface. 35 ref.—G. T. C.

***Low-Temperature Oxidation of Copper. II.—Reaction-Rate Anisotropy.** T. N. Rhodin, Jr. (*J. Amer. Chem. Soc.*, 1951, 73, (7), 3143–3146).—Cf. *ibid.*, 1950, 72, 5102; *Met. Abs.*, 1950–51, 18, 429. The oxidation rate of single-crystal Cu surfaces exposed to pure O at 78°–323° K. was highest for the least densely packed (100) face and smallest for the most densely packed (111) face. The rate anisotropy is adequately described by the treatment of Cabrera and Mott (*Rep. Progress Physics*, 1948–49, 12, 163; *Met. Abs.*, 1949–50, 17, 422)—though the phys. significance of their findings is considered ambiguous—and is consistent with the theoretical requirements of Frank and van der Merwe (*Proc. Roy. Soc.*, 1949, [A], 198, 205; *Met. Abs.*, 1949–50, 17, 573) for the existence of orientated overgrowths; according to the latters' model, pseudomorphic oxide structures may account for the oxidation-rate anisotropy.—J. R.

***Investigations Into Tropic Proofing of Electrical Materials, 1943–46. V.—The Corrosion of Copper Wire at D.C. Potential in Contact with Electrical Insulating Materials.** R. J. Meakins (*Australian J. Appl. Sci.*, 1950, 1, (1), 128–132).—Cf. *ibid.*, p. 80; *Met. Abs.*, this vol., col. 202: Parts II.–IV. are of no met. interest. Water absorption is the main cause of corrosion. Of 18 different insulating materials, alkyd resin varnish, some natural waxes, cellulosic sheet materials (except kraft paper), and wood-flour-filled phenolic resin or paper-base impregnated sheet, cause considerable corrosion. For testing, the samples were held at 90% relative humidity and 50° C., and a potential of 250 V. D.C. was applied across the sample by parallel Cu wires ~1·5 in. apart, for periods up to 10 days.—S. M.

***Stress-Corrosion of Cast Bronzes Evaluated.** J. T. Clenny (*Iron Age*, 1951, 167, (25), 85–89).—Tests of 10 different cast Cu alloys under stress in an NH_3 atmosphere indicate that 10% Al bronzes offer the best resistance. Four different methods of testing were used, including the detn. of stress-corrosion endurance limits and the % loss of U.T.S. due separately to stress and corrosion.—J. H. W.

***The Corrosion of Lead.** Rolf Uggla (*Svensk Kem. Tid.*, 1949, 61, 278–282; *C. Abs.*, 1950, 44, 5786).—[In German]. Experiments were performed on the corrosion of Pb in water in the presence of CO_2 with the object of confirming the corrosion theory of Kauko (*Trans. Chalmers Univ. Technol. Gothenburg*, 1947, (61)). Pure CO_2 at 1 atm. was passed through a water–Pb mixture at 25° C. The equilibrium Pb^{++} concentration was $\sim 10^{-5}$ g.-ion/l. and the *p*H was ~4. No precipitate of $Pb(OH)_2$ or $PbCO_3$ was visible on microscopic examination. A hypothetical corrosion diagram was drawn for the system $Pb–CO_2–H_2O$. The $K_{s.p.}$ of the reaction $H_2CO_3 = H^+ + HCO_3^-$, at 25° C. was determined as $3{\cdot}69 \times 10^{-7}$ by a potentiometric method and $3{\cdot}62 \times 10^{-7}$ by conductance measurements. Various preparations of $PbCO_3$ gave values at 25° C. of $\sim 10^{-13}$ for the solubility product.

***Anodic Corrosion of Lead in H_2SO_4 Solutions.** J. J. Lander (*J. Electrochem. Soc.*, 1951, 98, (6), 213–219).—Pure Pb was treated anodically in H_2SO_4 soln. at various const. potentials in order to attempt to obtain fundamental information which might be applied in regard to the corrosion of the positive grid in the Pb–acid battery. At potentials below those required for PbO_2 formation, a layer of tetragonal PbO is formed near the Pb. $PbSO_4$ is formed as an outer layer, probably as the result of a reaction between H_2SO_4 and the PbO layer. The potential range investigated was between 1·0 and −0·1 V. and in this range it was found that the corrosion rate increases with increase in temp. and with decreasing acid concentration. The weight-loss data are generally linear with time on a log/log plot. At potentials from 1·20 to ~1·50 V. a protective film is formed, presumed to be PbO_2. 10 ref.—G. T. C.

***Some Preliminary Studies of Positive Grid Corrosion in the Lead–Acid Cell.** J. J. Lander (*J. Electrochem. Soc.*, 1951, 98, (6), 220–224).—Preliminary experiments indicate that PbO_2 and Pb undergo the solid-phase reaction $PbO_2 + Pb \rightarrow 2PbO$, and a mechanism involving the Volta effect is proposed to explain this. It is suggested that a contact potential between Pb and PbO_2 may be an important part of the cell potential and explain self-discharge and the discharge reaction of the positive plate. It is further shown that an intermediate film of PbO_2 is formed at the positive plate which does not cycle to $PbSO_4$. For this reason it retains its compact nature and may in a large measure be responsible for slowing down grid corrosion. Grid growth appears to be proportional to the depth of corrosion after an initial period, the initial period and the rate of growth probably being characteristic of the tensile strength or creep-resistance of the metal. Electrolyte concentration and time of cycle are shown to be important factors in corrosion under cycling conditions in const.-concentration acid. These effects are discussed in terms of a cycling cell. Certain problems concerning the negative plate in connection with the concentration effect are briefly reviewed. 7 ref.
—G. T. C.

***Some Aspects of the Corrosion of Tinplate by Prunes.** V. W. Vaurio (*Corrosion*, 1950, 6, (8), 260–267).—Experimental evidence is presented which indicates that Sn corrodes at a const. rate without the evolution of measurable quantities of H, and that H evolution in canned prunes is related directly to the soln. of the Fe of the steel base. Increased Sn coating weight results in increased pack life, and experiments are described which illustrate that Sn affords cathodic protection to steel in contact with prunes. 11 ref.
—W. A. M. P.

***The Mechanism and Rate of Dissolution of Titanium in Hydrofluoric Acid.** M. E. Straumanis and P. C. Chen (*J. Electrochem. Soc.*, 1951, 98, (6), 234–240).—Ti dissolves rapidly in HF according to the equation: $2Ti + 6HF \rightarrow 2TiF_3 + 3H_2$. The formation of the trifluoride was confirmed by titration with $KMnO_4$ soln., and by measuring the vol. of H evolved. The rate of dissoln. of Ti in HF of concentrations between 0·02N and 0·2N is given by the empirical equation : $V = 459N^{2/3}$ mm.^{3}/cm.^{2}/min. For acid concentrations between 0·2N and 10N the equation becomes $V = 88 + 473N$ mm.^{3}/cm.^{2}/min. The rate of dissoln. of the Ti was increased by adding metal salts to the acid, the rate of increase being in the order : Fe^{2+}, Ni^{2+}, Ag^{+}, Cu^{2+}, Au^{3+}, Pt^{4+}, i.e. nearly in the order of decreasing overvoltage of the corresponding metals. Mg^{2+} was found to be inactive and Pb^{2+} hindered the dissoln. reaction. As the concentration of the HF varied from 0·0125N to 0·2N, the open-circuit potential of the Ti became more anodic, but with concentrations higher than 0·2N the potential remained practically const. and had the value −0·768 V. (H scale). A visible greyish-blue film was formed on the Ti in acid of concentrations below 0·5N. It is suggested that the dissoln. of Ti in HF is probably an electrochem. process, based on the activity of local elements. 19 ref.—G. T. C.

***Electrochemical Behaviour of Zinc and Steel in Aqueous Media.—II.** R. B. Hoxeng (*Corrosion*, 1950, **6**, (9), 308–312; discussion, 312).—Cf. *ibid.*, 1949, **5**, 330; *Met. Abs.*, 1949–50, **17**, 721. Experimental data are given to show that constituents present in natural waters have a pronounced effect on the electrochem. behaviour of Zn. It is concluded that the anodic or cathodic behaviour of galvanized steel in any water will depend on the equilibrium of the various constituents and cannot be determined from the amount of any one constituent.—W. A. M. P.

***Corrosion Tests with Zinc Alloy Die-Castings in Damp Gypsum Plaster.** L. A. J. Lodder and S. Baumgarten (*Metallurgia*, 1951, **43**, (260), 273–279).—The corrosion of Zn alloy die-castings, e.g. elect. conduit fittings, through contact with damp wall materials and methods of overcoming the corrosion have been fully investigated. Details of the results are described and illustrated. It has been found that the best protection is afforded by a chromate (Cronak) treatment. It is suggested that, in addn. to the chromate treatment, the castings be coated with an organic coating, e.g. a stoved phenol-formaldehyde enamel or bitumastic paint, specifically to protect the chromate coating.—F. M. L.

***The Oxidation of Heat-Resistant Alloys in the Presence of Foreign Oxides.** J. L. Meijering and G. W. Rathenau (*Philips Tech. Rev.*, 1951, **12**, (8), 213–220).—The oxidation of some Cu, Fe, and Ag-base alloys has been studied with a view to determining the mechanism by which alloys, which are normally protected against corrosion by a dense oxide skin, undergo accelerated oxidation when brought into contact with foreign oxides (MoO_3). The degree of oxidation of wires heated to various temp. in air, with and without contact with MoO_3, was determined by a simple tensile test. Discontinuities at a particular temp. in the oxidation rate/temp. curve (as in the case of an Al bronze) were found to agree with the formation of eutectics between the oxides, and the increased mobility of metal and O ions in this liquid is considered to be the cause of increased oxidation rate at certain temp. Increased rates at temp. below the lowest-m.p. eutectic are attributed to absorption of Mo atoms in the oxide skin lattice, with accompanying greater freedom of movement of electrons and metal or O ions (as in the case of a 25 : 75 Cr–Fe alloy). A third mechanism involved in the case of Ag is the formation of a complex oxide (having a lower dissociation pressure than the precious metal oxide and hence stable at atmospheric pressure) in which case, the two former mechanisms can operate.—D. K. W.

Corrosion of Metals in Fluorine and Hydrofluoric Acid. G. C. Whitaker (*Corrosion*, 1950, **6**, (9), 283–285).—A brief review based on experience gained in the use of metal equipment for the prodn., handling, and shipment of F and HF. Pb, exclusively used in the past, has now been replaced by other metals and synthetic rubbers. Ni and the Ni–Cu alloys such as Monel have been found to be most satisfactory, and mild steel has been extensively used in F cell construction, and also for the storage and shipment of anhydrous HF. 10 ref.—W. A. M. P.

The Theory of Scaling Processes in Metallic Alloys. Karl Hauffe (*Metalloberfläche*, 1951, [A], 5, (1), 1–7).—A detailed review of the theories developed by Frenkel, Wagner, Wagner and Schottky, and Hauffe. 24 ref.—E. N.

An Electrochemical Mechanism of Corrosion Inhibition by Chromates, Nitrites, and Other Oxidants. Marcel Pourbaix and Pierre Van Rysselberghe (*Corrosion*, 1950, **6**, (9), 313–315).—Corrosion inhibitors are classified, and a brief analysis of the mechanism of each type is given. The role of the inhibitor is defined in a discussion of the corrosion of Fe by a soln. free of O with inhibitor absent, and a further discussion considers the corrosion reaction when chromate or nitrite has been added. Min. inhibitor concentration for protection is shown to be that which for a given O concentration, brings the potential of the metal to a value corresponding to its passivation value. The electrochem. mechanism described shows that with insufficient inhibitor concentration, both the inhibitor and O are consumed in reduction reactions. 6 ref.—W. A. M. P.

†The Influence of Stress on Corrosion.—I.–II. Julius J. Harwood (*Corrosion*, 1950, **6**, (8), 249–259; (9), 290–307).—A comprehensive review. The effects of stress on the internal structure and energy characteristics of metals are discussed in relation to their influence on corrosion reactions. 81 ref. —W. A. M. P.

***Application of Corrosion-Resisting Materials to Railroad Electrical Construction.** H. F. Brown (*Corrosion*, 1950, **6**, (8), 268–273).—A summary of the work carried out by a committee of the Association of American Railroads in investigating the relative corrosion of ferrous and non-ferrous alloys in smoky atmospheres. All the materials tested were exposed for $>$5 years, and were examined periodically throughout the exposure period. Weight-loss and penetration data are given, and, in general, the high-Cu alloys were found to be superior to other non-ferrous metals tested. —W. A. M. P.

Some Fundamentals of Corrosion and Protection. M. A. Comley (*Electroplating*, 1951, **4**, (7), 229; and also *Indust. Finishing* (*Lond.*), 1951, **3**, (37), 976–978, 980).—Report of an address given to the Metal Finishing Association, Birmingham. —H. A. H.

The Principles of Temperature Humidity Testing Equipment. Leo Walter (*Brit. Indust. Finishing*, 1950, **2**, (25), 877–878, 880, 882, 884).—Details are given of a humidity test cabinet with automatic temp. and humidity control.—H. A. H.

Corrosion Guide. (Rabald). See col. 253.

7 — PROTECTION

(Other than by Electrodeposition.)

New Hard Coating Gains Wear Applications for Aluminium. —— (*Materials and Methods*, 1950, **32**, (2), 62–64).—MHC finish is a new electrochem. applied " file-hard " coating for Al and Al alloys contg. $<5\%$ Cu developed by the Glenn L. Martin Co. It is thought to be basically a film of Al_2O_3 with thickness from 0·0001 to 0·006 in., the usual value being 0·002 in. The coating has very high wear-resistance; properties and possible appn. are summarized.—J. W. C.

Automatic [Aluminium] Metal Spraying of Steel Tubes. —— (*Indust. Finishing* (*Lond.*), 1951, **3**, (34), 712–714, 716).—A short description of an automatic plant for the Al spraying (0·004 in.) of steel scaffolding tubes.—H. A. H.

The Chromium-Diffusion Process. —— (*Product Finishing* (*Lond.*), 1950, **3**, (10), 32–37).—A brief, general description of the Cr-diffusion process for the treatment of Fe and steel surfaces. The articles to be processed are packed in boxes,

together with a special powder contg. a high percentage of finely divided Cr and heated to 880°–1040° C. It is claimed that by careful control of the operating conditions the depth of the diffused layer can be controlled from 0·0005 to 0·025 in. The advantages and appn. of the process are discussed.
—H. A. H.

Sprayed [Molybdenum] Metal Coatings on Polished Surfaces. R. Corbett (*Indust. Finishing* (*Lond.*), 1951, **3**, (37), 955–956, 958–961).—Mo can be applied by the wire pistol to polished surfaces and good adhesion of the Mo obtained. Sprayed steel and other coatings have good adhesion to flash Mo deposits, so that grit blasting or machining can thereby be eliminated from the pretreatment cycle for normal metal spraying. Some details of spraying technique are discussed.
—H. A. H.

Outlook for Continuous Galvanizing Lines in U.S.A. E. A. Matteson (*Products Finishing*, 1951, **15**, (10), 64).—It is predicted that several new strip galvanizing lines will be required in the U.S.A. to meet the demand by 1953, though this depends on the availability of steel and Zn.—G. T. C.

Possibilities of Saving Zinc in Zinc Coating. A. Glazunov and L. Jeníček (*Chem. Listy*, 1944, **38**, 237–244; *C. Abs.*, 1950, **44**, 5726).—Galvanic Zn coatings are more economical than hot-dip ones, the galvanic coating being more regular and thinner than that obtained by immersion.

A Review of Factors Influencing the Life of Galvanizing Pots. D. N. Fagg (*Sheet Metal Ind.*, 1951, **28**, (290), 574–575).—F. reviews the factors influencing pot life and concludes that the life depends on: (1) Zn temp., (2) type of steel used for the pot, (3) fabricating defects in the pot, and (4) compn. of Zn. He discusses these factors and shows that the life of a pot is governed by the observance of two main conditions: (*a*) bath temp. must be below 480° C. and (*b*) the steel used must contain $\not>$0·07% Si.—R. J.

***The Flow of Liquid Metals on Solid Metal Surfaces and Its Relation to Soldering, Brazing, and Hot-Dip Coating.** (Bailey and Watkins). See col. 170.

Electroplating Aids in Clad-Steel Production. (Sangster). See col. 205.

Progress in Metal-Spraying Equipment. H. J. Plaster (*Electroplating*, 1951, **4**, (7), 215–217).—A short survey of the development of semi- and fully-automatic metal-spraying equipment.—H. A. H.

Metal Spraying of the Clunie Pipeline. W. R. Jennings (*Brit. Indust. Finishing*, 1950, **2**, (21), 583–586).—An account of the use of 6½ tons of Zn applied by the wire-spraying process as a 0·004-in.-thick layer for the protection of the three 12 ft. 3 in. dia. pipelines, the 21 ft. 6 in. dia. pipeline, and the 21 ft. 6 in. dia. steel tunnel lining for the Tummel-Garry scheme of the North of Scotland Hydro-Electric Board.
—H. A. H.

The Use of Protective Gases in Metal Spraying. — Püschel. Hans Biel (*Metalloberfläche*, 1951, [A], **5**, (2), 28–29).—Cf. Püschel, *ibid.*, 1950, [A], **4**, 141; *Met. Abs.*, 1950–51, **18**, 712. A discussion between P. and B. as to the relative costs, certainty of action (if any), and dangers of using protective atmospheres of H, coal gas, NH_3, C_2H_2, and C_3H_8 during metal spraying.—E. N.

Cathodic Protection of an Active Ship in Sea-Water. K. N. Barnard and G. L. Christie (*Corrosion*, 1950, **6**, (7), 232–234).—The cathodic protection of H.M.C.S. "New Liskeard" with Mg alloy anodes is described. Results to date indicate that cathodic protection of an active ship of this size is economically practical.—W. A. M. P.

Cathodic Protection of Fourteen Offshore Drilling Platforms. E. P. Doremus and G. L. Doremus (*Corrosion*, 1950, **6**, (7), 216–224).—An account is given of the cathodic protection with Mg anodes of over half a million ft.^{2} of submerged bare steel surfaces on drilling platforms and tenders in the Gulf of Mexico. Current and potential data are presented, and experiences and results obtained during installations on several different types of structures are discussed.
—W. A. M. P.

The Lacquering of Metal Articles. A. Hofmann (*Metalloberfläche*, 1951, [B], **3**, (4), 51–55).—A review of the factors involved in the choice of spirit or synthetic lacquers and lacquering procedures for the prodn. of coatings having good: (1) corrosion-resistance, (2) adherence and mech. strength, and (3) æsthetic properties.—E. N.

Equipment for Batch and Continuous Lacquering. H. Krause (*Metalloberfläche*, 1951, [B], **3**, (4), 58–60).—A brief, illustrated review, with special ref. to contemporary German equipment for vat and barrel lacquering and for the continuous lacquering of cables, &c.—E. N.

Modern Anti-Corrosive Paints. H. Hesse (*Metalloberfläche*, 1951, [B], **3**, (4), 56).—A review. Special ref. is paid to the fact that each type of pigment base has an individual, critical oil content; unless this is strictly adhered to the best results cannot be obtained.—E. N.

Rust Prevention Processes and Rust Preventatives Described in the Technical and Patent Literature. Richard Springer (*Metalloberfläche*, 1951, [A], **5**, (2), 22–28; (3), 43–46).—A review, with 113 ref.—E. N.

Chromate Finishes in War and Peace. H. C. Irvin (*Metal Finishing*, 1951, **49**, (7), 109–113).—The appn. of chromate treatments to Zn plating, hot-dip galvanizing, and Zn-base die-castings is described, and their use on Al, Mg, Cu, brass, and bronze is briefly mentioned. The nature and properties of chromate films are described and information given on the formulation of chromate baths.—G. T. C.

Phosphate Coatings in Defence Production. Alfred Douty (*Metal Finishing*, 1951, **49**, (7), 105–108, 120–121).—Various appn. of phosphate coatings are described including their use for paint bonding on Zn, Cd, Al, and their alloys.—G. T. C.

Protecting Surfaces from Deterioration. —— (*Pract. Eng.*, 1951, **23**, (578), 336–337).—A general review of corrosion and methods of protection.—E. J.

***Investigations Into Tropic Proofing of Electrical Materials, 1943–46. I.—The Protection of Electronic Equipment for Use Under Humid Tropical Conditions.** L. G. Dobbie (*Australian J. Appl. Sci.*, 1950, **1**, (1), 80–96).—D. reviews Australian investigations into service problems during World War II. Moisture collection during storage was the main cause of trouble, particularly where spot welding had destroyed the protective finish and left crevices. Testing methods needed: (*a*) closely controlled humidity (such as 96 ± 2%) at all temp.; (*b*) regulated temp. rise rate (affecting condensation); and (*c*) elect. testing of samples while still in the test chamber. Two humidity cabinets, of 5 and 20 ft.^{3} capacity, are described, but not illustrated. Hermetic sealing is still the only means of complete protection. Appendices give: (i) Diffusion of Water Vapour [in Organic Materials] and (ii) Data on Water Vapour and Evaporation.
—S. M.

8 — ELECTRODEPOSITION

Chromium Is Plated Directly on Aluminium by New Process. T. C. Du Mond (*Materials and Methods*, 1950, **32**, (4), 56–57).—Al is wet blasted using quartz crystals as abrasive and then transferred immediately into the plating bath. A coating of the suspension is retained on the surface and prevents oxidation while the transfer is made and a direct bond is thus formed between the Al and the Cr.
—J. W. C.

***Chromium-Plated Light-Metal Cylinders [for Internal-Combustion Engines].** P. Riekert and W. Hampp (*Metalloberfläche*, 1951, [A], **5**, (3), 33–37).—After briefly reviewing the development of successful Cr-plating techniques for light metals—Y alloy, Hydronalium, Silumin, EC 124, and KS 280—R. and H. describe the results of some unpublished war-time researches, by Rossenbeck and Kuhm, in relation to the use of hard Cr-plated cylinders and pistons—of the casting alloy EC 124—in internal-combustion engines. The Cr plating of cylinders was found to be more satisfactory than that of pistons. Such cylinders showed less wear and had lower working temp. than those made from grey cast Fe. The experiments also emphasized that for the best results: (1) great care must be exercised in the pickling operations before plating, and (2) the hard Cr surfaces must be honed, or when thick, polished before use. 17 ref.—E. N.

†**The Chromium Plating of Light-Metal [Aluminium Alloy] Cylinders [for Internal-Combustion Engines].** Erich Meyer-Rässler (*Metalloberfläche*, 1951, [B], **3**, (3), 33–42).—After comparing the phys. properties of grey cast Fe, Al alloys, and Cr, in so far as they are concerned with the characteristics of internal-combustion engines, M.-R. gives a critical analysis of the prodn. technique, and properties of Cr-plated Al alloy cylinders, viz. (1) hard Cr plating-bath compn. and operation, and the properties of the deposits produced—thickness, hardness, and the characteristics of the hair-cracks occurring in such films; (2) porous Cr—its prodn. by the following methods: (*a*) chem., (*b*) electrolytic—reversal of polarity within the Cr bath itself, or by cathodic treatment in acids or alkalies, (*c*) mech.—by knurling, and (*d*) rough grinding of the cylinder surfaces, so that the Cr deposit has a similar contour to that of the base metal; (3) suitable Al alloys and methods of prodn.—most types of casting alloys and methods of casting are satisfactory provided that porosity and segregation are kept extremely low; forging is also a suitable method of manufacture; (4) the action of heat on the Cr deposits—the variation of the coeff. of friction with temp.; (5) German and American operating experience with engines using Al cylinders which have been Cr plated in comparison with those which have Fe sleeve inserts; and (6) future lines of development as regards the manufacture and use of Cr-plated cylinders. 9 ref.—E. N.

Self-Regulating Chromium Bath Plates 20 to 80% Faster. A. H. Allen (*Steel*, 1950, **127**, (16), 98–99).—Describes the use of a new proprietary Cr-plating bath for which greater efficiencies are claimed.—E. J.

Hard Chrome Plating: The Industrial Workhorse. J. A. Williams (*Metal Finishing*, 1951, **49**, (7), 100–102, 120).—The properties of Cr plate are reviewed in general terms, and a list is given of the appn. of hard Cr plating.—G. T. C.

The Problem of Painting Bright Chromium Plate. (McMaster). See col. 232.

***Practical Observations on a Cyanide Copper [Electroplating] Bath.** W. Savelsberg (*Metalloberfläche*, 1951, [B], **3**, (1), 4–6).—Current/voltage curves are presented for the electrolysis of aq. soln. of: (1) $CuCN + KCN$, (2) $Cu(CN)_2 + KCN$, and (3) $CuCN + Cu(CN)_2 + KCN$. They show that although the decompn. voltage of soln. contg. only $Cu^{\cdot}$ ions is $\sim$2·81 V., when $Cu^{\cdot\cdot}$ ions are present there is a kink at $\sim$2·02 V. in the current/voltage curve, which corresponds to the transformation $Cu^{\cdot\cdot} \rightarrow Cu^{\cdot}$. This is considered to confirm the hypothesis that the porous, blistered deposits which are often obtained with $Cu^{\cdot}$ soln. are caused by the presence of $Cu^{\cdot\cdot}$ ions in the soln. Such trouble can be avoided by the addn. of Na_2SO_3 (10 g./l.) to the electrolyte; this salt: (1) is anodically oxidized preferentially to the $Cu^{\cdot}$, and (2) reduces to $Cu^{\cdot}$ any $Cu^{\cdot\cdot}$ present in the soln.—E. N.

***Colorimetric Determination of Zinc and Cobalt in Gold Alloy Plating Solutions.** (Mermillod). See col. 215.

***[Electrolytic] Lead Coatings in Protection of Metals Against Corrosion.** A. Glazunov, L. Jeníček, and V. Vondráček (*Chem. Listy*, 1944, **38**, 98–103; *C. Abs.*, 1950, **44**, 5726).—Successful Pb coating was accomplished by electrolysis in a complex Pb soln. (*p*H 5–6) (cf. G. and J., *Korrosion u. Metallschutz*, 1941, **17**, 384; *Met. Abs.*, 1943, **10**, 78). The coating is poreless at the average thickness of some thousandths of a mm. and has excellent adhesion. It sticks to rough surfaces; no special pretreatment is required. Results of corrosion tests with 1% H_2SO_4, KOH, and $K_2Cr_2O_7$ are given. Special passivation with 10% H_2SO_4, H_3PO_4, K_2CrO_4, Na_3PO_4 improves the resistance.

***Bright Nickel Plating on Metallic Single Crystals in the Absence of Addition Agents.** Henry Leidheiser, Jr., and Allan T. Gwathmey (*J. Electrochem. Soc.*, 1951, **98**, (6), 225–230).—Ni was electrodeposited on electrolytically polished single-crystal spheres of Cu and Ni under a wide variety of experimental conditions from a Watts bath which was generally operated in the *p*H range 3–4. The bath was made up with reagents of different purity, with exactly the same results in each case. The deposit formed on the (100) faces and on faces making a small angle with the (100) plane remained mirror bright to a thickness of 0·01 in. The deposits which formed on the (111) faces and on those faces making a small angle with the (111) plane were matt from the very start of the deposition. X-ray analysis at grazing incidence showed that in the case of the (100) faces the deposit remained mono-cryst., whereas in that of the (111) faces it was polycryst. The latter rapidly assumed a (100) preferred orientation under the conditions of the experiment. The hardness of the polycryst. deposit was found to be greater than that of the monocryst. one, but its ductility was less. The monocryst. deposit was compact, but the other was comparatively porous. A few tests were carried out on the relative corrosion-resistance of the two deposits using 10% HCl, the action of A.C. in the Watts bath, and electrolytic etching in HNO_3–acetic acid soln. In all cases the corrosion-resistance of the monocryst. deposit was the greater. A few preliminary experiments indicated that the high-temp. oxidation properties of monocryst. Ni, prepared by electrodeposition, were essentially the same as those of a massive single cryst. prepared from the molten metal. 3 ref.—G. T. C.

Heavy Silver Plating of Bearings. Frank C. Mesle (*Metal Finishing*, 1951, **49**, (7), 103–104, 121–122).—Various methods of producing bearings are described briefly and the advantages of Ag bearings emphasized. These are deposited to a thickness of 0·02–0·06 in. and have been shown to have a much longer life than Cu–Pb bearings. C.d. as high as 100 amp./ft.^{2} may be used in depositing the Ag. A brightener is normally added to the bath. After plating, the deposit is heated to 950°–1000° F. (510°–540° C.) in order to improve bonding.—G. T. C.

Deficient and Excessive Silver Plating of Cutlery. Carl Schaarwächter (*Metalloberfläche*, 1951, [B], **3**, (2), 17–18).—Graphs are given showing the deviation from standard of the weight of Ag found on individual members of sample batches of electroplated spoons and forks. They show how necessary it is to exercise regular and close control of the plating bath in order to avoid the prodn. of coatings which are too thin or too thick, and which, therefore, either wear badly or result in excessive prodn. costs.—E. N.

Continuous Electrolytic Tin Coating. —— (*Product Finishing* (*Lond.*), 1951, **4**, (7), 52–62).—A description of the manufacture of electrolytic Sn-plated sheet or strip at the Weirton Steel Co., W. Va., U.S.A., with special ref. to the most recently installed fourth tinning line.—H. A. H.

Electro-Tinning Steel Strip: Expenditure of Electrical Energy. John H. Mort (*Iron and Steel*, 1949, **22**, (4), 112–116).—Calculations are made of the energy required for deposition of Sn and for re-flowing the coating to give a bright appearance.—N. B. V.

Bright Zinc Plating. R. O. Hull and J. B. Winters (*Metal Finishing*, 1951, **49**, (6), 58–61, 63).—A general review of the process and its main appn. is presented. Details are included on the prepn. and compn. of soln. used for both vat and barrel plating, and information is given on operating details, &c. Methods of bath maintenance are described, and full details are included on the analysis of bright Zn baths. A method of preparing cast Fe for Zn plating is given, and there are brief notes on Zn bright dips and on H embrittlement.—G. T. C.

Determination of Plate Thickness on Zinc-Base Alloy Die-Castings. Harold E. Brown (*Plating*, 1951, **38**, (6), 556–557).—The thickness of Cr plate is determined by dissolving off the Cr from significant surfaces, having stopped-off non-significant areas, with HCl and then determining the Cr content of the soln. obtained. To determine the thickness of Cu and Ni, the casting itself is dissolved in HCl of definite strength, and the shell of Cu and Ni then removed and non-significant portions of the coating broken off. After checking that the shell is quite free from Zn, it is weighed. The Cu is then dissolved, either by the CrO_3–H_2SO_4 or the poly-sulphide–cyanide method, the former being recommended for fine-grained Cu. The Ni shell remaining is weighed, thus enabling the average thickness of both Cu and the plate to be calculated. The method is only suitable for comparatively small castings.—G. T. C.

[Discussion on a Paper by S. Senderoff and A. Brenner on :] Plating Experiments with Aqueous Solutions at High Temperatures. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 256).—Cf. *Met. Abs.*, 1950–51, **18**, 379.—G. T. C.

Electroplating Stainless Steel : A Critical Review with Recommended Practices. Joseph Haas (*Metal Finishing*, 1951, **49**, (6), 50–54).—The nature of the problem is briefly described and the literature and patents on the subject reviewed. A comprehensive chart is presented in which are summarized the various methods which can be employed for the preparation of stainless steel for plating and the actual plating processes, including Cu, Ni, and Cr plating. 11 ref. —G. T. C.

Electroplating Aids in Clad-Steel Production. S. E. Sangster (*Products Finishing*, 1951, **15**, (9), 12–16).—The use of Ni plate, 0·005 in. thick, as a means of ensuring a satisfactory bond between the composite layers of clad steel is briefly described.—G. T. C.

Barrel Deburring and Plating of Steel Stampings. Ezra A. Blount (*Products Finishing*, 1951, **15**, (8), 28–30, 32).—Brief details are given of the make-up of the load in the deburring barrels and of the compn. of the soln. and conditions of operation used for Ni plating. A treatment with a rust-proofing oil emulsion follows plating.—G. T. C.

Military Applications of Electroforming. M. H. Orbaugh (*Metal Finishing*, 1951, **49**, (7), 98–99, 118–120).—The technique of electroforming is described in general terms and its appn. to the prodn. of aircraft plastic tooling, Radar waveguides, aircraft fittings, surface-roughness gauges, and computing cams is mentioned.—G. T. C.

Electroforming of Precision Military Components. S. G. Bart (*Metal Finishing*, 1951, **49**, (7), 96–97, 116–117).—The various types of electroforming are briefly described, and some information is presented on the mech. properties of electrodeposited metals. Brief mention is made of the appn. of the technique.—G. T. C.

Finishing [U.S.] Army Ordnance Matériel. G. T. Viglione (*Metal Finishing*, 1951, **49**, (7), 89–91, 114–115).—Various finishing methods are briefly reviewed, including Zn and Cd plating and high-speed Cr plating.—G. T. C.

Plating in the [U.S.] Rearmament Programme. Seymour Senderoff (*Metal Finishing*, 1951, **49**, (7), 86–88, 114).—It is pointed out that the emphasis will cease to be on decorative plating and will instead be on plating for corrosion-resistance and for indust. appn. Some of the probable appn. of electroplating are reviewed.—G. T. C.

Plating's Role in Naval Ordnance : The Naval Gun Factory at Anacostia [U.S.A.]. —— (*Metal Finishing*, 1951, **49**, (7), 92–95, 115–116).—The development of electrodeposition in the factory is briefly reviewed, and information is presented on the equipment used and on the soln. compn. and operating conditions. Cd is recovered from the Cd stripping bath, which consists of a soln. of 8–16 oz./U.S. gal. NH_4NO_3, by first precipitating with NaOH, after proper precautions have been taken to get rid of the NH_3 fumes evolved. The precipitate is allowed to settle and the clear soln. siphoned off. Then follow 5 washings with water, and the precipitate is then dried by means of steam coils to a paste which is stored in earthenware jars and used to replenish the plating soln. as required, being first dissolved in NaCN. Experience shows that after such addn. have been made, the plate is less bright for an hr. or so and that extra brighteners are therefore required.—G. T. C.

Electrodeposition in the [Printing] Foundry.—I.–XII. J. S. Riley (*Electrotypers' and Stereotypers' J.*, 1949, **14**, (6), 122, 124 ; 1950, **15**, (2), 34, 36 ; (3), 58, 60, 62 ; (4), 82, 84 ; (5), 114, 116, 118 ; (6), 138, 140, 142 ; (7), 162, 164 ; (8), 186, 188 ; (10), 234, 236 ; 1951, **16**, (1), 10, 12 ; (2), 34, 36 ; (3), 59–60).—Elementary.—N. B. V.

Production Plating of Plastics. W. A. Raymond (*Metal Finishing*, 1951, **49**, (6), 55–57).—A brief illustrated description is presented of the operations and methods used by a firm engaged in this type of work. No details are included as to the compn. of the various soln. used.—G. T. C.

Review of Finishing Problems in Powdered Metal Products. —— (*Product Finishing* (*Lond.*), 1951, **4**, (5), 55–64).—Problems associated with the electroplating of powdered metal components are reviewed and discussed. Other finishing methods receive much less attention. 6 ref.—H. A. H.

Making Small Metal Tubes by Electrodeposition on Nylon Fibres. Rolf J. E. Gezelius (*Rev. Sci. Instruments*, 1950, **21**, (10), 886).—A method is described by which tubes several inches in length with inside dia. as small as 0·05 mm. can be made. Metals such as Cu, Ag, and Ni have been used, and it is possible to make hard and corrosion-resistant tubes of an Ni–Co alloy by simultaneous electrodeposition of these two metals.—E. J.

Modern Developments in Metal Finishing. I.—Recent Developments in Electrodeposition Processes. II.—Modern Developments in Electroplating Plants. E. A. Ollard (*Brit. Indust. Finishing*, 1950, **2**, (21), 563–564, 566–568, 570, 572 ; (22), 633–636, 638–639).—A review.—H. A. H.

Advances in Production Methods in Metal Finishing. Adolph Bregman (*Metal Progress*, 1951, **59**, (1), 77–80).—A review, with particular emphasis on plating.—F. A. F.

The Electrolytic Stripping of Electrodeposits. A. Pollack (*Metalloberfläche*, 1951, [B], **3**, (1), 8–9).—P. gives brief details of electrolytes and bath-operating conditions suitable for the removal of electrodeposits of Ni, Cr, Cu, Sn, Ag, and Au. 5 ref.—E. N.

***Cleaning and Preparation of Metals for Electroplating. III.—Degreasing Evaluation Tests : The Atomizer Test.** (Linford and Saubestre). See col. 233.

Plating Hints. Thomas M. Rodgers (*Metal Ind.*, 1951, **79**, (1), 6–7).—R. gives advice on degreasing, soak and electrolytic cleaning, Cr stripping, and checking of the condition and strength of cleaners.—J. H. W.

Solution-Heating Methods Improve Plating-Plant Operating Efficiency. Arthur Q. Smith (*Products Finishing*, 1951, **15**, (10), 58–62).—A description is given of the procedures used for the cleaning and plating of miscellaneous small articles, but only brief mention is made of tank heating, which is mainly done by gas immersion heaters.—G. T. C.

Rack Making. —— (*Plating*, 1951, **38**, (7), 709–711).—A series of photographs with descriptive captions illustrating the prodn. of plating racks.—G. T. C.

Specialist Attention to Plating Jigs. J. J. Preston (*Brit. Indust. Finishing*, 1950, **2**, (21), 601–603).—The importance is emphasized of correct plating jig design and construction. —H. A. H.

[Plating] Material Economy. Frank Wild (*Electroplating*, 1951, **4**, (7), 218–220).—Recommendations are made for obtaining the max. efficiency from cleaners, swills, acid etches and pickles, anodes, and electroplating soln. These refer particularly to Ni-plating procedures.—H. A. H.

***Disposal of Plating Room Wastes. III.—Cyanide Wastes : Treatment with Hypochlorites and Removal of Cyanates.** Barnett F. Dodge and Walter Zabban (*Plating*, 1951, **38**, (6), 561–566, 571–586).—Cf. *ibid.*, 1949, **36**, 723 ; *Met. Abs.*, 1949–50, **17**, 191. Work carried out under American Electroplaters' Soc. Research Project No. 10 is described. It covers the first part of an investigation for the disposal of electroplating wastes contg. cyanides and is limited to the method which is based on the oxidation of the cyanide ion by means of the hypochlorite ion, a method in which the cyanide is first oxidized to cyanate and the latter then converted to other C and N compounds. Full details of the experimental method and extensive numerical results are included. It is concluded that cyanides can be oxidized completely (to $\nless$0·2 p.p.m. CN) to cyanates by hypochlorites in <5 min. at room temp. Only cyanide soln. contg. Ni^{2+} require a longer time for the reaction to take place, viz. $\frac{1}{2}$ hr. In the latter case an excess of 20% of the reagent is required. Cyanates are completely destroyed (to $\nless 1$ p.p.m. CNO) by oxidation to CO_2 and N_2 at room temp. in ~ 1 hr. The oxidation of CN to CNO should preferably take place in the *p*H range 10–12·5, but the oxidation of CNO should be carried out in the range 7·5–9, though there is an indication that very rapid oxidation takes place at *p*H 6·0–6·5. It is stated that the formation of the explosive and toxic NCl_3 is unlikely if cyanates are oxidized at above *p*H 6·0. Detailed information is included as regards quantities of reagents to use. The use of $Ca(OCl)_2$ leads to more sludge formation than does that of Cl_2 or NaOCl, and it is recommended that one of the latter be used if the CN content be above 5000 p.p.m. A rapid test for the detection of traces of CN is described. The elimination of cyanates by acid hydrolysis is also described. 19 ref. —G. T. C.

A Method for Disposal of Acid and Alkaline Plating-Room Wastes. J. H. Monaweck and Clyde Kelly (*Products Finishing*, 1951, **15**, (9), 38, 40).—Brief details are presented of the mixing of acid waste with alkaline waste in rubber-lined tanks of suitable capacity. Stirring equipment and stainless-steel steam coils are required.—G. T. C.

The Purification of Plating Solutions.—II.–III. G. T. Colegate (*Electroplating*, 1949, **2**, (5), 307–315; (6), 417–423, 425–426).—Cf. *ibid.*, (4), 221; *Met. Abs.*, 1949–50, **17**, 190. [II.—] Deals solely with the purification of Ni electroplating soln. Consideration is given to the removal of metallic impurities by low- and high-c.d. electrolysis and by chem. pptn.; the removal of organic impurities by adsorption and chem. or electrochem. oxidation; and a method for dealing with CrO_3 contamination. [III.—] The purification of Cu, brass, speculum, Zn, Cd, Sn, Ag, Cr, Au, and Rh plating soln. is discussed.—H. A. H.

Control of Electroplating Solutions by Analysis and Observation. XII.—Electro-Analysis of Plating Solutions Using Home-Made Apparatus. XIII.—The Choice and Control of Cleaning Solutions. K. E. Langford (*Electroplating*, 1950, **3**, (12), 443–445, 454; 1951, **4**, (2), 41–44).—Cf. *ibid.*, (5), 166; *Met. Abs.*, 1950–51, **18**, 48. A home-made apparatus for the electrolytic analysis of electroplating soln. is described. It is claimed to be satisfactory for the following metals: Cu (in acid or cyanide soln.), Zn (cyanide soln.), Cd, Ni, and Ag.—H. A. H.

Sampling and Flow Measurement of Waste Water. —— (*Plating*, 1951, **38**, (6), 558–559).—Five systems for automatically sampling waste water are described. Two of these operate intermittently and the other three continuously. Diagrams of all the systems are included.—G. T. C.

Further Notes from Observations of U.S. Plating Practice. J. J. Dale (*Plating Notes*, 1951, **3**, (2), 53–55).—Brief notes on: (*a*) control and testing, and (*b*) agitation of electroplating baths.—T. A. H.

The Future of Electroplating. C. J. Cocks (*Plating Notes*, 1951, **3**, (2), 46–50; discussion, 50).—Some of the fields in which noteworthy advances may be expected are the more general use of addn. agents, automatic controls, electronic equipment, and alloy plating.—T. A. H.

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9 — ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

***The Electrochemical Behaviour of the Antimony Electrode.** S. E. S. El Wakkad (*J. Chem. Soc.*, **1950**, (Oct.), 2894–2896).—Stick, plated, and powdered Sb electrodes were examined in KCl–HCl, acetic acid–Na acetate, disodium hydrogen citrate–NaOH, boric acid–borax, and Na_2CO_3–HCl buffer mixtures covering the *p*H range 1·2–13·9. The elect. measurements were performed at 20° ± 0·05° C., a saturated HgCl electrode being used as the ref. half-cell. Powdered, massive, or plated Sb electrodes are shown to be reversible, since it is suggested that the fluctuations and drift observed in E_0 values by other authors can be explained by the gradual formation of tetroxide on the electrode and its very slow rate of attaining equilibrium, which depends upon the O supply and the surface of metal exposed. Equilibrium, when reached, corresponds to the reversible value of the system Sb_2O_3–Sb_2O_4. Provided that measurements recorded represent true equilibrium values, the Sb electrode is thus a true metal–metal oxide electrode.—H. A. H.

***The Mechanism of Exchange Between Chromium Ions in Solution and Metallic Surfaces.** M. T. Simnad (*Rev. Mét.*, 1951, **48**, (7), 522–524).—If a noble metal is placed in a radioactive soln. of ions of a more basic metal, it acquires a radioactivity corresponding to a deposit of the base metal of a thickness of many atomic layers. S. undertook preliminary experiments, as part of a general programme with radioactive tracers, on radioactive Cr in soln. in the state Cr^{+++} and CrO_4^{--}, with 18 : 8 stainless steel, Fe, Cr, and Pt. In alkaline soln. of chromate in the absence of O, all the specimens acquired the same activity by exchange, but the distribution of radioactive Cr on the specimens, as evidenced by autoradiography, differed greatly from one specimen to another. In the presence of O, the deposit of ^{51}Cr is greater for Fe and less for Pt. In these cases the distribution of the ^{51}Cr is a function of the metal and the treatment. In acid chromate soln. in the presence of O more marked activities occur. The exchange on the stainless steels is weaker after treatment in 5% HNO_3 + 1% K_2CrO_4 and stronger in 30% HNO_3. There is marked segregation of the exchanged Cr. The activity of the Fe could not be determined, as the metal was very badly corroded and rusted. The Cr was also very strongly attacked. In soln. of chromic Cr, the ^{51}Cr was deposited on the cathodic zones of the metal surface. These results corroborate those of similar experiments with radioactive Co and also those of Evans and Hoar (*J. Chem. Soc.*, 1932, 2476) with chromate soln. Other experiments are in hand with Cr of much greater activity, with a view to making a quant. study of the subject. 8 ref.—J. H. W.

***The Equilibrium Between Copper and Cupric and Cuprous Ions and the Behaviour of the Copper Electrode in Dilute Copper Sulphate Solutions.** S. E. S. El Wakkad (*J. Chem. Soc.*, **1950**, (Dec.), 3563–3566).—The value of K for the reaction $Cu + Cu^{++} = 2Cu^{+}$ was determined for conditions similar to those at a Cu surface in $CuSO_4$ soln., and the potentials of Cu electrodes in Cu^{++} salt soln. calculated using this value. The break in the potential/activity curve of the Cu electrode in dil. $CuSO_4$ soln. previously noted by Tourky and El W. (*ibid.*, 1948, 749; *Met. Abs.*, 1948–49, **16**, 570) is explained in the light of these results. The calculations are extended to Cu^{+} salt soln., and it is shown that in Cu^{++} salt soln. the potential of the Cu electrode tends towards that of the standard potential/activity curve of the Cu/Cu^{+} system at $a_{Cu^{++}} < \sim 0.5 \times 10^{-4}$. In Cu^{+} salt soln., however, the potential tends towards that of the standard curve of the Cu/Cu^{++} system at $a_{Cu^{+}} > \sim 0.5 \times 10^{-4}$. Using two methods of calculation, a mean value of 0·48 ± 0·01 V. was obtained for the standard Cu/Cu^{+} potential. $CuSO_4$ soln. shaken at 21° ± 0·5° C. with Cu for 5 days in an N atmosphere suffered a *p*H change of ~ 1.5 units in some cases. The bearing of this on the probable pptn. of Cu_2O or hydroxide on the Cu electrode in very dil. $CuSO_4$ soln. is discussed.—H. A. H.

***Relationships Between Germanium and Cadmium in the Electrolysis of Zinc Sulphate Solutions.** S. T. Ross and J. L. Bray (*J. Metals*, 1951, **3**, (6), 465–467).—A report of a study of the current efficiency in a model Zn cell, as a function of the concentration of Cd and Ge present. The electrolyte contained $ZnSO_4$ 100, Cd 0–2·5, and Ge 0–0·01 g./l., while the following variables were held const.: cell temp. (24° C.), voltage (10 V.), c.d. (30 amp./ft.^{2}), anode (Pb) spacing, relative electrode area, degree of agitation, cathode (2S Al alloy), prepn. technique, time (30 min.), acid concentration and soln. vol. (250 ml.). The results, which are shown in tables and graphs, substantiate the claim that definite interaction exists in the Zn cell between Cd and Ge concentrations. Ge alone, within normal operating limits, does not cause large current efficiency losses. Rather, the adverse effects of Ge are notably enhanced by the presence of Cd, and the presence of Ge in leach soln. of the commercial $ZnSO_4$ type precludes the complete removal of Cd from them. The findings are discussed in relation to the efficient electrowinning of Zn from Ge-tainted ores. 9 ref.—E. N.

[Discussion on a Paper by S. Schuldiner and R. E. White on :] Studies of Time–Potential Changes on an Electrode Surface During Current Interruption. I.—Zinc–Steel Couple in Synthetic Sea-Water. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 259).—Cf. *Met. Abs.*, 1950–51, **18**, 472.—G. T. C.

Possibilities of Extracting the Hydrogen Ion from Electrolytic Solutions. L. H. Collet (*J. Phys. Radium*, 1951, **12**, (6), 697).—A letter. Darmois has recently conceived the H-ion in electrolytic soln. as composed of H-ions, H^+, in a semi-free state, comparable with that of electrons in metals (*ibid.*, 1950, **11**, 577). C. shows that the energy of hydration of the H^+-ion corresponds, to a first approximation, with the energy necessary to extract the proton from soln., and suggests that protons may possibly be extracted from soln. by: (1) bombardment of the soln. by particles possessing considerable energy, e.g. by elect. discharge from a metal cathode, or by fine particles, or (2) by absorption of radiation of suitable wave-length.—J. S. G. T.

Hydrogen Overvoltage, Electric Double Layer, and Related Questions. (Mlle) Geneviève Sutra (*J. Phys. Radium*, 1951, **12**, (6), 673–681).—Recent theories of the nature and origin of H overvoltage, more especially that due to Darmois (*ibid.*, 1950, **11**, 577) are briefly reviewed. S. has recently shown that the H^+-ion is simply a proton; previous theories had conceived it as composed of a free proton encased in mol. of H_2O, and having the formula $(H_3O)^+$. It has been customary to distinguish the min. value of the overvoltage from that associated with variable current. S. considers the former voltage not to be an overvoltage, but the normal equilibrium voltage. The elect. double layer present at a metal|electrolyte interface is conceived as composed of widely spaced ions separated by H_2O mol. Overvoltage at the metal|electrolyte interface is essential to effect penetration of the elect. double layer at the interface during passage of current. The theory of overvoltage is developed on this basis, with satisfactory agreement between theoretical and experimental results. 36 ref.—J. S. G. T.

***Some Electrical Relations in Galvanic Couples.** H. D. Holler (*J. Research Nat. Bur. Stand.*, 1950, **45**, (5), 373–380).—The elect. theory of galvanic couples (short-circuited galvanic cells) is outlined and equations are derived for a couple with and without polarization by external current. Methods by which some of the elect. quantities may be measured or computed are described; in the usual galvanic couple only the p.d. E_G–E_S between the couple and a ref. electrode at const. potential E_S may be directly measured. From the relation of E_G to applied current I and applied p.d. E, current and potential criteria for cathodic protection for a couple having resistance R_G are derived. R_G cannot be measured independently, but when it is negligibly small the equation for cathodic protection $E = (E_A - E_S)$ is equivalent to the identity $(E_A - E_S) = (E_G - E_S)$. The theory applies to most electrochem. processes, including deposition and corrosion, and is also applied to multiple galvanic couples at different potentials.—J. W. C.

[Discussion on a Paper by H. D. Holler on :] Studies on Galvanic Couples. I.—Measurement of Electromotive Force and Internal Resistance of Cells During Current Flow. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 252–253).—Cf. *Met. Abs.*, 1950–51, **18**, 282.—G. T. C.

[Discussion on a Paper by G. Wranglén on :] Electrodeposition of Metal Powders. —— (*J. Electrochem. Soc.*, 1951, **98**, (6), 255–256).—Cf. *Met. Abs.*, 1950–51, **18**, 381.
—G. T. C.

***The Effect of Suspended Solid Substances in the Electrolyte on the Electrocrystallization of Metals.** Gösta Wranglén (*Svensk Kem. Tid.*, 1950, **62**, 43–49; *C. Abs.*, 1950, **44**, 5727).—[In English]. The effect of solid substances suspended in the electrolyte on the electrodeposit is discussed with ref. to 8 photomicrographs. Non-conducting solids, such as hydroxide flocks (I), may settle on the cathode and prevent crystal growth at such points. In the uncovered parts, isolated crystal groups are formed. I precipitated in the cathode film provides support for H bubbles, the normal cause of pitting. I may even cause powdery deposits. Conducting solids, such as flakes of anode slime, may settle on the cathode to form starting points for radial crystn. with the formation of outgrowths. With numerous inclusions of anode slime, the crystals diverge in all directions to form a mosaic structure. Crystal outgrowths are called "somatoids". They form where the deposit has advanced ahead of the surroundings either because of a protruding point on the cathode or because of the deposit of a conducting particle on the cathode. The isolated crystal groups formed when non-conducting solids deposit on the cathode are also called somatoids. Periodic banding is explained with ref. to a Ni deposit formed at low temp. from almost neutral soln. Under such conditions, a hydroxide sol is alternately built up to supersaturation and precipitated on the cathode. During the build-up period, layers of purer metal are deposited; this gives rise to the periodic banding.

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10 — REFINING

***Metallurgy of Cobalt Production [in Japan] from Cupriferous Pyrite.** Sanai Nakabe (*J. Metals*, 1951, **3**, (6), 445–451).—A detailed description is given of the Japanese war-time prodn. process for the recovery of Co from extremely low-grade (0·1% Co) pyrite concentrates obtained from Cu ores. The steps in the process were: (1) roasting, (2) leaching, (3) pptn. as $Co(OH)_3$ contg. Co 30–35%, (4) reducing fusion to eliminate Zn and Mn impurities, thereby producing crude Co contg. Co 90, Cu 2–5, Fe 2–5, and Ni 0·5–1·5%, and (5) electrolytic refining. The crude Co was cast into anode plates 380 × 590 mm., and the pure, metallic Co was deposited on Cu cathodes, 400 × 600 mm., 10 anodes being placed at a spacing of 160 mm. The wooden electrolytic tank was lined with sheet Pb, a separate chamber for each cathode being made of a wood framework with sheets of cotton cloth on both sides serving as a diaphragm, in order to prevent Fe from depositing on the cathode. The purified electrolyte, contg. Co 30 g./l. had SO_4'' and NH_4' in concentrations regulated to correspond to a nearly equimolecular mixture of $CoSO_4$ and $CoSO_4.(NH_4)_2SO_4$, with *p*H 5·5. It was fed continually—at a rate of 200 c.c./amp.hr.—to the cathode chamber, from which it passed through the diaphragm into the anode chamber, where it picked up Fe, after which it left for purification and recirculation. Although the cathode deposit was not smooth, a week of continuous electrolysis gave no serious difficulty at a c.d. of 1 amp./dm.^{2}, and a bath voltage of 1·5 V. at 50° C. Current efficiency was ~90%. Fe, Mn, and Zn, if any, went into soln., while Cu in the anode remained undissolved as anode slime. Addn. of H_3BO_3 to the electrolyte improved the deposit. The final product had a purity of 99+%, and ~7 tons of metallic Co were produced from ~37,000 tons of pyrite cinder.—E. N.

75 Years of Electrolytic Copper Refining at the Norddeutsche Affinerie, Hamburg. Karl Prior (*Z. Erzberg. u. Metallhüttenwesen*, 1951, **4**, (7), 245–249).—A brief review of the history and development of the refinery, which has, at the present time, a capacity of 90,000 tons/year. Some details are given of the main characteristics of the electrolytic cells employed.—E. N.

[Discussion on a Paper by R. R. Rogers and G. E. Viens on :] Refining Magnesium by Vaporization at Extremely Low Pressure. —— (*J. Electrochem. Soc.*, 1951, 98, (6), 257–258). —Cf. *Met. Abs.*, 1950–51, **18**, 435.—G. T. C.

[Discussion on a Paper by R. K. McKechnie and A. U. Seybolt on :] Preparation of Ductile Vanadium by Calcium Reduction. —— (*J. Electrochem. Soc.*, 1951, 98, (6), 253–254).—Cf. *Met. Abs.*, 1950–51, **18**, 284.—G. T. C.

Applications of Amalgam Metallurgy [to Refining]. H. Hohn (*Research*, 1950, **3**, (9), 407–417).—In continuation of a previous discussion (*ibid.*, (1), 16; *Met. Abs.*, 1950–51, **18**, 542) H. deals with selected techniques. In amalgam metallurgy free-flowing amalgams are almost exclusively concerned, which to outward appearance are similar to pure Hg. Examples given include the sepn. of Al from alumino-ferro-silicon, which is operated at high pressure (15 atm.) and is based on filtration; extraction of metallic Na from the Na amalgam derived from electrolysis of molten NaCl; sepn. of Zn from contaminated Zn scrap by soln. in Hg in the presence of dil. H_2SO_4 and then electrolysing; extraction of Zn of very high purity from Zn-contg. lyes obtained in the chlorinating calcination process, by electrolysis on to Hg cathode; extraction of electrolytic Pb from low-grade Zn ores by reaction between $PbCl_2$ soln. and Zn amalgam; extraction of Cd, In, and Tl from Zn ores or electrolysis soln.; prodn. of pure Mn alloys in powder form, in which a Na amalgam is used for a phase exchange with an aq. multimetal saline soln. contg. all the components of the desired Mn alloy in the correct proportions.—F. A. F.

11 — ANALYSIS

***Analysis of Aluminium and Aluminium Alloys Using Pin Samples : Direct-Reading Spectrochemical Methods.** R. W. Callon and L. P. Charette (*Analyt. Chem.*, 1951, **23**, (7), 960–966).—The methods for analysing Al and Al alloys using a direct-reading spectrometer (ARL research model quantometer), pin electrodes, and a high-precision source unit are described and discussed.—F. M. L.

***The Determination of Aluminium in Copper-Base Alloys.** G. W. C. Milner and J. Townend (*Analyst*, 1951, **76**, (904), 424–429).—The alloy is dissolved in HNO_3, and any H_2SnO_3 formed is filtered off. The *p*H of the soln. is adjusted to ~4, and Cu and Fe then reduced with hydroxylamine hydrochloride. Al is then precipitated by adding NH_4 benzoate. The precipitate is filtered off, dissolved in a hot soln. of NH_4 tartrate, and the Al precipitated as oxinate and determined volumetrically in known manner.—F. M. L.

***Cerimetric Determination of Arsenic and Antimony.** Rudolf Přibil (*Chem. Listy*, 1943, **37**, 205–207; 227–231; *C. Abs.*, 1950, **44**, 5759).—As^{3+} and Sb^{3+} were titrated potentiometrically with $Ce(SO_4)_2$ by using Pt and HgCl electrodes. A CO_2 stream was bubbled through the soln. Sb^{3+} can be titrated in a slightly acidic soln. in the presence of a little As^{3+}, and the latter can then be titrated after addn. of ICl. If the As content exceeds that of Sb (up to 30 times), the titration is carried out in the presence of ICl, 20% vol. H_2SO_4, and 40% vol. HCl. Procedure : Dissolve 0·5–1-g. sample in 20 ml. hot conc. H_2SO_4 and boil. Add 60 ml. H_2O and 15–20 ml. conc. HCl, and titrate the Sb^{3+} at room temp. Pb, Sn, Cu, Cd, Zn, and other constituents of alloys do not interfere. After the oxidation of Sb, dilute the soln. to 250 ml.; add to a 50-ml. aliquot 10 ml. 0·005*M*-ICl and 20 ml. conc. HCl, and titrate the As^{3+} with $Ce(SO_4)_2$.

***Reductometric Titration of Antimony, Copper, and Tin with Chromous Ion.** James J. Lingane and Clemens Auerbach (*Analyt. Chem.*, 1951, **23**, (7), 986–991).—Sb^{V}, Cu^{II}, and Sn^{IV} can be titrated potentiometrically in $\nless 5M$-HCl at 85° C. with standard $CrSO_4$. Sb–Cu mixtures in essentially all proportions can be analysed and small amounts of Sb and/or Cu can be determined in the presence of large amounts of Sn. Sn can be determined in the presence of equal or smaller amounts of Sb and/or Cu, but not with larger amounts of these metals.—F. M. L.

***Beryllium Determination in the Presence of Much Iron, Aluminium, and Magnesium.** A. Leibowitz and R. S. Young (*Iron and Steel*, 1949, **22**, (12), 486).—A procedure for the detn. of Be by pptn. with 8-hydroxyquinoline from acetic acid soln. is described. The original should be consulted for details.—N. B. V.

***Determination of Boron in Metal Borides.** Herman Blumenthal (*Analyt. Chem.*, 1951, **23**, (7), 992–994).—Finely ground B or metal boride, e.g. Fe, Cr, W, Ti, or Zr, is fused with Na_2CO_3 and, if necessary, $NaNO_3$ and the melt leached with cold HCl. Excess $BaCO_3$ is added to the soln., which is then boiled to precipitate the metal constituent. After filtration the soln. is titrated in the normal manner for detn. of B with standard NaOH in the presence of mannitol. —F. M. L.

***Rapid Test for Small Concentrations of Cadmium in Zinc Solutions.** R. S. Young and C. W. Barker (*Chemist Analyst*, 1948, **37**, (4), 81–82).—A works' test for the approx. detn. of 0·001–0·004 g. Cd/l. in $ZnSO_4$ electrolyte is based on the turbidity produced by Cd with β-naphthoquinoline.—F. M. L.

***Detection of Metallic Copper by Electrolysis.** C. Franklin Miller (*Chemist Analyst*, 1950, **39**, (1), 9).—A non-destructive test for the presence of Cu on the surface of objects comprises passing a current from the object as anode through a piece of cotton-wool saturated with a soln. of KBr in H_3PO_4 to a Pt cathode. Cu is indicated by a bright purple-red stain on the cotton-wool.—F. M. L.

***The Spectrochemical Determination of Zinc, Lead, and Iron in Copper and Copper Alloys.** Frederick V. Schatz (*J. Inst. Metals*, 1951–52, **80**, (2), 77–84).—A spectrochem. method suitable for the routine works' control of Cu and Cu alloys has been developed. The procedure outlined is applicable to high-speed analysis, has a precision capable of determining Zn in brass with an error (expressed as a standard deviation) of 1·18% of the content or less, and has a sensitivity capable of detecting common impurities in Cu alloys down to 0·005% or less. The detn. of Zn, Pb, and Fe in brasses and leaded brasses is considered in detail. Factors important to the accuracy of the analysis are also considered; these include discharge characteristics, sample polarities, effects due to the nature of the Pb dispersion, and curve shifts due to varying quantity of the matrix element.—Author.

***Determination of Germanium : Colorimetric Determination of Microgramme Amounts with Oxidized Hematoxylin.** Hanna Newcombe, W. A. E. McBryde, John Bartlett, and F. E. Beamish (*Analyt. Chem.*, 1951, **23**, (7), 1023–1027).—0·4–1·2 p.p.m. Ge is determined by distilling the sample with HCl and then extracting the Ge from the distillate, made up to 10*N*-HCl, into CCl_4. The Ge is then extracted back into water, the *p*H adjusted to 3·2 with a phthallate buffer, gelatin added, and the Ge then determined colorimetrically by adding a soln. of hematoxylin oxidized by warming with H_2O_2.—F. M. L.

***Anion Exchange of Complex Ions of Hafnium and Zirconium in HCl–HF Mixtures.** E. H. Huffman and R. C. Lilly (*J. Amer. Chem. Soc.*, 1951, **73**, (6), 2902–2905).—Cf. *ibid.*, 1949, **71**, 4147; *Met. Abs.*, 1949–50, **17**, 523. The elution of complex ions of Zr and Hf from an anion-exchange resin with HCl–HF mixtures has been studied in two concentration ranges. Differences in the dependence of the distribution coeff. upon the concentrations of Cl ion, H ion, and HF have been noted, and some possible ionic equilibria are discussed.—J. R.

Colorimetric Determination of Iron : A Review of Known Methods.—III. T. S. West (*Metallurgia*, 1951, **43**, (260), 311–313, 315–316).—Cf. *ibid.*, (259), 260; *Met. Abs.*, this vol., col. 143. Consideration is given to the colorimetric detn. of Fe using mercapto-acetic acid, kojic acid, nitroso-*R*-salt, orthonitrosophenol, dimethylglyoxime, 8-hydroxyquinolines, 4-hydroxydiphenyl-3-carboxylic acid, *iso*nitrosodimethyl-dihydroresorcinol, acetylacetone, alloxantin, pyramidone, resorcylaldoxime, resorcylic acid, cupferron, pyrocatechol, pyrogallol, and quinaldinic acid. Mention is made of other organic and inorganic reagents. 61 ref.—F. M. L.

***Separation of Iron (III) from Aluminium.** Harry Teicher and Louis Gordon (*Analyt. Chem.*, 1951, **23**, (6), 930–931).—Up to 2 mg. Fe is removed from up to 80 mg. Al as the negatively charged ferric thiocyanate complex ion on the anion exchange resin Amberlite IRA-400 A, which has previously been treated with HCl.—F. M. L.

***Separation of Lanthanons by Means of Complexes with Amino-Acids.** R. C. Vickery (*J. Chem. Soc.*, 1950, (Aug.), 2058–2061).—Complex formation between the lanthanons and amino-acids enables the lighter earths to be separated by oxalate fractionation. Histidine is more satisfactory than glycine, but cystine, glutamic acid, and aspartic acid are of little value. Using histidine, fractional pptn. with oxalic acid gave good concentration of La and Sm at the head and tail of a short series; La increasing from 40 to 95% in the first fraction and being absent after the fourth. Sm increased from 5 to 50% in 6 fractions, some sepn. of Pr and Nd also took place. Little sepn. could be obtained with the heavy, less-basic lanthanons, using either histidine or glycine. Indications of the possible existence of 2 co-ordination stages are reported.—H. A. H.

Antimonial Lead as Reductant for Stannic Ions. Charles Goldberg (*Chemist Analyst*, 1950, **39**, (1), 14).—Pb–15% Sb alloy is a cheap and efficient reducing agent for Sn prior to its detn. by titration with I_2.—F. M. L.

***An Arsenate–Cerimetric Method for the Determination of Magnesium.** J. P. Mehlig and B. E. Buell (*Chemist Analyst*, 1950, **39**, (4), 76–79).—Mg is precipitated as $MgNH_4AsO_4$ and the precipitate dissolved in dil. H_2SO_4. The arsenate is reduced to arsenite by boiling with red P in the presence of I_2 and, after removal of I_2 as AgI and excess Ag' with HCl, is titrated with standard $Ce(SO_4)_2$, using ICl as catalyst and *o*-phenanthroline-Fe^{II} complex as indicator. —F. M. L.

***The Use of Complexones in Chemical Analysis. X.—Colorimetric Determination of Manganese.** R. Přibil and E. Hornychová (*Coll. Trav. Chim. Tchécoslov.*, 1950, **15**, (8/9), 456–462).—[In English]. Cf. *ibid.*, (5/6), 260; *Met. Abs.*, 1950–51, **18**, 202. Small amounts of Mn are determined spectrophotometrically or photometrically by oxidizing the Mn in acetic acid in the presence of ethylenediamine tetracetic acid with Na bismuthate to form a Mn^{III}-complexone complex which is ruby-red in colour. The method can be adapted to be used in the presence of small amounts of Co and Cr.—F. M. L.

***Microchemical Detection of Mercury, Copper, and Cobalt by Means of 2-Isatoxime.** V. Hovorka and L. Diviš (*Coll. Trav. Chim. Tchécoslov.*, 1950, **15**, (10/11), 589–598).—[In English]. 2-Isatoxime is a sensitive reagent for the qual. micro-detection of Hg, Cu, and Co. It gives with Hg a red to carmine salt insoluble in dil. HNO_3, with Cu a H_2O-insoluble brown salt which dissolves in alkali or NH_3 to give a violet soln. decolorized by cyanide, and with Co a green salt soluble in excess alkali and NH_3 to give a dark-green soln. stable to cyanide. The sensitivities of the tests and the reactions of many other ions with the reagent are quoted.—F. M. L.

***Catalytic Analysis. XIII.—Microdetermination of Molybdenum with Ukena's Colorimeter.** Taranobu Shiokawa (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 770–773).—[In English]. Cf. *ibid.*, (4), 613; *Met. Abs.*, this vol., col. 143. The catalytic action of Mo upon the reduction of malachite green to the colourless state with Ti^{3+}, is used in the microdetn. of Mo using Ukena's colorimeter. The most suitable amount of Ti^{3+} to use is ~3·5–4·0 ml. of 0·10*N*-soln. Ni^{2+}, Fe^{3+}, NO_3^-, Pd^{2+}, and V^{5+} ions interfere with the reaction, but Cr^{3+}, Zr^{2+}, do not.—J. S. G. T.

***(I.—) Photometric Determination of Nickel in Scrap Nickel Silver: A Rapid Method of Grading. (II.—) Note on the Grading of Scrap Nickel Silver.** (I) Charles Goldberg; (II) Charles Goldberg and Thomas Dobbins (*Chemist Analyst*, 1950, **39**, (3), 56–57; 1951, **40**, (2), 38).—(I.—) A 1-g. sample is dissolved in 10 c.c. 1 : 1 HNO_3 and the soln. taken to fumes with 3 c.c. $HClO_4$. The soln. is washed with 25% HCl into a 50-c.c. measuring flask contg. 4 g. Na hypophosphite (to reduce Cu^{II} to Cu^I), shaken, and then made up to vol. with 25% HCl. Any SiO_2 or metastannic acid is filtered off before the Ni is determined photometrically. (II.—) To avoid any pptn. of metastannic acid, the sample is dissolved in HCl and 30% H_2O_2 for the detn. of Ni as described above. —F. M. L.

***Spectrophotometric Determination of Nickel in Aluminium Alloys.** M. D. Cooper (*Analyt. Chem.*, 1951, **23**, (6), 880–883). —The sample is attacked with NaOH soln. and the soln. filtered. The residue is dissolved in *aqua regia* and the soln. fumed with H_2SO_4 and, if much SiO_2 is present, HF. After dilution, Ni may be determined directly, but preferably Cu is first removed by H_2S or electrolysis. To the soln. are then added ammonium citrate, ammoniacal dimethylglyoxime soln., and I soln. in KI and the transmittancy measured at 540 mμ against a ref. soln. contg. everything except dimethylglyoxime.—F. M. L.

***Reaction of Nickel II and Beta-Isothioureidopropionic Acid.** Lydia Jane Uhlig and Henry Freiser (*Analyt. Chem.*, 1951, **23**, (7), 1014–1016).—β-Isothioureidopropionic acid forms a coloured complex with Ni in ammoniacal soln. which permits the detn. of small amounts of Ni in the presence of Al, Zn, and Cu. Fe and Co interfere.—F. M. L.

Modified Vacuum-Fusion Apparatus for Determination of Oxygen, Hydrogen, and Nitrogen in Certain Metals. A. F. Torrisi and Jean L. Kernahan (*Analyt. Chem.*, 1951, **23**, (6), 928–929).—In a modified vacuum-fusion apparatus ground-glass joints are located at strategic points to facilitate assembly, cleaning, and repairing, and two measuring systems are attached to a single melting furnace and pumping system to give a doubled rate of analysis. The melting chamber used is similar to that described by Guldner and Beach (*ibid.*, 1950, **22**, 366; *Met. Abs.*, 1949–50, **17**, 751).—F. M. L.

The Estimation of Oxygen in Metals by Hydrogen Reduction. W. A. Baker (*Metallurgia*, 1949, **40**, (238), 188–189).—A detailed description is given of a modification of the original apparatus described by B. (*J. Inst. Metals*, 1939, **65**, 345), which increases the accuracy and speed of analysis.—S. R. W.

***Colorimetric Determination of Phosphorus in Metals and Alloys [Copper and Brass].** Jorma Kinnunen and Bertil Wennerstrand (*Chemist Analyst*, 1951, **40**, (2), 33, 35).—A 2-g. sample of Cu or brass is dissolved in HNO_3 or, if Sn is present, in a HNO_3–HCl mixture and the soln. is neutralized with NH_4OH. 10 c.c. HNO_3 and 3 c.c. 2% $KMnO_4$ soln. are added and excess $KMnO_4$ then decomposed by dropwise addn. of H_2O_2, avoiding excess. 10 c.c. of a 0·25% soln. of NH_4 vanadate in 2% HNO_3 are added and the soln. boiled. After cooling the soln. 10 c.c. 10% NH_4 molybdate is added and the molybdivanado-phosphoric acid extracted into a 10 : 1 ether–butanol mixture. The extract is washed with 25 c.c. 10% HNO_3 and the transmittancy of the soln. then measured with blue filter against a blank on the reagents. —F. M. L.

***Amperometric Titrations of Sn^{IV} and Hg^{II} with Tetraphenylarsonium Chloride.** I. M. Kolthoff and Ralph A. Johnson (*J. Electrochem. Soc.*, 1951, **98**, (6), 231–233).—It is shown that Sn^{IV} and Hg^{II} can be titrated amperometrically with tetraphenylarsonium chloride using the dropping Hg electrode as the indicator electrode. Full details are presented of the procedures to be used in each case. In that of Sn^{IV} the standard deviation found was ±0·13 mg. and the mean deviation ±0·11 mg. In the range 15–35 mg. Sn, the errors approach a random distribution about zero; at higher concentrations of 35–50 mg. Sn there is a definite trend towards negative results, but greater accuracy can be obtained in this range by using a suitable coeff. of regression. Interfering ions include: Fe^{III}, Hg^{II}, Bi, Sb^{III}, Sb^{IV}, Cu^{II}, arsenious, uranyl, molybdate, vanadate, and selenite. Pb also interferes, but if present in concentrations of $<0\cdot005$ *M* the titration may be made by measuring currents at $-0\cdot4$ V. Ions which precipitate tetraphenylarsonium ion also interfere. In the detn. of Hg^{II}, the standard deviation is $\pm 0\cdot47$ and the mean deviation $\pm 0\cdot38$ mg. The errors are randomly distributed about zero in the range 30–80 mg., but in samples contg. >80 mg. the errors are significantly negative. Interfering ions include, Fe^{III}, Bi, Cu^{II}, uranyl, molybdate, vanadate, and selenite and in addn. those which precipitate the tetraphenylarsonium ion in $3N$-HCl. Nitrate and sulphate do not interfere. Attempts to determine Cd and Zn using the same method were unsuccessful. 4 ref.—G. T. C.

***Inorganic Chromatography on Cellulose. V.—The Use of Columns of Cellulose in Combination with Organic Solvent Extraction for the Separation of Uranium from Other Metals.** F. H. Burstall and R. H. Wells (*Analyst*, 1951, **76**, (904), 396–409; discussion, 409–410).—Cf. *ibid.*, p. 388; *Met. Abs.*, this vol., col. 216. U is separated from other metals by extraction of uranyl nitrate with ether contg. 5% HNO_3 by vol. on a cellulose column. Methods of overcoming difficulties due to interfering ions are fully described. Recommended methods are given for the detn. of U in several ores.—F. M. L.

***Ultra-Violet Spectrophotometric Determination of Vanadium.** George Telep and D. F. Boltz (*Analyt. Chem.*, 1951, **23**, (6), 901–903).—The ultra-violet absorption spectra of the peroxyvanadic complex has been studied and an absorbency max. at 290 mμ found which permits detn. of V in small amounts in the presence of greater amounts of Ti and Cr_2O_7'' than can be tolerated in the visible region.—F. M. L.

***3 : 3′-Dimethylnaphthidine as a Reagent for the Detection of Vanadium.** R. Belcher, A. J. Nutten, and W. I. Stephen (*Analyst*, 1951, **76**, (904), 430–431).—3 : 3′-dimethylnaphthidine gives a red-violet colour with VO_3' which is sensitive to 1 part in 500,000. Other oxidizing agents interfere. —F. M. L.

***The Accuracy of the Titrimetric Determination of Zinc with Potassium Ferrocyanide.** Yoshinaga Oka and Takuji Kanno (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], **2**, (5), 802–808).—[In English]. When $ZnSO_4$ soln. is titrated against $K_4Fe(CN)_6$ soln., using diphenylamine as indicator, the end-point cannot be determined accurately owing to gradual decolorization of the indicator. Addn. of NH_4Cl makes the colour change distinct, but accurate results are still not attainable. About 7% of the Zn was found present as $Zn_2Fe(CN)_6$. Excess of $K_4Fe(CN)_6$ favours the formation of $K_2Zn_3[Fe(CN)_6]_2$. Satisfactory results are obtained when $K_4Fe(CN)_6$ soln. with Zn, or a small excess of ferrocyanide, is added to the Zn soln. in one lot, and the excess of ferrocyanide titrated with $ZnSO_4$ or $K_2Cr_2O_7$ soln. Of the sulphates, chlorides, and nitrates of NH_4, K, and Na, only Na_2SO_4 can be present without affecting the accuracy of the titration; NH_4Cl produces the biggest effect. The effects of amounts of indicators, $K_4Fe(CN)_6$, and various concentrations of acids are studied.—J. S. G. T.

***Colorimetric Determination of Zinc and Cobalt in Gold Alloy Plating Solutions.** J. Mermillod (*Plating*, 1951, **38**, (7), 699–700, 703).—Full experimental details are presented of the detn. of Zn, using dithizone, and of Co, using 1-nitro-2-naphthol sodium disulphonate. In the former case, several interfering ions may be encountered, and special procedures are needed to eliminate them. The method enables the Zn content of the soln. to be determined to within 0·2 γ. It is not claimed that the method is any more accurate or any faster than the grav. method, but when a complete analysis is required the colorimetric method needs only a few more minutes' work after the other elements have been determined. In the method given for the detn. of Co, neither Cu nor Au ions interfere. 3 ref.—G. T. C.

***Colorimetric Estimation of Various Metal Derivatives of Sodium Diethyldithiocarbamate.** Rene J. LaCoste, Mason H. Earing, and Stephen E. Wiberley (*Analyt. Chem.*, 1951, **23**, (6), 871–874).—Bi, Co, Cr, Fe, Ni, and U, as well as Cu, can be accurately determined colorimetrically, using Na diethyldithiocarbamate and subsequent extraction in chloroform. It has been found that, since extraction of these metals is quant. over a wide range of acidity, unless they are removed they will interfere in the detn. of Cu, although by careful control of acidity Cu can be separated from U and Cr.—F. M. L.

Solution Methods of Spectrographic Analysis. —— (*Metallurgia*, 1951, **44**, (261), 45–49).—In a communication from the British Non-Ferrous Metals Research Association, an account is given of investigations carried out by a Panel into the applicability of soln. methods to spectrographic analysis. Four methods were studied, viz.: the cup method, the drip-feed method, the powder-spark method, and the Scheibe method (using an impregnated short graphite electrode). The cup and powder-spark methods were selected as most suitable for met. analysis. The appn. of the cup method is described with ref. to the detn. of Cr in Cr bronze, Pb and Zn in Cu alloys, Ca in Al, Mg, and Zn, B in Cu and Al, Li in Cu and Cu alloys, and Cd in Cd–Cu alloys. The powder-spark method is illustrated by descriptions of the detn. of Fe, Mg, Mn, and Si and Be in Al alloys, Ni in Cr bronze, and B and Al in Cu alloys, and the analysis of high-purity Al.—F. M. L.

***Inorganic Chromatography on Cellulose. IV.—Determination of Inorganic Compounds by Paper-Strip Separation and Polarography.** J. A. Lewis and J. M. Griffiths (*Analyst*, 1951, **76**, (904), 388–395; discussion, 395).—Cf. *Discussions Faraday Soc.*, 1949, (7), 179; *Met. Abs.*, 1949–50, **17**, 842. Sepn. are described in which problems of interference in quant. analysis are overcome by using paper-strip chromatography. Three types of sepn. are described, viz.: (*a*) the sepn. of one metal from a mixture exemplified by the sepn. of U from a large number of other metals; (*b*) the sepn. of several metals in a mixture, e.g. Co, Ni, and Cu in alloy steel and Cu and Co in pyrites; (*c*) the sepn. of a mixture of metals into groups contg. several metals that can be determined polarographically without further sepn., exemplified by the sepn. of ten metals into two groups contg. resp. V, Cu, U, Pb, and Ti and Fe, Mo, Bi, Sb, and Cd. The use of salicylic acid as a supporting electrolyte for polarography is described and discussed.—F. M. L.

***Polarography with Alternating Currents. I.—Outline of Theory, Apparatus, and Techniques. II.—A.C. Polarography of Cadmium, Zinc, Lead, Thallium, Indium, Bismuth, and Antimony.** B. Breyer, F. Gutman, and S. Hacobian (*Australian J. Sci. Research*, 1950, [A], **3**, (4), 558–566, 567–580).—[I.—] A sinusoidal alternating L.F. field was superposed on the direct potential applied to a dropping Hg electrode. Equations given enable calculation of ion concentration from the max. differential A.C. at the characteristic " summit potential " (analogous to $E_{\frac{1}{2}}$ in conventional polarography). This current is proportional to the ion concentration of the ion species being discharged. Advantages are: (1) dissolved O in soln. does not interfere; (2) single reading instead of plotted curve; (3) small concentrations of less noble ions can be determined in excess of nobler ions; (4) waves 40 mV. apart separable; (5) improved reproducibility; (6) easy recalibration; and (7) no delicate galvanometer. The sensitivity limit is $10^{-5}N$, with a working range of 10^{-3}–$10^{-4}N$. 12 ref. [II.—] Factors affecting max. differential current $i_{D\sim}$ in A.C. polarographic waves of Cd, Zn, Pb, Tl, In, Bi, and Sb are: (1) series resistance in cell circuit; (2) A.C. voltage amplitude; (3) temp. variation; (4) variation of supporting electrolyte; and (5) the presence or absence of air. A mixture of all seven ions is separable in concentrations of $<10^{-3}N$. 8 ref.—S. M.

Some New Polarometric Titrations. R. Kalvoda and J. Zýka (*Coll. Trav. Chim. Tchécoslov.*, 1950, **15**, (10/11) 630–638).—[In English]. Procedures are described for the polarometric titration of Cu with $K_4Fe(CN)_6$, of Tl with KI and $K_2Cr_2O_7$, of Ag with NH_4CNS, $K_4Fe(CN)_6$, and $K_2Fe(NO)(CN)_5$, and of WO_4'' with $Pb(NO_3)_2$. The diffusion current during the titration is measured using a dropping Hg electrode with a saturated HgCl electrode as ref. 21 ref. —F. M. L.

12 — LABORATORY APPARATUS, INSTRUMENTS, &c.

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

A Simple Temperature-Controlled Laboratory Furnace. R. S. Barnes (*J. Sci. Instruments*, 1951, **28**, (3), 89–92).—The furnace described was wound with Ni–Cr and Pt wires threaded through twin-bore alumina quills and then wrapped around a thick-walled Ni–Cr cylinder. The Ni–Cr and Pt windings function as two arms of an A.C. bridge. Because the temp. coeff. of the two windings are different, the impedance ratio of the other two arms governs the balance temp. of the bridge. A relatively simple control circuit is required. During 8 hr. with mains fluctuations of 13 V. and an ambient temp. change of 3° C., the temp. inside the furnace block remained between 750·9° and 751·1° C. Below 800° C. no detectable drift in several days is claimed, but above this temp. the resistance of the Pt wire steadily increases with time.—P. C. L. P.

A Vacuum Furnace for Use in the Temperature Range 1000°–2000° C. J. W. Tomlinson and J. O'M. Bockris (*Rev. Sci. Instruments*, 1950, **21**, (6), 507–508).—A laboratory vacuum furnace with a glass envelope and a heater of W wire is described.—E. J.

An Improved Vacuum Fusion Furnace. W. G. Guldner (*Bell Lab. Record*, 1951, **29**, (1), 18–20).—G. describes a furnace in which the charge is contained in a graphite crucible (¾ in. dia., 3 in. long) surrounded by graphite powder within a quartz thimble. The thimble is suspended within a Pyrex tube which has a plug in the base, normally waxed into position, through which the thimble may be removed. 2650° C. has been attained with a 5-kW., 525-kc. generator. —D. M. D.

Preparation of Single-Crystal Copper Ribbons from Single-Crystal Copper Rods. A. A. Petrauskas, E. A. Coomes, and J. E. MacDonald (*Rev. Sci. Instruments*, 1949, **20**, (12), 961–962).—A method is described whereby a rod having a known orientation of the crystallographic planes with respect to the axis is carefully machined, and oxidized and etched in suitable reagents repeatedly until a smooth single-crystal Cu ribbon, 5 in. long, 0·13 in. wide, and ~0·003 in. thick, is obtained.—E. J.

***[Furnace for the] Growth of Molybdenum Single Crystals.** N. K. Chen, R. Maddin, and R. B. Pond (*J. Metals*, 1951, **3**, (6), 461–464).—The construction and operation of a modified Andrade furnace for the prodn. of single-crystal rods of Mo are described. It is capable of producing single crystals ~1½ in. long in the relatively short time of 7 hr., when the centre 4 in. of the 6-in.-long rod is heated to an average temp. of 2000° C., and the ends are kept at ~35° C. It is not known whether the steep temp. gradient or the high temp. is responsible for the rapid grain-growth obtained. 3 ref.—E. N.

***The Melting Point and the Density of Neptunium Metal : A Micro-Melting-Point Apparatus for Metals.** E. F. Westrum and L. Eyring (*U.S. Atomic Energy Commission Publ.*, 1950, (**AECD-2922**), 8 pp. ; also (**UCRL-1056**) ; and *J. Amer. Chem. Soc.*, 1951, **73**, (7), 3399–3400).—The metal was produced by reduction of NpF_3 with Ba at ~ 1300° C. and contained various impurities. It was formed as silvery globules in mg. quantities, and the m.p. was determined by clamping a globule in a 35-mil-dia. W wire " hairpin " (the jaws of which were ground plane and polished), and heating in a high vacuum. The temp. at which the jaws converged was taken as the m.p. and values of 650° and 419° C. were obtained for Mg and Zn, resp. The m.p. of various samples of Np was in the range 639°–641° C. The *d* was determined using the displacement of dibutyl phthalate in a capillary tube and was in the range 16·9–20·1 g./c.c. for 4 samples of different prepn. The lower *d* may be due to a metallic phase stable at higher temp. and retained owing to the presence of an impurity, and the higher values were considered representative of the *d* of the room-temp. form. X-ray photographs gave a f.c.c. structure with $a = 4{\cdot}89$ Å. and a calculated *d* of 14·7 g./c.c., but this is probably due to an oxide film.—B. W. M.

***Density Control in the Manufacture of Rhodium [Optical] Filters.** W. Zehden (*Vacuum*, 1951, **1**, (1), 38–39).—A letter. Rh is a suitable material for making high-quality opt. filters by evaporation and deposition *in vacuo*. It is found, however, that : (1) the *d* of the deposit increases for some time after the end of the deposition, until a final stable value is attained ; (2) burnishing alters this final value ; and (3) the comparatively low values of the *d* at the edges reduces the mean value of the *d* of the film, more especially in the case of small filters.—J. S. G. T.

The Deposition of Rhodium on Perspex. B. L. Ginsborg and O. S. Heavens (*Rev. Sci. Instruments*, 1951, **22**, (2), 114–115).—The successful coating of Perspex by the evaporation of Rh from specially prepared W spirals is described. A hard, durable, opaque layer, ~1000 Å. in thickness, can be obtained.—E. J.

A Method of Annealing Metal Powders without Sintering [Prior to X-Ray Examination]. F. W. Von Batchelder and H. E. Strauss (*Rev. Sci. Instruments*, 1951, **22**, (6), 396–397).—Before the X-ray detn. of lattice parameters, it is necessary to anneal the metallic powders, and by mixing the powder with a larger quantity of a powder which is chem. inactive to the metal, annealing, and then separating the inactive powder by dissolving, the sintering of the metal is prevented. Examples of the use of the technique are given.—E. J.

Apparatus for the Preparation of Metals with an Exactly Known Content of Impurities. J. D. Fast (*Philips Tech. Rev.*, 1950, **11**, (8), 241–244).—F. describes equipment for the prepn. of extremely pure Fe and for the addn. thereto of known quantities of impurities. Melting is done by means of induction heating, and the melting process can be followed throughout on account of the crucible being contained in a water-cooled glass envelope. The influence of C, N, and O, alone and in combination, upon the properties of Fe, can readily be studied.—D. K. W.

An Electronic Instrument for the Measurement of the Damping Capacity of Materials. A. D. N. Smith (*J. Sci. Instruments*, 1951, **28**, (4), 106–109).—Equipment is described which will measure the time interval for the amplitude of a vibration to decay to one-half its initial value. It is applicable to the frequency range 50–10,000 c./s., and could be modified to extend to 100,000 c./s.—P. C. L. P.

An Oscillation-Type Magnetometer. J. H. E. Griffiths and J. R. MacDonald (*J. Sci. Instruments*, 1951, **28**, (2), 56–58).—A simple method of measuring the saturation magnetization of thin samples of ferromagnetic materials is described which has been applied to Ni and Supermalloy sheet 0·11 mm. thick and to evaporated Ni films 0·1–2 μ thick. The results obtained agreed with values taken from the literature.—P. C. L. P.

A Simple Goniometer for the Study of Crystal Surfaces. P. R. Rowland (*J. Sci. Instruments*, 1951, **28**, (2), 61–62).—In the instrument described, the specimen is mounted in the centre of a transparent Perspex sphere. Measurement of the distance along the surface of the sphere, between reflections from a beam of light, gives the angle between reflecting faces. The reflections from an etched metallographic specimen of Cu established, without calculation, the orientation of the crystals, the fact that they were twinned, and the twinning plane. —P. C. L. P.

An Application of Geiger Counter Tubes for Spectrochemical Analysis. O. G. Koppius (*Philips Tech. Rev.*, 1950, **11**, (7), 215–219).—Describes equipment for the continuous detection of small traces of Pb in indust. atmospheres. Air is drawn through a continuous spark discharge between two Cu electrodes, and the resulting radiation is analysed by means of a quartz spectrograph and a Geiger counter, focused on the portion of the spectrum contg. the strongest Pb line, i.e. at 2203 Å. The Geiger counter is connected to a counting-rate

meter. The sensitivity of the apparatus is limited to ~0·60 mg. Pb/m.^{3} air, but the versatility is far superior to that of the more common and sensitive photographic apparatus.
—D. K. W.

A Simple Automatic Recording Microphotometer. R. M. Fisher and D. S. Miller (*Rev. Sci. Instruments*, 1950, **21**, (11), 938–939).—By using a balanced opt. and electronic system, a precision microphotometer has been constructed, and its use and advantages are described.—E. J.

***An Emission Microscope for Photoelectron Autoradiography.** A. N. Barker, H. O. W. Richardson, and N. Feather (*Research*, 1950, **3**, (9), 431–432).—A letter. Contact autoradiography is limited in resolution, and magnification is only unity. Magnified autoradiography is possible when the radioactive source emits electrons of a single energy. A microscope operating on this principle is briefly described; the specimen is placed in the lower gap of a two-component magnetic lens, the field of which is maintained between soft Fe pole pieces by horseshoe-type permanent magnets. The performance of the microscope was studied using Th active deposit as source material. The image of a brass button activated with radio-Th is reproduced at an electron-opt. magnification of ×7: a best resolution of ~6 μ was achieved, shown by a study of the image of intersecting grooves engraved on the button.—F. A. F.

An Exposure Meter for the Electron Microscope. E. H. Frei and F. L. Hirshfeld (*Rev. Sci. Instruments*, 1951, **22**, (4), 231–232).—A simple electrostatic exposure meter is described and by its use the characteristic of the photographic plate has been plotted at const. intensity of the electron beam.—E. J.

A Single-Crystal X-Ray Camera for Direct Recording of the Reciprocal Lattice. J. M. Torroja, E. Pajares, and J. L. Amorós (*J. Sci. Instruments*, 1951, **28**, (2), 44–46).—In the camera described the crystal is rotated about a zone axis and the film rotated synchronously about a parallel axis. An annular slit is placed between the specimen and the film to prevent more than one diffraction cone recording at a time. The path lengths traversed by diffracted beams from each reciprocal lattice plane are equalized by varying the angle between the axis of rotation and the collimated incident X-ray beam. By this procedure undistorted records of the reciprocal lattice are obtained which can be indexed at sight.—P. C. L. P.

Low-Temperature X-Ray Diffraction Apparatus. Donald F. Clifton (*Rev. Sci. Instruments*, 1950, **21**, (4), 339–342).—When studying transformations at sub-zero temp., it is often convenient to carry out treatments such as cold working, &c., and to transfer the specimen still at low temp. to the X-ray-diffraction apparatus for examination. Two suitable apparatuses are described. One uses cold gas blown over the specimen and the other, having a working range from 77° to over 375° K., uses liq. N as a coolant. Temp. control is effected by balancing the heat conduction along a tube with a heater near the specimen. The methods are compared. 20 ref.—E. J.

On Adapting a Powder X-Ray Diffraction Camera for Reflection Patterns. Jonathan Parsons (*Rev. Sci. Instruments*, 1950, **21**, (2), 185–186).—A method is described by which an X-ray-diffraction powder camera may be adapted, easily and quickly, to accommodate small flat mounts. The special sample mount makes it possible to examine variations in the compn., &c., of the specimen.—E. J.

A Monochromator for X-Ray Powder Camera. Frances M. Wrightson and I. Fankuchen (*Rev. Sci. Instruments*, 1951, **22**, (3), 212–213).—The design and construction of a monochromator assembly is described in detail, and its performance compared with a camera using filtered radiation when X-ray fluorescent material is present in the powders under study.
—E. J.

The Mounting and Centring of Specimens in High-Temperature X-Ray Powder Cameras. E. G. Steward (*J. Sci. Instruments*, 1951, **28**, (1), 29–30).—A device made in Monel metal is described, to support and centre specimens for examination in the Unicam high-temp. X-ray powder-diffraction camera.
—P. C. L. P.

Optical Centring System for X-Ray Diffraction Samples. Eugene L. Perrine (*Rev. Sci. Instruments*, 1950, **21**, (3), 262–263).—The positioning of a sample on the axis of rotation for X-ray diffraction studies is facilitated by the use of an opt. system which is described in detail.—E. J.

An Electrodeposited Surface Roughness Standard. P. M. Aitchison (*Australian J. Appl. Sci.*, 1950, **1**, (1), 71–74).—Strips of Cr 0·0035 in. in width and spacing were plated on a slip gauge for 10 min. at 2·8 amp./in.^{2}, giving a depth of 0·00015 in., followed by lapping. A wax developed by the British Iron and Steel Research Association was found best for stopping-off. The abs. height of the strips can be measured to better than 0·1 micro-in. (0·0025 μ).—S. M.

Two Improved Methods of Graticule Production, Using Evaporated Metal Films. P. M. Aitchison and R. E. Aitchison (*Australian J. Appl. Sci.*, 1950, **1**, (1), 75–79).—An undercoat of 0·00003 in. Zn followed by a coat of a wax resist developed by the British Iron and Steel Research Association prevents rounding of edges during etching of ruled lines. Lines of similar sharpness are obtained by depositing 0·00003 in. Zn, followed by ruling, then depositing 0·000005 in. Cr. As the Cr deposited on the clean glass is not affected, that with the Zn undercoat can be removed with 0·01N-NaOH soln. at 50° C. Inferior adhesion of Cr occurs if wax is used instead of the Zn, although the method is feasible.—S. M.

A Mechanical Arrangement for Measuring Small Changes in Length. Günther Leibfried (*Z. Physik*, 1950, **127**, (5), 580–587).—A description is given of a mech. apparatus for measuring small changes in length of a rod or wire under tension. Two points on the rod are fastened to two slides carrying parallel plates, so that as the rod extends these move relative to each other. Between the plates are two fine Cu wires (dia. 0·01–0·05 mm.) which rotate under the motion of these plates. Each wire carries a mirror, so that slight rotations can be measured by an optical lever, of length 8 m. Oil damping is used. By this means a magnification of 3×10^5 is obtained. The apparatus is stable in use, and changes of 10 Å. can be detected. The apparatus is used to detect the flow of 2 Al single crystals. No sharp changes in elongation were observed. The elongation curve shows a wavy structure, with a time period of between 1 sec. and 1 min.—A. C.

13—PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

Hardness Testing in Relation to Anisotropy of the Indenter. H. Meincke (*Metalloberfläche*, 1951, [A], **5**, (2), 17–21).—Consideration of the results—mainly of other workers—of various indentation hardness tests leads to the following conclusions: (1) the Knoop indenter and, more particularly, the Grodzinski double-cone indenter are most reliable for single-crystal and anisotropic polycryst. materials; (2) tensile strengths can be more accurately calculated from the results of tests which produce almost unidirectional indentations; and (3) microhardness tests, also, are best carried out with almost unidirectional indenters; the Vickers 136° pyramid gives erroneous results. 15 ref.—E. N.

How Can Hardness Values Be Used in Metal Design? John B. Campbell (*Materials and Methods*, 1950, **32**, (6), 43–47).—The difficulties of defining hardness are described, and the relations between indentation hardness and other mech. properties are discussed.—J. W. C.

***A Comparison of Mohs' Scale of Hardness with the Results of Other Hardness Tests.** Nikolaus Ludwig (*Metalloberfläche*, 1951, [A], **5**, (3), 38–42).—An attempt is made to correlate the results (mainly obtained from the literature) of various hardness tests with Mohs' scale of hardness. The hardness tests considered are: (i) scratch (Bierbaum), (ii) indentation (Brinell, Rockwell, Vickers, and Knoop), and (iii) Shore scleroscope. Correlation between Mohs' hardness and: (1) scratch hardness was poor, (2) Brinell and Rockwell hardness could be obtained over only a limited range, (3) Shore hardness was non-existent, and (4) Vickers and Knoop hardness was good, especially when the tests were carried out under light loads, i.e. microhardness testing. Chruschtschow's formula: $M = 0{\cdot}7\sqrt[3]{H_V}$, for the relationship between Mohs' hardness (M) and Vickers hardness (H_V) appears to be valid. 23 ref. —E. N.

***An Automatic Light-Load Bergsman-Type Hardness Tester.** P. Grodzinski (*J. Sci. Instruments*, 1951, 28, (4), 117–121).—A Bergsman microhardness tester was modified to make it more robust, fitted with automatic loading and unloading controls, and attached to a Vickers projection microscope. A double-cone diamond indenter, giving boat-shaped impressions, whose length was measured, was found to be less liable to fracture hard materials than either the Vickers or Knoop indenters. Comparisons are made between the hardness investigated with the double cone and with the Vickers indenter on sintered carbide H and on high-speed steel, in the range of loads 10–1000 g. A sharp drop of apparent hardness was found with loads $<$100 g. —P. C. L. P.

***A Micro-Indentation Hardness Tester for Attachment to the Vickers Projection Microscope.** S. J. Lloyd and D. J. Norris (*J. Sci. Instruments*, 1951, 28, (3), 81–84).—A beam-loading instrument is described, capable of applying loads in the range 0·1–10 g. and, with a different pivot, from 10 to 50 g. The 136° pyramid diamond indenter used could be aimed with a precision $>$1 μ. The duration of appn. of the load is controlled by a cam, and the rate of approach of the indenter to the specimen is limited to a max. of 1·0 mm./min. by a two-cylinder oil dash-pot. Impressions were made in coarse-grained electrolytically polished specimens, with loads of 0·1–1·2 g. in Al and in β-brass with loads of 1·0–10 g. The instrument gave consistent results provided that the impression diagonal exceeded 6 μ, corresponding to loads of 0·4 g. on Al and 2·5 g. on β-brass. The discrepancies at lower loads to give unduly large impressions may be due to vibration or to inertia of the beam.—P. C. L. P.

***Results of Shear Fatigue Tests of Joints with $\frac{3}{16}$-in.-Diameter 24S-T31 Rivets in 0·064-in.-Thick Alclad Sheet.** M. Holt ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2012), 45 pp.; *Appl. Mechanics Rev.*, 1951, 4, 93).—Rivet shear fatigue tests were made of simple lap joints with a single $\frac{3}{16}$-in.-dia. 24S-T31 rivet in 0·064-in.-thick Alclad sheet. 24S, 75S, and 14S sheets were tested in several tempers. Designs were varied to include countersunk-head rivets in drilled holes and in dimpled holes, and brazier-head rivets. S/N curves for various alloys and tempers were found to lie fairly close together in relatively narrow scatter bands. Design of joint gave a wider range of fatigue strength than did the choice of material; no one sheet alloy showed superiority over the others.

***The Measurement of Young's Modulus of Metals and Alloys by an Interferometric Method. I.—Young's Modulus of Nickel–Copper Alloys.** (Fukuroi and Shibuya). See col. 181.

***Determination of Elastic Constants of Solids by Pulse Method.** — Krishnaji (*Proc. Nat. Inst. Sci. India*, 1950, 16, (4), 227–234).—An apparatus is described for measuring the elastic const. of solids by means of ultrasonics. A pulse of H.F. sound waves is passed through a plane parallel plate of the solid suspended in a liq. bath. The angle of incidence of the beam is changed by rotating the solid plate about a vertical axis. At certain angles of incidence the longitudinal and shear waves are totally reflected, thereby giving min. in the transmitted energy. From these angles of incidence giving total reflection, together with the velocity of sound in the liquid, the elastic const. can be calculated. Results are given for measurements on Al, brass, and Fe. 6 ref.—E. C. P.

***Ferromagnetic Metals: Identification by Measurement of Internal Stresses. I.—Magnetic Tests. II.—Mechanical Tests.** Albert Borowik (*Iron and Steel*, 1948, 21, (1), 3–6; (2), 39–43; (3), 81–86; (4), 117–121).—The work described was carried out on a range of steel wires, but the methods used are of more general interest.—N. B. V.

Compensating for Inaccuracies of Measurement in G. Sachs's Boring Method for Determining the Internal Stresses in Bars and Tubes. H. Bühler and W. Schreiber (*Metall*, 1951, 5, (3/4), 53–57).—The inaccuracies of measurement are discussed from a math. standpoint, and S's. formulæ (*Z. Metallkunde*, 1927, 19, 352; *J. Inst. Metals* (Abstracts), 1928, 39, 648) are corrected accordingly. 10 ref.—E. N.

That Specification: Master or Servant? [With Particular Examples from the Cold-Rolled Non-Ferrous Industry]. Thomas B. Crow (*Metallurgia*, 1950, 42, (248), 14–17; (249), 59–63; correspondence, (250), 144–145).—A general discussion of the uses and limitations of specifications. They are not a guarantee of suitability and safety, except where the user is well versed in the met. technique to be employed. Examples of the use of specifications of cold-rolled material are given in which dissatisfaction arises through lack of knowledge, co-operation, or sufficient attention to the requirements of the job. Mis-appn. or mis-interpretation may even lead to difficulties. C. views certain aspects of specifications critically, e.g. ambiguity of wording, testing, tolerances, tempers, and analysis, and suggests that the purchaser should explain to the metal-roller exactly what he wishes to make and how he intends to make it and that the manufacturer should choose the supply of material which can be tested before purchase. Thus a satisfactory individual specification can be evolved.—S. R. W.

International Conference on the Non-Destructive Testing of Materials, at Saarbrücken, 30 Nov.–2 Dec. 1950. H. J. Seemann (*Metall*, 1951, 5, (3/4), 68–69).—Brief reports on the following papers presented at the meeting: *R. Berthold*: "Non-Destructive Methods for Measuring the Thickness of Walls and Coatings"; *H. Trost*: "The Counter-Tube as an Accessory for Testing Materials"; *Th. Wuppermann*: "Thickness Measurements During the Production of Tinplate"; *O. Vaupel*: "Experiences of Radiographic Inspection in Berlin and the Eastern Zone of Germany Since 1945"; *H. Gerbeaux and M. Evrard*: "Experiences in Inspecting Materials with Gamma Radiation"; *C. Brachet*: "The Use of Irradiated Materials for the Radiological Inspection of Large Objects"; *F. Gottfeld*: "The Use of Radon and Radioactive Isotopes in England for Non-Destructive Testing"; *A. Guinier*: "The Role of Fine-Structure Investigations in Non-Destructive Testing"; *C. Legrand*: "The Investigation of Surface Films by X-Rays and the Details of a Suitable Type of Debye Camera for This Purpose"; *H. Möller*: "New Possibilities in the Measurement of Stresses by X-Rays, by the Use of Short-Wave Radiation"; *J.-J. Trillat*: "Micrography with Secondary Electrons"; *H. J. Seemann*: "Physico-Metallurgical Considerations in the Ultrasonic Testing of Materials"; *P. Bastien*: "Sources of Error in the Ultrasonic Testing of Metallic Materials"; *I. Krautkrämer*: "A New Ultrasonic Testing Instrument Based on the Impulse-Reflection Principle, and Some Results Obtained with It"; *G. A. Homès, J. Ots,* and *E. Symon*: "Belgian Methods of Ultrasonic Testing"; *R. Pohlman*: "Sources of Error in Acoustic Methods of Testing"; *R. Cabarat*: "An Acoustic Method for Determining the Elastic Modulus and the Internal Friction of Solids"; *J. Jellinghaus*: "Magnetic and Electrical Methods of Non-Destructive Testing"; *F. Förster*: "New Methods of Non-Destructive Testing Based on Magnetic Properties"; *H. de Leiris*: "Methods of Inspection Based on the Principle of the Milk of Lime Process"; *A. Forster*: "Methods and Instruments for the Practical Testing of Surfaces"; and *E. Lœuille*: "Spectral Analysis as an Accessory in Non-Destructive Testing".—E. N.

Fluoroscopic Inspection of Light-Metal Alloys. Justin G. Schneeman and T. E. Piper (*Non-Destructive Testing*, 1950-51, 9, (3), 20–22).—Fluoroscopy saves 75% of film inspections on light alloy castings. Using direct-viewing of the fluoroscopic screen (through lead glass), sections of Mg and Al may be inspected at a source–screen distance of 20 in., with 100 kV. (10 m.amp.) X-rays. By increasing the penetration to 150 kV. (10–15 m.amp.), thicknesses up to 1½ in. may be examined. Routine fluoroscopy permits the detection of defects $\nless$0·080 in. in thickness or $\nless$10% of the specimen thickness, whichever is the greater. The technique will not replace radiography, which is essential for the detection of micro-shrinkage in Mg alloys, and of fine porosity in Al alloys.—L. M.

Gamma-Radiography in Shipbuilding and Engineering. J. D. Hislop (*Trans. Inst. Marine Eng.*, 1951, 63, (5), 83–87; discussion, 87–96; and (abridged) *Engineering*, 1951, 171, (4442), 327–330).—H. defines the terminology employed in radiography, and indicates the relative activities and "half-value" periods of the various γ-ray sources available to industry. The factors governing the correct choice of source are discussed, and techniques employed for the examination of typical castings are described. H. compares the merits of methods applying γ-ray sources with those using X-rays and outlines the limitations of both techniques. —D. K. W.

Radiography of Welded Joints for Power Plants. J. H. Lawson (*Australasian Eng.*, 1951, (Feb.), 78–82).—Equipment and techniques are described.—T. A. H.

Industrial Radiography: Safety Precautions. R. King (*Iron and Steel*, 1949, 22, (14), 633–635).—N. B. V.

14 — TEMPERATURE MEASUREMENT AND CONTROL

***Thermoelectric Properties of Titanium, with Special Reference to the Allotropic Transformation.** (Worner). See col. 168.

A Two-Colour Infra-Red Radiation Pyrometer. A. F. Gibson (*J. Sci. Instruments*, 1951, 28, (5), 153–155).—If it is assumed that the emissivity is independent of the wave-length, the surface temp. of a body can be determined from the ratio of the radiated energy in two narrow wave-length bands. A null method is adopted in the apparatus described which uses a PbS photoconductive cell to detect radiation of wave-length 1·3–2·6 μ. With suitable choice of filters, temp. of 200°–1200° C. can be measured to within ±3° C. At lower temp. the signal : noise ratio decreases until at 150° C. the accuracy is only ±25° C. Possible modifications and improvements are discussed.—P. C. L. P.

A Method of Temperature Programme Control for Thermal Analysis. J. G. Ball and E. T. Adams (*J. Sci. Instruments*, 1951, 28, (2), 47–49).—A method to give a pre-arranged temp. programme smoothly in systems of small heat capacity is described. A cam, which is attached to the shaft of a variable transformer, is driven by a spindle rotating at a const. speed. To prevent slipping, the Tufnol cam edge was trimmed with thin rubber sheet and shallow flutes were cut along the spindle. Variation of the cam radius alters the rate of change of applied voltage; this enables const. rates of heating and cooling to be obtained. Using an amplified differential couple in conjunction with a direct-reading thermocouple, temp. of 663° and 657° C. were found for the melting and solidification points, resp., of a 0·05-c.c. sample of pure Al.—P. C. L. P.

The Optical Temperature Scale and the Radiation Constants. Fr. Hoffmann (*Z. angew. Physik*, 1950, 2, 88–95; *C. Abs.*, 1950, 44, 5658).—The decision in 1948 of the International Bureau of Weights and Measures to make Planck's radiation const. $C_2 = 1{\cdot}438$ cm.°C. was premature, not so much because this value is higher than the previous one (1·432) as because of a lack of agreement between the theoretical and experimental values of the Stefan–Boltzmann const. σ. This introduces an uncertainty of at least 2° C. in the m.p. of Au and makes imperative a comparison of the opt. pyrometric and gas thermometric scales below the Au point.

Temperature Measurement: New Instruments and Developments in German Practice. —— (*Iron and Steel*, 1948, 21, (10), 421–423).—N. B. V.

15 — FOUNDRY PRACTICE AND APPLIANCES

Aluminium Alloy Die-Castings. [—I.–II.]. Floyd A. Lewis (*Foundry*, 1951, 79, (4), 124–127, 271–274; (5), 96–100, 230–235).—[I.—] L. discusses the appn. of Al alloy die-castings and describes the plant and techniques used in their manufacture. The alloys used and their properties are considered. [II.—] Modern plant and techniques for the prodn. and after-treatment of Al-alloy pressure die-castings are described. The design of the castings is also considered. 10 ref.—R. W. R.

How Beryllium Copper Parts Are Designed for Investment Casting. John T. Richards (*Machinist* (*Eur. Edn.*), 1951, 95, (28), 1047–1049).—R. summarizes the type of investment-cast parts that are designed to take advantage of the particular properties of Cu–Be alloys, discusses the design limitations and advantages inherent in this method of casting, and tabulates the phys. and mech. properties of the alloy Berylco 20C (Be 20, Co 0·5%).—J. H. W.

The Removal of Non-Metallic Inclusions in the Manufacture of Bond-Cast Bearings, Particularly [Steel-Backed] Lead Bronze [Bearings]. A. Rühenbeck (*Metall*, 1951, 5, (3/4), 57–58).—The successful prodn. of steel-backed Pb bronze bearings depends on the prevention of oxide films, borax particles, and gas bubbles remaining at the interface. R. reviews some American, British, and German patents for this purpose.—E. N.

The Use of Electric Furnaces in the Casting of Copper Alloys. Georges Blanc (*Fonderie*, 1951, (65), 2467–2480).—B. briefly describes the particular appn., advantages, and limitations of the elect. furnace for melting Cu. The furnaces considered are: (1) resistance crucible, (2) reverberatory resistance, (3) arc, (4) L.F. induction, and (5) H.F. induction furnaces. In general, the advantages of these furnaces are flexibility, possibilities of high-temp., elimination of harmful constituents acquired from furnace gases, use of any given atmosphere or a vacuum, and avoidance of the inconveniences of a crucible. On the other hand, the initial cost is high, the running costs are greater than those of gas furnaces, and skilled workers are required. 6 ref.—J. H. W.

Melting Tin Bronze in Indirect-Arc Electric Furnaces. Bruce W. Schafer (*Foundry*, 1951, 79, (5), 132–135).—S. discusses gas pick-up in the melting of Sn bronzes in indirect-arc furnaces and suggests means of minimizing it. —R. W. R.

Melting Practices for Magnesium Alloys. K. S. Sealander (*Foundry*, 1951, 79, (5), 94–95, 204–207).—S. gives a short account of modern melting practice for Mg-base alloys. Be addn. of 0·0002% and N degassing are recommended for all alloys; the N degassing operation is carried out before superheating, but if necessary, an additional Cl degassing treatment may be given after refining. Grain refinement by C inoculation is also described.—R. W. R.

***Zinc-Base Alloys for Die-Casting. I.—Effect of the Quality of Zinc on the Decay Phenomena of Zamak.** (Hikage). See col. 183.

***The Nucleation of Cast Metals at the Mould Face.** (Reynolds and Tottle). See col. 191.

***Contribution to the Measurement of the Casting Properties of Metals and Alloys.** Raymond Guillemot (*Fonderie*, 1951, (65), 2459–2466).—Amongst other important factors in casting is the rate of filling of the mould. In the flowability of a metal or alloy, very small variations of temp. (down to 2·5° C.) are of importance, and it is difficult in the foundry to measure such small differences. G., by using a crucible divided into two compartments, measured accurately the relative temp. of a standard metal and the alloy under investigation poured simultaneously into two concentric spirals. The ratio of the length of the alloy spiral to that of the standard metal spiral was taken as a measure of the relative flowability of the alloy. The standards used were: (1) sublimed Mg for ultra-light alloys, (2) 99·99% Al for light alloys, (3) electrolytic Cu for Cu-base alloys, and (4) Armco Fe for ordinary and special steels. Practical tests with A-S4G (Cu 0·10, Mn 0·60, Mg 0·55, Fe 0·52, and Si 4·26%) against Al at 720° C. gave values of 88·4 and 83% for the flowability. G. points out that Kondic and Kozlowski (*J. Inst. Metals*, 1948–49, **75**, 665; *Met. Abs.*, 1949–50, **17**, 852) controlled the temp. in their experiments to ±1° C., and studied only relative low-m.p. alloys (up to m.p. of Al–Si), but that his method is, if less accurately controlled, better adapted for Cu and Fe alloys. Further, that for the highest accuracy it is necessary to assume that, since flowability is a linear function of the temp., the ratio of relative flowabilities should be const. over a range of temp. Owing to the existence of a temp. range of solidification in many alloys this is not actually the case. This method is especially useful for measuring the effect of a flux on an alloy and for comparing the flowability after one and two meltings. 22 ref.—J. H. W.

***Choke Control in Finger Gating.** W. H. Johnson, H. F. Bishop, and W. S. Pellini (*Foundry*, 1951, **79**, (4), 116–121, 275–278).—The paper describes the further results obtained in the course of the high-speed cinematographic study of the flow of metal in sand moulds, being carried out at the U.S. Naval Research Laboratory (J., Baker, and P., *Amer. Foundryman*, 1950, **17**, (4), 106; *Met. Abs.*, 1949–50, **17**, 852). The present work was confined to a study of the metal flow in multiple-finger gates, in which attempts were made to secure uniform flow by the use of chokes in each finger. It was found that if the finger gates are slightly choked, the running system becomes pressurized and the flow from each finger is uniform. Too heavy choking of the fingers results in undesirable jet effects at the gates. A choke placed between downgates and runner bar merely limits the rate at which the casting is run and does not ensure uniform flow through each gate. Fingers providing high frictional resistance to flow require less choking than those offering little resistance. The use of flared fingers confers little benefit in multiple-gating systems.—R. W. R.

Running and Feeding of Castings. H. S. Farmer (*Found. Trade J.*, 1951, **90**, (1805), 357–361; discussion, 361–362).—Paper read before the Birmingham branch of the Institute of British Foundrymen. A survey of the various techniques employed for the running and feeding of castings.
—J. E. G.

Precision-Investment Non-Ferrous Casting Alloys. —— (*Materials and Methods*, 1950, **32**, (5), 97, 99).—Materials Engineering File Facts No. 201. Compn., phys. and mech. properties are summarized for Cu, Co, Al, and Ni alloys suitable for making precision-investment castings.
—J. W. C.

Precision-Investment Castings Replace Parts Produced by Other Methods. Rawson L. Wood and Davidlee Von Ludwig (*Materials and Methods*, 1950, **32**, (3), 49–53).—Examples of economy and improved performance obtained by the substitution of investment castings for small assemblies previously produced by other methods, are given.—J. W. C.

Metal Parts with High Accuracy and Finish Produced by New Casting Process. B. N. Ames, S. B. Donner, and N. A. Kahn (*Materials and Methods*, 1950, **32**, (2), 43–46).—Cf. *Met. Abs.*, 1950–51, **18**, 679. A description of the Croning or "C" moulding process developed in Germany, which utilizes moulds made from silica sand and powdered phenol-formaldehyde resins. Tests were made to determine the casting properties of Al alloys, bronze, grey Fe, and medium-C and stainless steels in these moulds. The authors conclude that the "C" casting process is very suitable for long-run and repetition work and that at present it is most suited to non-ferrous alloys and cast Fe.—J. W. C.

Intricate Castings from a Durable Loam Mould. J. Currie (*Found. Trade J.*, 1951, **90**, (1806), 383–388; (1807), 419–425).—Paper read before the Scottish branch of the Institute of British Foundrymen. An illustrated account, giving details of the moulding technique employed.—J. E. G.

***Metal Penetration [into Moulding Sand].** S. L. Gertsman (*Foundry*, 1951, **79**, (5), 84–89).—Cf. *Met. Abs.*, this vol., col. 154. The results are described of tests in which the influence of several factors on the penetration of moulding sand by steel was investigated; among these factors were the ferrostatic pressure, pouring temp., the presence of gas in the metal, the grain-size of the sand, the type of binder, the hardness of ramming, and the type of mould wash. It was concluded that sand penetration is minimized by the use of low pouring temp., combined with fine-grained sand (or sand contg. wood flour) and a mould wash which does not crack when heated.—R. W. R.

What's Wrong with Castings? J. B. Caine (*Foundry*, 1951, **79**, (6), 106–108, 267–271).—C. points out that rounded defects in castings are relatively innocuous compared with sharp cracks which produce strong concentrations of stress, and suggests that foundrymen should therefore devote most attention to the avoidance of such defects as cracks and hot tears.—R. W. R.

The Cup, Saucer, and Spoon [Casting Test]. Pat Dwyer (*Foundry*, 1951, **79**, (5), 106–109).—D. describes the essential steps in the one-piece prodn. of cup, saucer, and spoon castings frequently used as tests of the abilities of foundry apprentices.—R. W. R.

Design of Core Boxes and Driers for Use on Core Blowers. Elmer Blake (*Foundry*, 1951, **79**, (6), 114–117, 260–264).—B. describes and discusses the design and manufacture of the boxes and driers used to produce a typical core.—R. W. R.

Olivine Sand Research. Gilbert S. Schaller and W. A. Snyder (*Foundry*, 1951, **79**, (5), 104–105, 168, 170).—A brief account is given of work in progress at the University of Washington on the properties of moulding sands made from olivine sands.—R. W. R.

Synthetic Resins in the Foundry. P. G. Pentz (*Metal Ind.*, 1951, **78**, (25), 504–507).—Abstract of a paper presented to the British Plastics Convention, discussing the properties necessary in a resin-bonded core sand.—J. H. W.

Naturally-Bonded versus Synthetic Sands. J. J. Sheehan and W. B. Parkes (*Proc. Inst. Brit. Found.*, 1950, **43**, B46–B48).—See *Met. Abs.*, 1949–50, **17**, 384.

Experimental Work on Oil-Sand Practice. D. T. Kershaw (*Found. Trade J.*, 1951, **90**, (1796), 115–122).—Read before the Slough section of the London branch of the Institute of British Foundrymen. An account of the appn. of sand-testing techniques to the elimination of casting defects usually associated with faulty oil sand.—J. E. G.

Fumes from Oil-Bonded Cores. —— (*Found. Trade J.*, 1950, **89**, (1789), 499–500).—A summary of the Tech. Sub-Committee report submitted to the Joint Advisory Committee on Conditions in Iron Foundries.—J. E. G.

New Sand-Testing Apparatus. W. H. Moore (*Found. Trade J.*, 1951, **90**, (1793), 46–49).—A description of a high-temp. sand-testing apparatus and its employment for the elimination of buckles, rat tails, veining, and allied casting defects.—J. E. G.

Rationalization of Sand Preparation. J. F. Goffart (*Proc. Inst. Brit. Found.*, 1950, **43**, A189–A197; discussion, A197–A201; and *Found. Trade J.*, 1950, 89, (1779/80), 319–326; discussion, (1787), 448–451).—Official Exchange Paper from Association Technique de Fonderie de Belgique to the Institute of British Foundrymen. A detailed and critical survey of the process of sand prepn., reclamation, and regeneration.—J. E. G.

Pattern-Making for General Engineering Casting. H. S. W. Brittain (*Proc. Inst. Brit. Found.*, 1950, **43**, B81–B96; discussion, B96–B98).—See *Met. Abs.*, 1949–50, **17**, 855.

Pattern Equipment for Production Moulding. S. A. Horton (*Proc. Inst. Brit. Found.*, 1950, **43**, B70–B79; discussion, B79–B80).—See *Met. Abs.*, 1949–50, **17**, 855.

Pattern Jigging on Moulding Machines. H. Haynes (*Proc. Inst. Brit. Found.*, 1950, **43**, B49–B54).—See *Met. Abs.*, 1949–50, **17**, 855.

Sprayed-Metal Core-Boxes. F. Schumacher (*Found. Trade J.*, 1950, **89**, (1788), 472).—A brief account of prodn. and advantages of sprayed-metal core-boxes.—J. E. G.

Castings for Machine Tools. A. G. Thomson (*Found. Trade J.*, 1951, **90**, (1795), 101–104).—A general illustrated description of the works of Technaloy, Ltd.—J. E. G.

G. and J. Weir's Foundry and Heat-Treatment Facilities. —— (*Found. Trade J.*, 1951, **90**, (1804), 333–334, 343).—A description of a foundry manufacturing auxiliary machinery for marine and land engineering plant.—J. E. G.

A Visit to the Fonderie Bruneau Frères at Orléans. Maurice Victor (*Rev. Aluminium*, 1949, (161), 413–418).—An illustrated description of a French foundry engaged in the prodn. of high-duty castings for aircraft.—N. B. V.

Technical Control Emphasized at a Rochester Foundry. —— (*Found. Trade J.*, 1951, **90**, (1804), 335–340).—An illustrated account of the foundry of Kent Alloys, Ltd., making a wide variety of high-precision castings.—J. E. G.

Wetting Agents Useful in the Foundry. Paul Winters (*Foundry*, 1951, **79**, (6), 118).—W. describes ways in which dust may be reduced in foundries by use of water contg. wetting agents.—R. W. R.

Foundry Trades in the Great Exhibition, 1851. —— (*Found. Trade J.*, 1951, **90**, (1808), 447–449).—A general account.—J. E. G.

Fundamental Production Planning in the Foundry. P. Rohrhirsch (*Neue Giesserei*, 1950, **37**, (26), 588–592; and (abridged) *Found. Trade J.*, 1951, **90**, (1803), 315–318).—A general illustrated account.—J. E. G.

Foundry Conference at Leipzig, 14–15 Dec. 1950. —— (*Metall*, 1951, **5**, (3/4), 69–71).—Brief notes on the papers—dealing with both ferrous and non-ferrous metals and alloys—presented at the meeting.—E. N.

Work in a Jobbing Foundry. J. F. Dowell (*Proc. Inst. Brit. Found.*, 1950, **43**, B134–B146).—See *Met. Abs.*, 1950–51, **18**, 556.

Jobbing Foundry on Scientific Lines. —— (*Found. Trade J.*, 1950, **89**, (1785), 383–386, 398).—An illustrated account of typical practice in a jobbing foundry.—J. E. G.

What is Ahead for Non-Ferrous Foundries in Defence Work? W. A. Mader (*Amer. Foundryman*, 1951, **19**, (4), 105–106).—Met. and foundry trends are reviewed.—V. K.

Role of a Metallurgical Engineer. Peter E. Kyle (*Amer. Foundryman*, 1951, **19**, (4), 100–103).—The syllabus of the metallurgy course at Cornell University is described, with particular emphasis on the foundry aspects of the course. —V. K.

Industrial Engineer in the Foundry. Douglas C. Williams (*Amer. Foundryman*, 1951, **19**, (5), 88–90).—The syllabus of the foundry course at Ohio University is described.—V. K.

The Foundry, the Engineer, and the Future. J. G. Pearce (*Found. Trade J.*, 1950, **89**, (1781/82), 343–346; discussion, 346–348, 350).—Read before the Lancashire branch of the Institute of British Foundrymen. A critical review of the interrelationship of design, chem. compn., microstructure, and mech. properties of castings. P. attaches considerable importance to functional qualities.—J. E. G.

[Foundry] Foremanship. S. Lewis (*Proc. Inst. Brit. Found.*, 1950, **43**, B147; discussion, B148).—See *Met. Abs.*, 1949–50, **17**, 974.

Craftsmanship in the Foundry. C. Gresty (*Found. Trade J.*, 1951, **90**, (1815), 625–626).—Presidential address delivered at 1951 Annual Conference of the Institute of British Foundrymen.—J. E. G.

Time Study in Foundries. H. S. Ward (*Found. Trade J.*, 1950, **89**, (1788), 473–478).—An account of the appn. of time study to rate-fixing.—J. E. G.

Foundry Cost Estimation. —— (*Found. Trade J.*, 1951, **90**, (1812), 555–557).—General discussion at a meeting of the South African branch of the Institute of British Foundrymen. —J. E. G.

Safety and Hygiene in the Foundry. J. Steele (*Found. Trade J.*, 1951, **90**, (1793), 41–45).—Read before the South African branch of the Institute of British Foundrymen. A general account.—J. E. G.

Safety in the Foundry. H. Allen (*Found. Trade J.*, 1951, **90**, (1798), 181–182).—A. discusses amongst other topics the hazards arising from rough edges of moulding boxes, from the handling of chem. substances such as Na and degassing mixtures, from moving machinery, from using drying stoves without an emergency escape hatch, and from hot metal. —J. E. G.

Safety in a Foundry. I.—General Principles of Accident Prevention. II.—Application of These Principles to a Foundry. F. S. Harper and R. E. Hurnall (*Australasian Eng.*, **1951**, (Jan.), 92–95; (Feb.), 44–47).—[I.—] General principles of safety and accident prevention are described, and their appn. to sp. problems in the foundry discussed. [II.—] Foundry workers are injured because of faulty lifting equipment, unsafe working areas, faulty storage, or the manual handling of heavy materials instead of using mech. means. Eye injuries, contusions, lacerations, and burns are common. Corrective measures are outlined.—T. A. H.

Foundry Conditions in Great Britain. J. W. Gardom (*Found. Trade J.*, 1951, **90**, (1797), 153–156; (1798), 183–186; discussion, (1804), 341–343).—Read before the South African branch of the Institute of British Foundrymen. An account giving details of training, working conditions, developments in foundry equipment, met. control, sand practice, mechanization, melting practice, and special alloys. 31 ref.—J. E. G.

Review of the South African Foundry Industry. H. G. Goyns (*Proc. Inst. Brit. Found.*, 1950, **43**, A174–A183; discussion, A183–A188; and *Found. Trade J.*, 1950, **89**, (1781/82), 333–341; discussion, (1783/84), 365–370).—Presented by the South African branch of the Institute of British Foundrymen at the Institute's Annual Conference. An account of the extent and scope of the South African foundry industry is given.—J. E. G.

Foundry Management Today. J. Stott (*Found. Trade J.*, 1951, **90**, (1808) 435–443; discussion, 443–445).—A general review.—J. E. G.

Aids to Production [in the Foundry]. K. Docksey (*Found. Trade J.*, 1950, **89**, (1790), 537).—Presidential address to the East Midlands branch of the Institute of British Foundrymen.—J. E. G.

The Scientific Approach. H. T. Angus (*Found. Trade J.*, 1950, **89**, (1798), 558, 562).—Presential address to the Birmingham branch of the Institute of British Foundrymen.—J. E. G.

A Way of Living. John J. Sheehan (*Proc. Inst. Brit. Found.*, 1950, **43**, A22–A24).—See *Met. Abs.*, 1949–50, **17**, 974.

Foundry Practice. (Salmon and Simons). See col. 254.

Centrifugal Casting by the Cire Perdue Process. (Rosen). See col. 255.

16 — SECONDARY METALS: SCRAP, RESIDUES, &c.

***On the Action of Fluorides on an Aluminium–[5·5%] Magnesium Alloy.** Wilhelm Dautzenberg (*Z. Erzberg. u. Metallhüttenwesen*, 1951, **4**, (1), 15–19).—After considering the results of Schneider and Berblinger (*Z. Metallkunde*, **1949**, **40**, 411; *Met. Abs.*, 1950–51, **18**, 52) appertaining to the removal of Mg from Al alloys by treatment with various alkali and alkaline-earth chlorides, D. presents a report on similar work, carried out on an Al–5·5% Mg alloy, in which such chlorides were replaced by their corresponding fluorides. The results show that the efficiency of such fluorides in the removal of Mg increases in the order: (NaCl) $\rightarrow CaF_2 \rightarrow AlF_3.NaF \rightarrow NaF$—all of which are available commercially—and that the efficiency of each increases with increase in temp. between 850° and 1050° C.—E. N.

Reclaiming and Refining Light Metal Scrap. —— (*Modern Metals*, 1951, **7**, (2), 54–55).—Taken from "Fabrication of Aluminium in Germany" by the Technical Industrial Intelligence Division of the U.S. Department of Commerce, this article discusses German methods of recovering Al from scrap aircraft. The use of "dry-hearth" furnaces, methods of removing Mg, and the Beck dual refinery process for Al and Mg are briefly reviewed. In the latter process vacuum C-rod furnaces were used, and an important part of the process is the degassing technique.—R. J.

Direct Extrusion Applied to Light Metal Scrap. Max Stern (*Iron Age*, 1951, **167**, (26), 71–73).—Dispersion of oxides, destruction of chip layers, and almost normal properties were obtained in bars of Al and Mg scrap extruded at 750° F. (400° C.) and 15 tons/in.^{2}.—J. H. W.

***[Recovery and] Separation of Copper from Zinc [in Dilute Solutions and Waste Liquors] by Ion-Exchange [Resins].** Ernest J. Breton, Jr., and A. W. Schlechten (*J. Metals*, 1951, **3**, (7), 517–521).—The concentration and selective sepn. of Cu and Zn in dil. soln. by means of ion-exchange resins of the carboxylic and the sulphonic types has been studied as a function of flow rate, *p*H, Cu : Zn ratio, and concentration. The results show that: (1) the sulphonic resin has a high concentration capacity under wide variations of *p*H and flow rates, although its selectivity is low; it would be useful for the removal of ions under acidic conditions; (2) the carboxylic resin has a very high concentration capacity for the recovery of Cu above $pH = 7$; below $pH = 5$ the capacity becomes too low for most purposes; the resin would be useful where a sepn. is desired; and (3) ion-exchange resins, especially those of the carboxylic type, can be used to concentrate ions from dil. soln. It is concluded that the ion-exchange method can extend the range of metal recovery far beyond that obtainable by conventional methods; it is particularly suitable for the recovery of metal ions from waste soln., with consequent recovery of metal values, prevention of stream pollution, and, possibly, reclamation of the water. 6 ref.—E. N.

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17 — FURNACES, FUELS, AND REFRACTORIES

Induction Furnace [for Galvanizing] Cuts Zinc Loss. Edward W. Burd (*Steel*, 1950, **127**, (1), 88–90, 92).—The use and advantages are described of L.F. induction heating of the galvanizing furnace in the prodn. of Zn-coated steel sheets.—E. J.

Modern Heat-Treatment Furnaces. L. G. A. Leonard (*Found. Trade J.*, 1951, **90**, (1795), 89–94; and also *Indust. Gas*, 1951, **14**, (160), 105–108, 110, 112, 114, 116).—Paper presented to the Manchester Association of Engineers. An account of developments in furnace construction with descriptions of typical examples of modern furnace types. —J. E. G.

Improvements in Electric-Furnace Design. F. V. Lewis (*Found. Trade J.*, 1950, **89**, (1789), 505–506).—A description of the appn. of rotary controllers fitted with "Amplidyne" to a 3-phase, direct-arc melting furnace. A new type of electrode gland designed to reduce electrode consumption is also described.—J. E. G.

Improvements in Furnaces and Methods for Heat-Treatment. Floyd E. Harris (*Metal Progress*, 1951, **59**, (1), 63–68).—A general review is given, mainly devoted to furnaces for ferrous materials.—F. A. F.

Solar Furnace for Ferrous Metal Melting Experiments. F. Trombe and M. Foëx (*Rev. Mét.*, 1951, **48**, (5), 353–358; discussion, 358).—A 2-kW. solar furnace is described in detail. —T. G.

Pure Refractory Materials. F. H. Norton (*U.S. Atomic Energy Commission Publ.*, 1948, (**AECD-2237**), 3 pp.).—A list of the thermal properties of stable sulphides of various elements including those of Ce, Th, and U. Of these CeS is the most promising as a crucible material, and a flow sheet is included showing its prodn. from either the oxide or the metal. Crucibles may be formed by pressing in steel moulds. CeS has a metallic Au colour, is stable in air, has a *d* of 5·1–5·8 g./c.c., and melts at >2200° C. It is an excellent crucible material for vacuum melting of metals where freedom from O_2 is essential. ThS may be produced by a similar process and is silvery with a *d* of 9·6 g./c.c.—B. W. M.

Some Aspects of the Use of Refractory Materials. F. H. Clews and A. T. Green (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (II), 933–961; discussion, 1047–1093).—See *Met. Abs.*, 1949–50, **17**, 298.

Refractories in the Foundry. A. Gunn (*Proc. Inst. Brit. Found.*, 1950, **43**, B55–B61; discussion, B61–B62).—See *Met. Abs.*, 1949–50, **17**, 763.

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18 — HEAT-TREATMENT

The Heat-Treatment of Aluminium Alloys. J. Crowther (*Metallurgia*, 1951, **43**, (259), 243–247).—The heat-treatment of Duralumin alloys B, S, H, and K is discussed in detail. —F. M. L.

Electric Annealing Furnace Assures Uniform Grain-Size of Seamless Copper Tubing. —— (*Indust. Heating*, 1951, **18**, (2), 230, 232, 234).—An illustrated account of an elect. roller-hearth furnace for the annealing of seamless Cu tubing. Grain-size control is possible to ±0·5 mm.^{2} Two controlled-atmosphere generators supply the 73-ft.-long furnace, which is capable of handling 3000 lb./hr. of Cu tubing. —D. M. L.

Large Cast-Brass Cakes Heated in Continuous Cycle in American Brass Plant. —— (*Indust. Heating*, 1951, **18**, (5), 806, 808, 810).—An illustrated description of a large oil-gas furnace for the continuous heating of $1\frac{1}{2}$-ton brass slabs for hot rolling. The furnace capacity is nearly 20 tons/hr. in a continuous cycle. Two zone temp. are maintained within the furnace chamber.—D. M. L.

Annealing of Ti and Zr. A. M. Bounds and H. W. Cooper (*Metal Progress*, 1951, **59**, (1), 69, 100, 102).—A discussion is given of the problem of furnace atmosphere for annealing these metals: vacuum annealing is recommended. Molten caustic (NaOH) is useful as a pickling medium, but its use may be dangerous, leading to violent reaction between the caustic bath and Ti.—F. A. F.

Recent Heat-Treatment Installations: Exacting Demands of Modern Engineering Govern Design. —— (*Metallurgia*, 1951, **43**, (259), 221–233).—Recent installations incorporating latest developments are briefly described.—F. M. L.

19 — WORKING

The Stretch-Wrap Forming of Aluminium Sheets. Henri Leboutеux (*Rev. Aluminium,* 1950, (**164**), 96–99).—An illustrated account of the process used by the Hufford Machine Works, U.S.A. Cf. Kluge, *Light Metal Age,* 1949, **7**, (7/8), 10; *Met. Abs.,* 1949–50, **17**, 859.—N. B. V.

Deep Drawing and Stamping of Airplane Parts. J. J. Sloan (*Machine moderne,* 1950, **44**, (Feb.), 9–14; *Appl. Mechanics Rev.,* 1951, **4**, 104).—After some general remarks on deep drawing, a number of examples are given of aeroplane parts produced by deep-drawing and stamping processes. Limits of height-to-width ratios of finished product and the min. radii of curvature related to sheet thickness are given for steel and Al.

The Fabrication and Erection of the Arvida Aluminium Bridge. C. J. Pimenoff (*Eng. J.,* 1950, **33**, (6), 446–452).—Cf. *ibid.,* 1949, **32**, 186; *Met. Abs.,* this vol., col. 239. P. describes the procedures used in fabricating and assembling the bridge in the shop and in erecting it on the site. Assembly was carried out by riveting; see Anders and Elliot, *ibid.,* 1950, **33**, 453; *Met. Abs.,* this vol., col. 235.—N. B. V.

Extruding Copper and Brass. W. W. Cotter, Sr. (*J. Metals,* 1951, **3**, (6), 440–442).—After briefly reviewing the history of the extrusion process, C. describes the construction, operation, and performance of Loewy standard single and double-actioned hydraulic extrusion presses of the universal vertical and horizontal types.—E. N.

[Making] Cartridge Cases from Round Stock. Arthur F. MacConochie (*Steel,* 1950, **127**, (11), 88–91; (12), 108–110).—Discusses the Manuchin process for making small-arms cartridge cases in brass from extruded bar stock instead of from strip. Details of the machines used for slug making, cupping and drawing are given, as well as the advantages of the process.—E. J.

Compression-Formed Steel [and Non-Ferrous] Tubing Offers Close Tolerances and High-Strength Properties. L. A. Karg (*Materials and Methods,* 1950, **32**, (2), 52–53).—The Rockrite process for making compression-formed tubing is described. Compression-formed tubing is produced to close tolerances, is fine-grained, has uniform microstructure, and a shallower depth of decarburization than cold-drawn tubing. The process is suitable for Cu, Ni, Be, and Al alloys, as well as steels.—J. W. C.

Magnesium Sheet Hot Formed at Less Cost. Gilbert C. Close (*Steel,* 1950, **127**, (22), 56–58).—Experiments are described on the successful hot forming of Mg alloy aircraft components from sheet.—E. J.

***Can You Friction-Saw Nickel-Base Alloys?** H. J. Chamberland (*Steel,* 1950, **127**, (10), 91–92).—Results are given on experiments using a band-saw on various proprietary Ni alloys.—E. J.

Sheet and Tinplate Mills: Measurement of Roll Camber. J. H. Mort (*Iron and Steel,* 1948, **21**, (4), 111–115; (5), 147–149).—N. B. V.

How to Work Titanium and Its Alloys. Anderson Ashburn (*Machinist* (*Eur. Edn.*), 1951, **95**, (26), 961–969).—A. summarizes the at present limited information available on the working characteristics and properties of Ti and its alloys. He suggests that this report can serve as a guide to those with little experience of Ti, but emphasizes the fact that few operations have been performed sufficiently often for it to be certain that they are the best methods possible.
—J. H. W.

Rearmament and the Aviation Gas Turbine. [Methods of Making Parts for Gas Turbines]. F. R. Banks (*Aeroplane,* 1951, **80**, (2073), 438–440).—Of present prodn. methods B. favours the stamping of compressor and turbine blades to finished size and using the sintered-Fe process for compressor stator blades. In the case of high-alloy turbine blades, a new die may only take 1000 pieces of work before being resunk or replaced. Hence a large set-up of hammer and stamps backed by a battery of die-sinking machines are necessary. High-alloy nozzle guide vanes are precision cast by the lost wax process; they can, however, also be made from alloy sheet formed to proper profile and welded at the trailing edge. Compressor blades can be made by rolling strip steel to finished aerofoil form and then chopping it off to suitable lengths and brazing on the root.—H. Pl.

Annual General Meeting and Proceedings of the Institute for Research on Sheet-Metal Working [November 1950 at Darmstadt]. —— (*Metall,* 1951, **5**, (1/2), 22–23).—A brief report on the papers presented at the meeting: *O. May* described a new type of press which he had designed; *O. Kienzle* discussed "Some Fundamental Principles Associated with the Forming of Sheets"; *O. Klein* spoke on "The Qualities of Sheet" in relation to various forming operations; *H. Beisswänger* dealt with "Flow Patterns Resulting from Deep Drawing"; *A. Burkhardt* described "A New Deep Drawing Process"; *H. Bablik* discussed "Wet or Dry Galvanizing?"; *A. Kutzelnigg* gave a "Survey of the Fields of Application of Electroplating"—for the protection of metal surfaces; *H. Biel* dealt with "Sprayed Metals as a Means of Corrosion Protection"; *F. Bollenrath* reviewed the development of "Electrostatic Methods for Spraying Paints" in the U.S.A.; *R. von Linde* discussed "The Use of Controlled Atmospheres in the Forming of Sheets".—E. N.

Cold Working of Metals. (——). See col. 255.

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20 — CLEANING AND FINISHING

A Method for Chemical Brightening of Aluminium. C. E. Naylor (*Plating Notes,* 1951, **3**, (2), 59–61).—The brightening soln. contains H_3PO_4 (sp. gr. = 1·59) 3 and H_2SO_4 (sp. gr. = 1·84) 1 part by vol. The article is immersed in this soln. at a temp. of 100°–120° C. for 2 min., rinsed for 5 sec., then immersed in a soln. contg. H_2SO_4 (sp. gr. = 1·84) 3% by vol., and H_2CrO_4 112 g./l. at 60°–80° C. to prevent etching of the Al.—T. A. H.

Modern Developments in Metal Finishing: Methods for the Protection and Decoration of Aluminium.—I.–II. V. F. Henley (*Brit. Indust. Finishing,* 1950, **2**, (25), 868–870, 872, 875, 876; **3**, (26), 33–34, 36–40, 42).—[I.—] Methods for the mech. polishing, cleaning, pickling, and bright dipping of Al are reviewed, with particular ref. to modern developments. [II.—] Electrobrightening, anodizing and dyeing, chem. pretreatment, and etch primers are discussed.
—H. A. H.

Automatic Spray-Finishing of Aluminium Siding. Richard H. Nappi (*Products Finishing,* 1951, **15**, (9), 44–46).—Al sheets used for house "siding" are first painted with a Zn chromate primer and then with a baking urea-formaldehyde synthetic lacquer.—G. T. C.

Aircraft Finish Demonstrates Durability. Gilbert C. Close (*Products Finishing,* 1951, **15**, (9), 36).—A brief description is given of the condition of an Alclad wing-flap which had been immersed in the sea for five years. The material had received a CrO_3 treatment, a coat of Zn chromate primer, and a top coat of lacquer. In no place had the finish failed sufficiently to permit the corrosion of the underlying metal.
—G. T. C.

***The Problem of Painting Bright Chromium Plate.** Wardley D. McMaster (*Plating,* 1951, **38**, (7), 696–698, 703).—Various precautions which have been taken in the past in an endeavour to ensure good adhesion of paint on Cr plate are briefly described, and it is shown that in all cases a certain number of failures was invariably encountered. However, the use of a hot CrO_3 rinse contg. 0·11 g./l. CrO_3 before painting ensured good adhesion. The work is allowed to air-dry

without further rinsing. Full details are included on experiments carried out with other concentrations of CrO_3. The use of H_3PO_4 in place of CrO_3 was not satisfactory, nor was a thorough degreasing treatment. Apart from the use of the CrO_3 pretreatment, the following conditions are suggested as important in ensuring good adhesion of organic coatings: (1) careful control of the plating bath, (2) adequate cleaning in the case of buffed parts, (3) the use of a paint or enamel which adheres well to steel, (4) an adequate baking, and (5) the use of a primer for those enamels which contain a significant proportion of Al flake pigment. It is shown that if failure does not occur when the painted specimen is exposed for 24 hr. to an atmosphere of 100% relative humidity at 38° C. or to a water soak test of 2–4 hr. at 38° C., it is unlikely that failure will occur in service. No difference in paint performance could be detected between specimens in which the Cr plate had been applied over steel, brass, or Zn.
—G. T. C.

Surface Finishing the Copper Alloys. [—I.–II.]. W. F. Walker (*Mech. World*, 1951, 129, (3358), 501–503; (3359), 523–526).—[I.—] Recommendations are made for the fettling, polishing, descaling, and cleaning of castings in Cu-base alloys. [II.—] W. describes various methods of surface finishing Cu alloys, including scratch-brushing, burnishing, chem. etching, chem. colouring, and lacquering.
—R. W. R.

Finishing Magnesium. I.—Cleaning, Pre-Treatment, and Chemical-Finishing Methods. —— (*Product Finishing* (*Lond.*), 1951, 4, (7), 40–51, 92).—A review of acid and alkali cleaning procedures, mech. surface treatment methods, and chem. and electrochem. produced conversion coatings for Mg. The various types of treatment included in the latter groups include: chrome pickle, sealed chrome pickle, bright chrome pickle, R.A.E. 30-min. bath, alkaline dichromate, modified alkaline dichromate, chrome-sulphate, Melanizing, chrome-alum, chrome-phosphate, dichromate, caustic pressure, selenious acid, galvanic anodizing, caustic anodizing, dichromate-phosphate anodizing, and the Manodyz process.
—H. A. H.

[Steel] Sheet and Tinplate Manufacture: The Pickling Process. J. H. Mort (*Iron and Steel*, 1948, 21, (10), 403–407; (11), 437–441; (12), 472–474).—N. B. V.

Application of Organic Coatings to Metal.—I. B. M. Letsky (*Paint Manuf.*, 1951, 21, (3), 83–89, 109).—Brief information is included on the roller-coating method of applying paints, &c., to tinplate.—G. T. C.

Cleaning Castings Chemically. R. J. Peters (*Found. Trade J.*, 1951, 90, (1811), 529–530).—A description of a method of cleaning castings by immersion in a catalysed molten salt bath contg. NaOH. The castings are subsequently immersed in cold and hot water rinses.—J. E. G.

Soft Grits Provide Low-Cost Method for Blast-Cleaning Metals. T. F. Clark and E. C. Lathrop (*Materials and Methods*, 1950, 31, (5), 67–69).—C. and L. discuss the use of grits made from agricultural residues, such as corncobs, rice, and nutshells, for blasting metal surfaces in order to degrease, remove weld scale, and clean before electroplating.
—E. C. P.

Blast Cleaning: A Recently Developed Method of Using the Cleaning Power of Abrasive Blast without the Limitations of an Enclosed Booth. —— (*Product Finishing* (*Lond.*), 1951, 4, (4), 40–43).—A new dustless, blast cleaning method for the removal of paint, scale, rust, corrosion products, &c., from metal surfaces is described.—H. A. H.

***Cleaning and Preparation of Metals for Electroplating. III.—Degreasing Evaluation Tests: The Atomizer Test.** Henry B. Linford and Edw. B. Saubestre (*Plating*, 1951, 38, (7), 713–717).—Cf. *ibid.*, (4), 367; *Met. Abs.*, 1950–51, 18, 778. A report on American Electroplaters' Society Research Project No. 12. The atomizer test is a new method for determining the presence of grease on a metal surface. The test involves the following steps: (1) evaporation of all surface moisture, (2) spraying of water against the test panel, using an atomizer, (3) measuring the areas covered by a thin, uniform film of water. Full details are presented of the equipment and the procedure used. Panels oiled with either mineral oil or lard oil were used for the tests, and it was established that the former was removed more readily than the latter. In the case of the lard oil, the atomizer test was able to distinguish between various degrees of cleaning as shown by the effect of time and temp. on the cleaning index. For lard oil, the cleaning index increases approx. linearly with time, the higher the temp. the steeper the slope. It is clear that the atomizer test provides greater sensitivity than a visual inspection and that it shows up the presence of oil far more strikingly than the water-break test. The reproducibility of the atomizer test is not very good.
—G. T. C.

[Discussion on a Paper by J. C. Harris, R. E. Kamp, and W. H. Yanko on:] Detection of Soil Removal in Metal Cleaning by the Radioactive-Tracer Technique. —— (*J. Electrochem. Soc.*, 1951, 98, (6), 258–259).—Cf. *Met. Abs.*, 1950–51, 18, 442.—G. T. C.

Developments in Drying [and Stoving Lacquered] Sheet Metal. A. Merz (*Metalloberfläche*, 1951, [A], 5, (4), 57–60).—Modern German equipment for the semi-automatic and automatic drying and stoving of lacquered sheet is described and illustrated.—E. N.

Chemical Colouring (Patina) of Decorative and Everyday Metal Articles. H. Krause (*Metalloberfläche*, 1951, [B], 3, (1), 1–3; (2) 22–24).—Methods for the chem. colouring of Cu and its alloys (including German silver and Al bronze), Al and Mg and their alloys, Fe, Sn, Zn, and Pb, are discussed.
—E. N.

[Stoving] Japans and [Synthetic] Lacquers [Suitable for] Electrodeposited Metals. H. Weise (*Metalloberfläche*, 1951, [B], 3, (4), 50–51).—E. N.

Mechanical Finishing of Metals: For Decorative Purposes. John B. Campbell (*Materials and Methods*, 1950, 32, (5), 81–96).—Materials and Methods Manual No. 64. A comprehensive review of mech. methods of producing attractive surface finishes on various metals and their alloys.—J. W. C.

Finishes Successfully Baked on Non-Ferrous Products by Induction Heating. Frank Netschert (*Materials and Methods*, 1950, 32, (3), 60–61).—The baking of enamel on Al and brass products has been effected by induction heating. The product is superior to those from gas oven or infra-red heating, principally because the enamel in contact with the metal is baked first and gases are expelled as hardening proceeds outwards.—J. W. C.

Finishes for Torch Cases. H. L. Turner (*Product Finishing* (*Lond.*), 1951, 4, (6), 56–62).—A fairly detailed description is given of the finishing operations used by the Ever Ready Co. (Great Britain), Ltd. These include the bright Ni and Cr plating of brass torch bodies, &c.; Ni followed by Ag plating of brass reflectors; the electropolishing and anodizing of Al torch cases; and the Granodine or Alocrom treatment of steel and Al components., resp., as a pretreatment before painting.—H. A. H.

Die-Castings Polished for Plating by Fadgenizing and Gyro-Finishing at Jernstedt. Ezra A. Blount (*Products Finishing*, 1951, 15, (8), 48–49).—Fadgenizing consists of mounting the parts on a cylindrical jig and rotating this immersed in a slurry consisting mainly of flour silica. In Gyro-finishing the slurry contains a mixture of liq. buffing compound and Maizo meal. It is claimed that large numbers of parts can be economically treated by these methods which replace conventional conveyor-type polishing equipment.
—G. T. C.

Increased Belt Life and Cut with New Contact Wheel. W. K. Van Ormer (*Plating*, 1951, 38, (6), 554–555, 557).—A new type of rubber contact wheel for abrasive belt polishing is described. It is claimed that the use of this wheel enables the operator to double the output from a standard abrasive belt in those operations where glazing normally occurs, since it causes a controlled breakdown of the bond and hence reduces glazing. The wheel is a serrated one with wide-spaced, narrow bands with an optimum relief angle. Laboratory and works trials with this type of wheel are briefly described. It is claimed that such wheels show advantages in rough snagging and that in addn. to producing longer belt life, they give higher stock removal rates, a more uniform finish, that they reduce the number of belt changes and result in less operator fatigue.—G. T. C.

Metal-Finishing Research. —— (*Product Finishing* (*Lond.*), 1951, 4, (5), 85–87).—A summary of the sections of the 31st Annual Report of the British Non-Ferrous Metals Research Association which deal with metal finishing. —H. A. H.

Finishing Research in the U.S.A. —— (*Product Finishing* (*Lond.*), 1951, 4, (4), 44–52).—Notes on the current work of the Battelle Memorial Institute and the U.S. National Bureau of Standards on the pretreatment and finishing of metal surfaces.—H. A. H.

Marking Processes.—I.–X. E. E. Halls (*Product Finishing* (*Lond.*), 1950, 3, (4), 75–91; (5), 72–82, 84; (6), 79–82, 84–89; (9), 85–93; (10), 49–64; (12), 64–71, 79; 1951, 4, (2), 80–95; (4), 74–84; (6), 78–90, 106; (7), 82–92).—[I.—] *Engraving and Filling*: a survey of the engraving and filling of ferrous and non-ferrous metals and plastics. General procedures described include those for Cu, brasses, Al and its alloys (including anodized and dyed), Zn and Zn alloys, &c.; [II.—] *Branding, Coding, and Hot-Press Marking*; [III.—] *Transfer Marking Processes and Techniques*; [IV.—] *The Gelatine Transfer Marking Method*; [V.—] *Marking by Means of Stencilling Apparatus*; [VI.—] *Rubber Stamp Marking*; [VII.—] *Chemical Etching or Engraving*; [VIII.—] *Silk Screen Stencil Marking Methods*; [IX.—] *Adhesive Label Marking Methods*; [X.—] *Marking of Laminated Sheet and Sandwich Board*.—H. A. H.

21 — JOINING

Methods of Joining the Copper Alloys. W. F. Walker (*Mech. World*, 1951, **129**, (3357), 475–479).—W. briefly describes the soldering, brazing, and welding techniques used to join Cu, brass, gun-metal, P bronze, high-tensile brass, Al bronze, Si bronze, cupro-nickel, and nickel silver. —R. W. R.

Prepared Pastes for Brazing Material. L. G. Klinker (*Metal Progress*, 1951, **59**, (1), 71–73).—A discussion is given of the advantages of the use of a paste rather than solid metal when furnace brazing. Pastes termed "Cubond" 151, 153, and 155 are discussed; these contain the metal and wetting agent, but no fluxes for 153 and 155, although fluxes are present in 151. Cubond 155 contains 4% Fe in the metallic part.—F. A. F.

The Silver-Brazed Pipeline. John B. Ross (*Weld. J.*, 1950, 29, (8), 636–639).—Some uses of Ag-brazed pipe joints are described.—W. D. B.

***Recent Canadian Developments in the Cold Riveting of Aluminium.** Ernest Anders and D. G. Elliot (*Eng. J.*, 1950, 33, (6), 453–460; also (abridged) *Engineering*, 1950, **170**, (4411), 141–143).—A. and E. describe work carried out on the development of large Al alloy rivets (>½ in. dia.) for use in constructing the Arvida Bridge (see Pimenoff, *ibid.*, 1949, 32, 186; 1950, 33, 446; *Met. Abs.*, this vol., col. 231, 239). Tests carried out on rivets of Alcan 26S-T alloy, $\frac{51}{64}$ in. dia., showed that an annular type of rivet could be successfully driven cold by means of a pneumatic gun, using a suitable set. Rivets of this type were used for the bridge and promise to extend considerably the structural appn. of Al alloys. —N. B. V.

***Optimum Flash-Welding Conditions for Aluminium Alloys: The Importance of Upset Variables.** E. F. Nippes, W. F. Savage, J. J. McCarthy, and P. Patriarca (*Weld. J.*, 1950, 29, (10), 497S–511S).—Joint efficiencies of nearly 100% were attained in the flash welding of Al alloys 61S-T, 14S-T, 75S-T, and 24S-T. With the exception of the last alloy upset current magnitudes of 50,000 or 70,000 amp./in.^{2} for durations of 25 or 9 cycles, resp., are not detrimental. Greater upset currents result in peak temp. >600° C., thus causing over-ageing; 61S-T is the most, and 24S-T the least susceptible to this. The magnitude of the upset current was the most important variable influencing the upset pressure requirements; in general, slower flashing velocities and larger upset currents were more effective in reducing pressure. A reduction of 45–65% in pressure was possible without reducing joint efficiency by >10%. With 24S-T conditions could be selected such that a 50% reduction in upset pressure was achieved without decreasing the weld properties.—W. D. B.

Weld [Aluminium] Without Flash. Charles Bruno and G. W. Birdsall (*Steel*, 1950, 127, (2), 86–87, 102).—Describes a technique for the flash welding of Al components with the min. of flash.—E. J.

Welded [Aluminium Alloy] Oil Pipelines: A Review of Techniques and Progress. Rolt Hammond (*Petroleum*, 1951, **14**, (4), 87–90).—The use of an Al alloy contg. 0·7% Mg and 0·4% Si for constructing 1½ miles of 4-in.-dia. pipe-line is described, though the article is mainly devoted to steel pipes. The Al pipe was used in 40-ft. lengths and welded by the Argonarc process, which is also briefly described. The use of Al for this purpose is in the development stage, and it will be some time before the real value of Al to resist corrosion under these conditions can be assessed. —G. T. C.

Aluminium Boxes Inert-Arc Welded. —— (*Iron Age*, 1951, **167**, (18), 99).—Heliarc welding is used in the mass prodn. of rectangular Al boxes, whereby the corners of the boxes can be welded without burning through the material and warpage is virtually eliminated.—J. H. W.

Welding Aluminium with Inert-Arc D.C. John W. Mortimer (*Weld. J.*, 1950, 29, (10), 879–881; discussion, 1951, **30**, (1), 49–52).—Advantages claimed include low initial cost, absence of H.F. interference, and lower cost of He compared with A (in the U.S.A.).—W. D. B.

Shielded-Arc Welding of Aluminium. Fritz Albrecht (*Weld. J.*, 1950, 29, (7), 542–546).—A comparison between welding costs for some typical prodn.-quality Al assemblies welded by the O–H torch and by the A-shielded A.C. arc indicates that the latter is more economical, provided that continuous utilization is possible.—W. D. B.

Welded Aluminium Ammonium Nitrate Prilling Tower. H. N. Hockensmith (*Weld. J.*, 1950, 29, (9), 729–732).—Materials and procedures are described.—W. D. B.

Gas Welding with Bronze in Production. Clinton E. Swift (*Weld. J.*, 1950, 29, (8), 646–650).—Appn. of bronze welding to thin-gauge parts is described and illustrated.—W. D. B.

Welding of Zinc-Base Alloy Die-Castings. Roland H. Ogden (*Weld. J.*, 1950, 29, (6), 453–457; and also *Weld. Eng.*, 1950, **35**, (1), 38–41).—The compn. and properties of the commoner Zn-base die-casting alloys are discussed, and recommendations are made for filler rod, techniques, &c. The use of scrap-metal for filler rods is unsuitable.—W. D. B.

The Weldability of Non-Ferrous Metals and the Results Obtained by Oxy-Acetylene Welding. Ed. Henrion (*Rev. Soudure*, 1949, 5, (1), 16–18).—The weldability of pure Al and Cu and Al–Mg alloys is briefly discussed and typical test results are given. It is concluded that no difficulty need be experienced with the pure metals, and that mech. properties at least equivalent to those of the parent plate may be obtained in the Al–Mg alloys by using a filler rod contg. 7% Mg, and by lightly hammering after welding.—W. D. B.

Flash-Welding Non-Ferrous Metals. F. L. Brandt (*Weld. J.*, 1950, 29, (9), 718–722; discussion, (12), 1087–1088).—Recently developed techniques are described, and recommended machine settings are given in tabular form for a number of welds including Cu, Al, Ti, Ag, &c., and some of their alloys.—W. D. B.

***The Fatigue-Resistance of Welded Joints.** [—I.–II.]. W. Soete and R. van Crombrugge (*Rev. Soudure*, 1950, 6, (2), 72–82; (4), 199–212).—Repeated tension-fatigue tests have been carried out on a number of typical welded joints in mild steel. (1) *Unrestrained butt welds*: the fatigue limit (at 2×10^6 cycles) in both the "as-welded" and "welded and stress-relieved" conditions was 15–16 kg./mm.^{2} for unmachined joints; machining raised this figure by some 15%, machining and stress-relieving by 25%. Hence stress-relieving alone has no effect; the surface condition is of paramount importance. (2) *Restrained butt welds*: the results were contradictory, but with lightly restrained butt welds there is some evidence to suggest that residual stresses have a negligible effect on the resistance to fatigue; the question of severe restraint is much more complex, and will be investigated in a future research. (3) *Cruciform welds*: the resistance to fatigue of a joint welded with ordinary electrodes is either ~ 5 kg./mm.^{2} or up to 16 kg./mm.^{2}, according to the depth of penetration; the latter figure is obtained when penetration is complete using deep-penetration electrodes. The resistance to fatigue varies with differing depths of penetration, and while a broad relationship is indicated, further work is required.—W. D. B.

***Variables in Cross-Wire Welding of Dissimilar Metals.** I. S. Goodman (*Weld. J.*, 1950, **29**, (10), 863–875).—For Fe, Ni, or Fe–Ni welds the optimum strength varies with different setdown percentages (decrease in combined wire dia. after welding) depending on dia. or temper. Differences in welding time have little effect. W or Mo weld properties are variable, depending upon amount of recrystn. and stress. Welds of Fe to refractory metals are difficult; max. stress occurs close to 100% setdown, providing embrittlement is not excessive. —W. D. B.

The Order of Succession of Welding Passes and the Sequence of Welded Joints. W. Bonhomme (*Rev. Soudure*, 1949, **5**, (1), 31—39).—Examples are given of the welding sequences used for the vertical members and girders of a Vierendeel bridge, a Diesel-motor frame, a large flat-bottomed vessel, and a vessel with a conical bottom.—W. D. B.

A Simple Method for Butt Welding Fine Wires [for Thermocouples]. Maurice Spielman (*Rev. Sci. Instruments*, 1951, **22**, (3), 216–217).—A method using ordinary apparatus is given. —E. J.

†The Aircomatic Welding Process. A. Muller, G. J. Gibson, and E. H. Roper (*Weld. J.*, 1950, 29, (6), 458–482).—The fundamentals of metal-arc transfer in inert-gas-shielded arc welding are discussed, and the Aircomatic process is described. The characteristics of welds made by this process in Al alloys, and Al bronze, using both manual and automatic methods, are given. All the normal types of weld can be produced. Gas porosity can be eliminated in flat Al alloy welds, giving mech. properties which are the max. that can be expected. Al bronze welds and overlays can be readily produced; in the latter case the bond is stronger than that of the plate material. Typical appn. are described and illustrated, the versatility of the process is shown where rigid control of deposit analysis is required. 13 ref.—W. D. B.

Automatic Inert-Gas-Shielded Arc-Welding Operations. H. O. Jones (*Weld. J.*, 1950, **29**, (8), 655–656).—Some recent developments are described, viz. automatic head, mech. wire feeder, &c.—W. D. B.

Applications for Helium in Inert-Arc Welding. M. J. Conway (*Weld. J.*, 1950, **29**, (7), 533–536).—A brief historical review of the development of He-shielded arc welding, its present appn., and possible future developments.—W. D. B.

"Heliarc" Welding in Production. T. E. Piper (*Weld. J.*, 1950, **29**, (9), 723–728).—The use of He-shielded arc welding in aircraft plant is described.—W. D. B.

Jigs and Fixtures for Inert-Gas Arc Welding. H. A. Huff, Jr., and A. N. Kugler (*Weld. J.*, 1950, **29**, (10), 893–903).—Special problems of jig design for inert arc welding are presented by such factors as magnetic and H.F. interference, back-up, &c. Some of the requirements are discussed and examples given.—W. D. B.

Modern Resistance Welding: A Precision Machine Tool. René Bouillon (*Rev. Soudure*, 1950, **6**, (3), 144–159).—A review of resistance-welding methods and equipment and the development of the mech. and elect. sections of welding machines. The advantages of rectification are stressed. —W. D. B.

Electronic Control of Modern Resistance-Welding Machines. F. Viart (*Rev. Soudure*, 1950, **6**, (3), 160–168).—V. discusses switching phenomena, synchronous control, and electronic contactors. Electronic control apparatus for welding at 50 c./s. and lower frequencies is described.—W. D. B.

Electronic Control of Welding Equipment. C. E. Wright (*Australasian Eng.*, **1951**, (Feb.), 50–56).—Read at the Welding Engineering Symposium, Sydney, Aug. 1950. Electronic control is used in automatic arc welding, in flash welding, in stud welding, in brazing, and in the allied field of metal cutting, but its largest field of appn. is in resistance welding. Regulators and electronic timing sequence controls are described, which give the precise control of welding current necessary for satisfactory work. All important types in regular commercial use, both in Australia and overseas, are included. 8 ref.—T. A. H.

Cold Pressure Welding. —— (*Mass Prod.*, 1951, **27**, (4), 74–83).—An account is given of the principles and practice of cold welding. The properties of the joint, the design of the welding tools and the weldabilities of different non-ferrous metals and alloys are considered. Appn. of the process are briefly discussed.—R. W. R.

***Metal Spraying of High-Temperature Metals and Alloys.** Robert T. Thurston and John Wulff (*Weld. J.*, 1950, **29**, (7), 313S–319S).—The metal spraying of some high-temp. materials (Mo, Ni, Co–Cr, Ni–Mo, and Cr–Mo–Fe) has been studied. Hydrostatic bursting tests showed a relationship between strength and oxide content. Marked densification may be obtained by sintering the deposit in H, especially in the presence of a liq. phase.—W. D. B.

***Determination of the Temperature of Sprayed-Metal Particles.** James E. Cline, Robert T. Thurston, and John Wulff (*Weld. J.*, 1950, **29**, (7), 320S–322S, 342S).—Tests on Zn, Al, Cu, Ni, Fe, and Mo show that sprayed metal particles are at their m.p. at the moment of impact. Calculations show that the contribution of the kinetic energy to heat content and temp. is negligible.—W. D. B.

Werkstoff und Schweissung. (Erdmann-Jesnitzer). See col. 256.

22 — INDUSTRIAL USES AND APPLICATIONS

A Survey of Commercial Wrought Light Alloys and Some Engineering Applications. H. A. Wainwright (*J. Record Trans. Junior Inst. Eng.*, 1949–50, **60**, 1–27).—A review of the properties and some appn. of wrought Al and Mg alloys. —N. B. V.

Basic Rules for the Application of Aluminium Castings. A. Buckeley (*Giesserei* (Sonderausgabe *Konstrukteur und Giesser*), **1951**, 64–69).—Phys., chem., and mech. properties of Al-base alloys are reviewed in relation to the appn. of Al alloy castings.—V. K.

Aluminium and Its Alloys in the Chemical and Processing Industries. S. M. Lawrence (*Indust. Chemist*, 1951, **27**, (316), 219–224).—L. describes the grades of Al and its alloys used in chem. engineering; fabrication methods and surface treatment are briefly considered. Recommendations are made for the operation and cleaning of Al vessels. Some recent appn. of Al in the gas, rubber, petroleum, canning, dairy, baking, and textile industries and in sewage disposal, H_2O_2 manufacture, the handling of foodstuffs, &c., are described.—R. W. R.

Use of Aluminium in Petroleum-Refinery Equipment. E. E. Kerns and W. E. Baker (*Petroleum Refiner*, 1951, **30**, (5), 123–129).—A number of actual refinery installations of Al are reviewed. These include its use for the roof of storage tanks, for tank vent valves and flame arresters, and for bubble caps in the distillation of crude oil and other refinery processes. Several examples are quoted of the use of Al heat-exchanger tubes, and detailed information is included on operating conditions. It has also been used for instrument tubing and for weather-proofing insulation on pipes, &c. Brief details are included on the use of Al for piping and also for pump and turbine packing. It is concluded that Al is suitable for use in several types of service, and in many cases is cheaper than alternative materials. It is stressed that considerable care is necessary in evaluating the corrosive nature of the environment, and attention is necessary at the design stage to reduce or prevent the possibility of galvanic corrosion.—G. T. C.

The Arvida Bridge : Design of the Aluminium Superstructure. C. J. Pimenoff (*Eng. J.*, 1949, **32**, (4), 186–195).—A general description is given of the first all-Al highway bridge at Arvida, P.Q., Canada, with details of specifications, loads, stresses, &c. Problems peculiar to Al construction are discussed, with indications of the methods by which they have been overcome.—N. B. V.

Light Alloy Hangar at London Airport. —— (*Metallurgia*, 1951, **43**, (260), 295–296; also *Engineer*, 1951, **191**, (4975), 721; *Engineering*, 1951, **171**, (4453), 654–655; and *Light Metals*, 1951, **14**, (160), 410–414).—The first all-Al aircraft hangar ever to be built is described.—F. M. L.

The New Aluminium Roof of the Milan Railway Station. —— (*Rev. Aluminium*, 1949, (161), 430–431).—An illustrated note.—N. B. V.

A Circus Has Been Built in Duralumin in Holland. —— (*Rev. Aluminium*, 1950, (165), 162–163).—N. B. V.

A Light-Alloy Dragline Bucket Lasts Four Times as Long as a Steel Bucket. —— (*Rev. Aluminium*, 1950, (163), 48–49).—Refers to experience gained at London Airport.
—N. B. V.

Prefabricated [Aluminium] Withes for Agriculture. M. V[ictor] (*Rev. Aluminium*, 1950, (164), 100–102).—A coil of hard-drawn Al wire, notched at short intervals, can easily be broken at the notches and the lengths used for securing vines, plants, &c., to wires or sticks.—N. B. V.

Aluminium Conductors. W. W. Schumacher (*Aircraft Prodn.*, 1951, **13**, (147), 21–26).—Appreciable savings in weight in aircraft elect. systems can be obtained by using Al instead of Cu conductors. The protection of Al terminals against corrosion, methods of terminal attachment, and cable-assembly tests are reviewed and discussed. 6 ref.
—W. A. M. P.

Cables Sheathed with Aluminium : Development in Great Britain. P. M. Hollingsworth (*Conf. Internat. Grandes Réseaux Elect., Paper* **203**, 1950, 29 pp.; *Elect. Eng. Abs.*, 1950, **53**, 453).—Of the various methods for sheathing cables with Al, only the swaging method has been developed in Great Britain to the stage of full-scale commercial prodn. A brief account of the technique and scope of the method is given. Tabular information on the phys. and mech. properties of Al-sheath cables so produced, and similar information on the characteristics of the finished cables, including mech. strengths and current ratings, are presented. Installation, jointing, and terminating procedure is described and illustrated in detail; finally, Al gas-pressure-type cables and Al conductor cables with Al sheaths are referred to briefly. About 200 miles of Al-sheathed cable have now been installed. Experience has shown that this type of cable is capable of economic prodn. and, apart from tech. considerations, acquires importance in view of the threatened shortage of Pb.

The Evolution of the Bicycle. [Use of Light Alloys]. Maurice Victor (*Rev. Aluminium*, 1950, (163), 71–79).—The use of light alloys in motor-assisted bicycles is described and illustrated.—N. B. V.

[The Use of Light Alloys in] The Equipment of the Paul-Emile Victor Expedition to Greenland. —— (*Rev. Aluminium*, 1950, (167), 248–251).—Prefabricated buildings, sledges, and other items of equipment were made in Al alloy.
—N. B. V.

[Use of Aluminium Alloys in] Mobile Medical Units in Algeria. —— (*Rev. Aluminium*, 1950, (164), 122–123).
—N. B. V.

[Light Alloy] Pradère Road Tankers. —— (*Rev. Aluminium*, 1950, (162), 22–24).—A number of French road tankers, mostly made in Al–Mg alloy, are illustrated.—N. B. V.

The C.I.M.T. Light-Alloy Railway Coach. Maurice Victor (*Rev. Aluminium*, 1949, (157), 239–255).—A detailed description is given of the design and construction of a light-alloy railway train running on pneumatic tyres.—N. B. V.

Light-Alloy Rolling Stock for London Transport. —— (*Engineering*, 1951, **171**, (4454), 705–707; also *Engineer*, 1951, **191**, (4970), 552–553).—Gives details of the construction of a light-alloy coach, employing an alloy of the following compn.: Mn 0·4–0·9, Mg 0·5–1·0, Si 0·75–1·25%, balance Al. In the heat-treated condition, this alloy possesses a U.T.S. of 20 tons/in.^{2}, and an elongation of 10% on 2 in.—D. K. W.

[Use of Aluminium Alloys in] The New High-Speed Launches of the Messageries Royales of Laos. —— (*Rev. Aluminium*, 1950, (165), 160–161).—N. B. V.

[Use of Light Alloys in] Air Transport. André Chevrier (*Rev. Aluminium*, 1950, (166), 184–191).—Some of the latest types of transport aircraft are described and illustrated.
—N. B. V.

[The Use of Light Alloys in] High-Power Electric Generators and Motors [for Aircraft]. Henry Lanoy (*Rev. Aluminium*, 1950, (164), 109–112).—N. B. V.

No Backlash with This Aluminium Fishing Reel. —— (*Modern Metals*, 1951, **7**, (2), 50).—The construction of an Al fishing-reel, including a backlash control device, is given.
—R. J.

New Projects in Light Alloys. —— (*Metal Treatment*, 1950, **17**, (62), 81–82).—An account of the construction of the "Dome of Discovery" at the London South Bank Exhibition, and of telescopic gangways for use with the "Queen" vessels at Southampton.—J. W. C.

Report of [A.S.T.M.] Committee B-7 on Light Metals and Alloys, Cast and Wrought. —— (*Proc. Amer. Soc. Test. Mat.*, 1950, **50**, 173–180).—See *Met. Abs.*, 1950–51, **18**, 315.

How to Select and Use Low-Melting [Bismuth] Alloys as Production Aids. O. J. Seeds (*Materials and Methods*, 1950, **32**, (3), 64–68).—Appn. of low-melting Bi alloys are reviewed, most of them resulting from the property that Bi-rich alloys expand on solidification and continue to expand when cooled.—J. W. C.

Highly Corrosion-Resistant Spring Material [Elgiloy] Finds Varied Use. Kenneth Rose (*Materials and Methods*, 1950, **32**, (3), 54–55).—Elgiloy, a Co–Cr–Ni alloy developed originally for use as a watch-spring material, has found several appn. as a corrosion-resistant spring material with mech. properties superior to those of steel.—J. W. C.

How Copper-Base Casting Alloys are Used in Electrical Industry. R. A. Colton (*Materials and Methods*, 1950, **32**, (6), 57–61).—The properties required and alloys available are briefly reviewed.—J. W. C.

Beryllium Copper Shoulders Defence Jobs. John T. Richards (*Steel*, 1951, **128**, (20), 83–84).—R. discusses the properties of Cu–Be alloys and describes typical appn. in the cast and the wrought states.—E. J.

[Copper Alloy] Cylinder Heads for Sleeve-Valve Aero Engines. J. Hinde (*Metallurgia*, 1951, **43**, (255), 11–14).—The developments in design and construction, necessitated by increased power demands, in sleeve-valve engine heads are outlined. Some of the problems involved in the manufacture of a two-piece head comprising a Ni-plated Cu–0·45–0·8% Cr base with a brazed-on O.F.H.C. Cu fin pack, are described.—F. M. L.

Magnesium Tubing Spun and Brazed to Form Critical Part. Don C. Law (*Materials and Methods*, 1950, **32**, (5), 65–67).—A small float used in hydraulic car lifts has been successfully produced by spinning and brazing Mg alloy tubing.—J. W. C.

Magnesium Alloy Star. —— (*Metal Ind.*, 1951, **79**, (2), 28–29).—Describes the construction of a Mg-alloy four-pointed star, 40 ft. from tip to tip and weighing 1790 lb.
—J. H. W.

[Nickel-] Plated Iron as a New Material in Vacuum Technique. W. Espe (*Slaboproudy Obzor*, 1950, 11, (Feb.), 31–39; *Physics Abs.*, 1950, 53, 803).—[In Czech]. The question of a reliable substitute for Ni in electronics is complicated by the fact that Ni combines a great number of valuable and important characteristics which can only be obtained, if at all, by a carefully balanced compn. of baser metals, in particular low-C Fe alloyed with Al. The best of these alloys has to be plated with Ni in order to obtain the required surface characteristics of the Ni, for which there can be no substitute. E. gives a very comprehensive and detailed survey of successful work in this direction in various countries.

Commercial Uses of Zirconium Developing. G. L. Miller (*Iron Age*, 1951, 167, (20), 83–87).—In Great Britain Zr is produced by a modified Kroll method. M. tabulates the mech. properties and compn. of commercial Zr, and describes its working characteristics and appn.—J. H. W.

Designers Look to Stronger Airframe Materials. Leo Schapiro (*Steel*, 1951, 128, (16), 77–80).—After discussing the progress in aircraft construction, S. reviews the developments in new ferrous and non-ferrous alloys, the latter including Al, Mg, and Ti alloys.—E. J.

Spinning Jets [for Production of Man-Made Fibres]. J. M. Pirie (*Textile Industries and Fibres*, 1951, 12, (7), 216–219).—The first requirements of metals for the spinnerets used in the prodn. of viscose rayon is the ability to resist the simultaneous actions of strong NaOH and dil. H_2SO_4 soln., thus immediately limiting the choice to noble-metal alloys, Ta, or Zr. Other requirements are suitability for a high degree of polishing, and resistance to "ring" formation owing to electrochem. action. The most widely used material for the spinnerets is probably Au–30% Pt, though Pt–10% Rh, which has approx. similar hardness but is slightly more difficult to work, is also employed. Ta jets are relatively more prone to blockage owing to ring formation, and the surface finish is relatively inferior. In the prodn. of cuprammonium fibres, the chem. conditions are not so severe and the jets are almost always formed from Ni. With the especially severe conditions in glass-fibre prodn. only Pt or Pt alloys give reasonable working life. It is considered that any future advances in spinnerets will be met. in character, leading perhaps to the use of alloys which can be hardened to a high degree by heat-treatment after forming and polishing. Methods in the prodn. of various types of fibres and the fabrication of the spinnerets are briefly outlined, and typical spinnerets are illustrated.—J. R.

Heat-Resistant Castings: Their Selection and Application. H. R. Clauser (*Materials and Methods*, 1950, 32, (4), 79–90).—Materials and Methods Manual No. 63. A comprehensive review. Data are given for high-temp. properties, fatigue, corrosion-resistance, machining, and welding properties, &c.—J. W. C.

Recommended Carbide Grades for Various Applications. —— (*Materials and Methods*, 1950, 32, (6), 85).—Materials Engineering File Facts No. 203. Manufacturers' recommendations for carbides for chip removal, wear appn., and impact appn. are collected on a single chart.—J. W. C.

Extruded Metal Shapes and Their Uses. T. C. Du Mond (*Materials and Methods*, 1950, 32, (2), 65–76).—Materials and Methods Manual No. 61. A general review.—J. W. C.

Aircraft Designers Look to "Newer" Structural Metals. —— (*Steel*, 1950, 127, (15), 82–84).—A report of a lecture by Fred R. Dent, Jr., who discusses the development of certain materials, such as Mg alloys and Ti, and the requirements from such materials in aircraft construction.—E. J.

23 — MISCELLANEOUS

Science in the Service of the Community. (Sir) John Anderson (*J. Inst. Metals*, 1951, 79, (6), 407–414).—Institute of Metals May Lecture 1951. The remarkable developments of the first half of the present century in the practical appn. of sci. discoveries to the fields of industry, metallurgy, medicine, biology, agriculture, &c., are briefly reviewed. The potentialities, both for construction and destruction, of the discovery of atomic fission are touched upon, and an urgent plea advanced for a wider dissemination of sci. knowledge as a means of directing these new forces towards ends that will benefit mankind.—Author.

†Le Chatelier's Principle and Its Applications to Metallurgy. B. Bever and R. Rocca (*Rev. Mét.*, 1951, 48, (5), 363–368).—B. and R. give a short historical survey of Le Chatelier's principle (briefly, that action and reaction are equal and opposite) and discuss the law math. They point out that this law is of very general appn. and give examples in metallurgy, such as the reduction of Fe_2O_3 to FeO, the increase of the solubility with temp. of H_2 in most solid and liq. metals, phase changes, and the theory of elasticity. 32 ref.—J. H. W.

Some Modern Developments in Copper Pyrometallurgy (with Particular Reference to North American Practice). W. B. Boggs (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (II), 771–783; discussion, 1047–1093).—See *Met. Abs.*, 1949–50, 17, 468.

Bronze Welding, Riveting, and Wiremaking by the Ancient Greeks. E. Anastasiadis (*Metal Progress*, 1950, 58, (3), 322–324).—A hollow-cast horse in bronze of the period 350 B.C. was made in several parts welded together: casting and weld metal were analysed and gave results showing them both to have Sn contents in the range from 11 to 13%, and Zn in the range 0·5 to 1%, together with Fe and Pb as impurities. Such compn. are found among present-day commercial bronzes. Cu was shown to have been welded with a 3% Sn bronze; bronze pieces were joined with rivets of bronze. Wire from the decoration of a Cu cup was microscopically examined and was seen to have been made by a drawing process.—F. A. F.

A Review of Productive Capacity in the Magnesium Industry. J. D. Hanawalt (*Metal Progress*, 1950, 58, (4), 512–518, 530).—At the end of the war military orders stopped and there was little peace-time demand for Mg. Since 1947 U.S. consumption has risen again, and one ingot plant is now producing at the rate of 22,500 tons/year. Since 1947 there has been a change-over in the relative importance of wrought and cast prodn., and the former now exceeds the latter in quantity. In spite of general inflation, Mg prices have remained unchanged or have advanced only ~10%. Considerably improved rolling-mill plant is now becoming available, and slabs weighing up to ~2000 lb. have been rolled. Extrusion of Mg alloys formed by mixtures of fine powders is now under development, and alloys can be produced in this way which cannot be cast. New finishing processes are also mentioned.—F. A. F.

The Nickel Industry: Twenty Years On. (Sir) William Griffiths (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (II), 848–880; discussion, 1047–1093).—See *Met. Abs.*, 1949–50, 17, 468.

The Zinc-Smelting Industry in Great Britain. S. W. K. Morgan (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (II), 784–808; discussion, 1047–1093).—See *Met. Abs.*, 1949–50, 17, 468.

Occupational Diseases in the Light-Metal Industry. Karl H. Sroka (*Metall*, 1950, 4, (17/18), 394–397).—Although there are certain inherent, secondary dangers associated with the electrothermal prodn. of Al, proper precautions render them negligible. The working and fabrication of Al and its alloys are harmless, except in so far as they are associated with high-speed machinery, with consequent danger from fine metal splinters, &c. Light-metal household utensils are completely harmless; Al dust and certain Al compounds are used in medicine.—E. N.

The Toxicity of Beryllium. Maurice Déribéré (*Usine nouvelle*, 1950, 6, (47), 40).—The many uses of Be are briefly surveyed, and it is stated that although the toxicity of the fluorides and oxyfluorides is established it is unreasonable to attribute similar injurious properties to Be ions and its oxide or silicate. In fact, no case of " poisoning " has been reported in Europe from the use of fluorescent lamps containing a Be-bearing powder. After discussing the conditions relating to the manufacture and use of fluorescent lamps, it is concluded that except for certain very exceptional risks occasioned by a special sensitivity to Be, the use of Zn–Be silicate in fluorescent lamps has no hazards for manufacturer or user.—H. A. H.

Beryllium [Poisoning] Symposium. —— (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-1803), 68 pp.).—A review of the known cases of Be poisoning, including the conditions of exposure.—B. W. M.

Cadmium Poisoning from Cadmium-Plated Food Containers: Literature Review. —— (*Plating Notes*, 1950, 2, (6), 189–194).—Severe poisoning can occur if Cd is used for plating food containers, or any article with which food or drink may come in contact. Abstracts of 11 articles are given.—T. A. H.

The Possibilities of Mercury Poisoning in Instrument Shops. Robert F. Angell (*Petroleum Eng.*, 1951, 23, (2), C18).—The methods by which Hg can be absorbed into the body are described, and the symptoms of Hg poisoning are listed. A number of precautions for avoiding Hg poisoning in enclosed shops, laboratories, &c., are given. 6 ref.—G. T. C.

Occupational Diseases Encountered in Working Metal Surfaces. [I.—] Dust Dangers, Skin Injuries. [II.—] Metallic Poisoning in Working Metal Surfaces. Karl H. Sroka (*Metalloberfläche*, 1950, [A], 4, (10), 154–157; (12), 187–189).—[I.—] A description of the hazards encountered: (1) pneumokoniosis (silicosis, &c.) arising from the dusts generated during grinding and polishing operations; this can be overcome by the use of synthetic SiO_2-free materials for these operations; (2) skin discoloration and dermatitis caused by the absorption of (*a*) abrasives and metal particles during grinding and polishing, (*b*) oils during machining operations, and (*c*) chemicals during pickling; these can be overcome by the use of barrier creams; and (3) burns during welding operations. [II.—] A discussion of the toxicity, pathology, and prophylaxis of the following metals (or their alloys or compounds): Pb, Sn, Zn, Fe, Cu, Ni, Mg, Cd, Hg, As, and Cr.—E. N.

Industrial Diseases. —— (*Indust. Finishing* (*Lond.*), 1951, 3, (33), 677).—A detailed analysis of the 493 cases reported in the United Kingdom during 1950 under the Factories Act, 1937, or under the Lead Paint (Protection against Poisoning) Act, 1926. These include 57 cases of Pb poisoning, and 143 cases of Cr ulceration, of which 100 were reported from the Cr-plating industry.—H. A. H.

Possible Hazards from Chemical Contamination in Water Supplies. C. W. Muehlberger (*J. Amer. Water Works Assoc.*, 1950, 42, (11), 1027–1033; discussion, 1033–1034).—M. considers the dangers to health which may aries from drinking chem. contaminated water over a considerable period of time. The effects of Al, Cu, Fe, Pb, Mn, Se, Na, and Zn, are briefly discussed. The toxicity of Cr and Cd is more fully discussed, and results obtained from experiments on animals are given, but the available data are considered to be inadequate, conflicting, and unsatisfactory, and it is impossible to form any opinion as to safe limits for these contaminants. 17 ref.—W. A. M. P.

Industrial Hygiene at American Smelting and Refining Company. John N. Abersold and K. W. Nelson (*Trans. Amer. Inst. Min. Met. Eng.* (in *J. Metals*), 1951, 191, (1), 33–36).—Although dealing mainly with the occupational hazards—dusts, fumes, gases, toxic materials, heat, humidity, radiation, and noise—encountered in the smelting and refining of Pb, Cu, Zn, Ag, and Au, and by-product metals such as As and Cd, this article is of general interest. It describes, in some detail, the apparatus and techniques employed in the evaluation of the potentially harmful factors through air sampling, micro-analysis, and other means, and the appropriate methods of medical and engineering control which are utilized wherever necessary.—E. N.

Working Conditions in Certain Melting and Smelting Processes. W. A. Attwood and W. B. Lawrie (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (I), 441–448; discussion, 449–483).—See *Met. Abs.*, 1949–50, 17, 613.

Physiological Effects of High Temperature and Humidity on Workers in the Mining and Metallurgical Industries. David Brunt (*Proc. Fourth Empire Min. Met. Congr.*, 1950, (I), 325–338; discussion, 449–483).—See *Met. Abs.*, 1949–50, 17, 613.

Research Associations: The B.N.F.M.R.A. —— (*Mass Prod.*, 1951, 27, (1), 90–95).—A brief account of the organization, equipment, and work of the British Non-Ferrous Metals Research Association.—R. W. R.

New Hot Laboratories at Argonne. H. L. Hull, R. C. Goertz, and K. R. Ferguson (*U.S. Atomic Energy Commission Publ.*, 1949, (AECD-2990), 13 pp.).—A description of laboratories in which samples up to 100 Curies, 1 meV. gamma can be handled with safety. In the met. cave, a Rockwell hardness tester has been operated successfully by the Master-Slave manipulator, and other apparatus includes a tensile-strength machine, profile-measuring apparatus, and sp.-gr. apparatus. Special opt. equipment is under design for macro and micro inspection and for complete remote-control metallography.—B. W. M.

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[For review of Vol. I, see *Met. Abs.*, 1949–50, 17, 141. This supplement brings the information in Vol. I (published in 1948) up to date by giving additional new structures and correcting earlier data.]

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25 — BOOK REVIEWS

Non-Ferrous Metals (Wrought). Report of a visit to the U.S.A. in 1950 of a Productivity Team representing the Wrought Non-Ferrous Metals Industry. 9¾ × 7½ in. Pp. xiv + 96, with 28 illustrations. 1951. London: Anglo-American Council on Productivity, 21 Tothill Street, S.W.1. (4*s*. 6*d*., post free.)

The report of the Productivity Team on the American Non-Ferrous Metals Industry which visited America last year, under the ægis of the Anglo-American Council on Productivity, has just been issued. It makes very interesting reading. One opens a report of this kind with something of the keen expectancy expressed by Keats when he first read Chapman's " Homer ", and perhaps it is not inapt to say that the reaction produced on the mind of the reader of this Report, albeit in a more prosaic way, was not unlike that in the poet's mind.

It is not proposed in this notice, to discuss the Report in any detailed way; that will be done, it is hoped very extensively and very conscientiously, in all our establishments given to the production of semi-manufactures in the non-ferrous metals. Rather is it the object of the writer to underline the general conclusions arrived at by the Team.

In the general details given of the working and metallurgical processes examined there appears to be little that is new, and the British product may be compared quite favourably with the American product in all that goes under the general term of quality. Indeed, in some cases the British product is reported as being of slightly superior quality. What has impressed itself on the minds of the Team more than any other single factor—and this is the burthen of the Report—is the greatly increased productivity in the United States factories. It has often been claimed that in the small jobbing factory given over to the processes of making sheet, tube, and the like, what are vaguely called " mass-production methods " cannot be applied, and that high productivity must be allied to large expensive pieces of plant. The Report shows quite clearly that this is a misconception, and the value of the Report lies in the clear and concise way in which the Team have pin-pointed the essential methods whereby output is significantly increased without the use of mammoth equipment.

Perhaps the first thing that strikes the reader is that although productivity is very much greater in the United States, there is no magic in the means of its achievement. The writer was particularly impressed by one phrase in the Report which stated that management in the American factories was aggressive. It is clear that there is a tireless enthusiasm for making two blades of grass grow in the place of one. It is significant that a great number of these gains have been made at little cost, and with little expensive equipment.

Among a number of factors which contribute to this end there are two which seem to be most relevant. One is most certainly the fact that the American worker can reap personally and immediately, the benefit of this increased productivity. He gets what he earns and is able to spend freely. This is an inevitable result of conditions of teeming productivity, as the economists so often remind us. It points quite clearly to the fact that the most important factor in our economy in this country lies in the industrial field. Nevertheless the fact has to be faced, that unduly high taxation, which reduces the worker's earnings, inhibits his effort.

The second factor is the way in which effort is integrated. Both management and trade unions are at one in this matter of productivity. They appear to have learned the lesson that the only true bargain is one which profits both participators. The real key to the success of the joint efforts of management and workers is not primarily the rate for the job, but the extra ton which can be produced for the same effort. This is a very vital and significant emphasis, and the Team rightly make much of this in their Report.

Unanimously the Team indicate that the solution of this problem in this country lies in the hands of management. The writer, while agreeing that this must always be so, feels that the problem is not a simple one, for it is not in the first instance a problem in mechanics, but a problem in human relations. If it is the problem of management, then it demands wisdom and persuasiveness. It calls not only for technical ability, but for a very real understanding of human nature, and a very large measure of patience and winsomeness. From the workers it demands an intelligent appreciation of the essential rightness of the American conception. It has been said that there is no better turf in the world than that on the Cumberland Fells, and that the best sheep in the world are Cumberland sheep. Each has made the other supreme. This is the simple parable which has to be understood, and if this can be achieved, much can be done in the future. There is a need for a statesmanlike approach to the problem which appears to be the first essential step for success.

The various sections of the Report offer much encouragement to the reader. Standards of quality and tolerance, tools and methods are not very different in the two countries. The various products are dealt with adequately, except perhaps in the section devoted to extrusion, which appears a little inadequate. The sections on economics, administration, and costing are of considerable interest, and in the appendices there is a quantity of condensed information which will prove to be of much value.

In conclusion, the writer would like to congratulate the Team on a workmanlike, comprehensive, and understandable report. In particular he would comment on the wisdom and common sense of their various conclusions.

CHRISTOPHER SMITH.

Atom Movements. A Seminar on Atom Movements Held During the Thirty-Second National Metal Congress and Exposition, Chicago, 21–27 October 1950, Sponsored by the American Society for Metals. 9 × 6 in. Pp. [vii] + 250, illustrated. 1951. Cleveland (3), O.: The Society, 7301 Euclid Avenue. ($5.00.)

This volume is another in the series which contain the papers presented at Seminars of the American Society for

Metals. It covers more or less the information which one would expect to find in a monograph entitled "Diffusion in Metals".

The obvious advantage of this type of publication is that each paper is written by a competent authority on his own part of the field covered by the general title. In fact, each author in this volume is known to have made essential contributions to the special field he is reviewing, and the contributions are therefore of a high standard. Generally, the disadvantage of this type of publication is an unevenness in treatment. This drawback has been avoided, however, in the present volume to an extent which the reviewer has not encountered hitherto with similar publications. In a few places only might some more space have been devoted to the subjects.

C. Wells and R. E. Hoffmann deal with the experimental techniques and the analysis of diffusion data; the tracer methods include a new autoradiographic technique; the discussion of the other methods is very brief, too brief perhaps, but the bibliographies are fairly complete. Some readers might also wish to see a complete list of diffusion data, but this gap is not serious, as an excellent tabulation is available in this country in Smithells's "Metals Reference Book".

Much attention is paid to the recently discovered Kirkendall effect, particularly in the article by J. Bardeen and C. Herring, while H. B. Huntington discusses the mechanism of diffusion primarily from an energy point of view. C. E. Birchenall's empirical comparison of diffusion data is cautious in speculation and shows the "weak spots" which future research may profitably attack. D. Turnbull's article reveals the considerable advances which the theory of grain-boundary and surface diffusion has made during recent years.

There are three reviews dealing with problems in which diffusion phenomena play an important role: oxidation of metals (C. Wagner), internal oxidation (F. N. Rhines), and sintering (P. Duwez). Wagner confines his considerations to the parabolic oxidation of metals, almost the only type which appears to be amenable to theoretical treatment at present. Rhines's review includes some special features of gas–metal diffusion and then turns to the effect of internal oxidation. Duwez includes a discussion of Kuczynski's method (for determining self-diffusion rates) and its possible limitations. There is finally a review on migration of grain boundaries by J. E. Burke.

The introduction dealing with the phenomenological description of the diffusion process is written by L. S. Darken and the conclusion by R. F. Mehl. If the reader is still unaware of it, Mehl's summary shows most clearly that we are still in a very early stage of interpreting diffusion data and understanding the true mechanism of atom movements. Certainly no one can be blamed for this. Many research workers, including the authors of this book, have done excellent work, but the mechanism of diffusion, even in atomic solutions, is complicated, and much work remains to be done.

The bibliographies are full, but without making a special point of doing so the reviewer has noticed at least fifteen misprints in authors' names, particularly foreign ones—an excessive number.

The book, of which the paper, print, and binding leave nothing to be desired, is thoroughly recommended to anyone concerned with or interested in diffusion in metals.

O. Kubaschewski.

Métallographie : Les diagrammes d'équilibre par questions et réponses. Par A. Glazunov. 24 × 15 cm. Pp. [vii] + 240, with 425 illustrations. 1951. Paris (6^{e}): Dunod, 92 rue Bonaparte. (Br., 1280 fr.)

The author, who is a Professor of the Pribram School of Mines, states in his preface that, in his experience, textbooks on metallography do not stress sufficiently the importance of deriving the maximum of information from an equilibrium diagram, and this point of view is endorsed by Professor Chaudron, who contributes a preface to the French edition of the book. As in Dr. Hume-Rothery's recently published dialogues between an ancient and a modern metallurgist, the information is given in the form of questions and answers, and, as one reviewer of Dr. Hume-Rothery's work commented, one can only marvel at the ability of the student to ask exactly the right question at the right time.

The first section of the book deals with such concepts as components, phases, degrees of freedom, entropy, thermodynamic potentials, and free energy, and covers the derivation of the phase rule and the phenomena observed in the freezing of single substances. The second, and much the longer section, deals with binary systems. All possible forms of equilibrium diagram are considered, and the illustrations are based on salt mixtures as well as on alloy systems. The treatment generally follows classical lines, but the relationship between the phase fields of the diagram and the free-energy curves of the phases is dealt with adequately.

H. W. L. Phillips.

Metallographie : Grundlagen und Anwendungen. Von Otto Mies. 3. Auflage. 22·5 × 15·5 cm. Pp. 68, with 186 illustrations. 1949. Berlin–Göttingen–Heidelberg: Springer-Verlag. (DM 3.60.)

This book is intended as an introduction to metallography for the use of students and trainees in industry. Of its 68 pages, 46 are devoted to the ferrous metals, and of the remainder, 11 deal with copper alloys, 9 with light alloys, and 2 with bearing metals and zinc. Each section gives a brief account of manufacturing processes, a glimpse at the appropriate part of the constitutional diagram, and a description of a few typical microstructures. The information given appears to be accurate.

H. W. L. Phillips.

Engineering Materials Manual. Edited by T. C. Du Mond. 11½ × 8½ in. Pp. [iii] + 386, with many illustrations and tables. 1951. New York: Reinhold Publishing Corp. ($4.00); London: Chapman and Hall, Ltd. (36*s*.)

This volume describes in the minimum of words, with the aid of many diagrams, tables, and illustrations, most of the materials now being used in engineering works. The 28 sections into which it is divided each present an overall picture of a single material or group of materials. They are essays on engineering materials which have been written, by 14 separately named authors, for the series of "Materials and Methods Manuals" which have been published in *Materials and Methods* during recent years. They were produced primarily for those practical engineers who may be potential users of the materials, and the limitation of space, in view of the width of field covered, has been met by a praiseworthy economy in words.

There are, to begin with, sections on the standard engineering steels, stainless steels, high-strength low-alloy steels, tool steels and free-cutting steels, clad steels, and grey-iron castings. Wrought and cast aluminium and magnesium alloys, nickel alloys, engineering bronzes, beryllium copper, and bearing metals are then described in sections which constitute the non-ferrous metals group of subjects. The latter half of the book deals with cemented carbides, ceramics, rubber, plastics, hard-facing materials, finishes for metals, and adhesives. The sections dealing with subjects of which the reviewer has a more intimate knowledge proved on reading to be well-balanced presentations from the American point of view. They do not cover European-made material specifications, but there are a few brief references to European developments.

Although not intended primarily for metallurgists, there are few members of the Institute who would not find this volume a useful addition to their library, and it will be especially valuable to those connected with production engineering. Its value as a reference book will be greater in the United States than in other countries, but it provides a mass of authoritative information which would be difficult to find, had the authors not gathered it conveniently in this manner. The book will be a timely aid to the many designers and technical assistants who must day by day consider the choice of materials for essential components, when the use of any scarce material must be justified, and almost all are scarce.

T. Henry Turner.

Die metallischen Werkstoffe. Aufbau–Prüfung–Eigenschaften. Von Karl Wellinger und Paul Gimmel. 24 × 17 cm. Pp. [vii] + 130, with 142 illustrations. 1950. Stuttgart 1: Verlag Konrad Wittwer. (Brosch., DM 8.50; geb., DM 9.80.)

This excellent little manual presents in a concise form, mainly for undergraduates, but also for graduates in industry, the present state of knowledge in the field of ferrous and non-ferrous metals. Professor Wellinger's and his co-author's researches at the Staatliche Materialprüfungsanstalt Stuttgart are well-known beyond the frontiers of their country, but it is nevertheless a very laudable achievement that, after having written numerous papers on highly specialized subjects, they can condense into 125 pages a general statement of the problems of testing and of the characteristics of all metals which are of interest to engineers.

While the first part of the book, dealing with the structure of metals, does not go beyond the usual material presented in similar text-books, the second and third on testing methods and machines discuss in an outstandingly clear manner such topics as the significance of fatigue tests, the influence of various factors on fatigue life, up-to-date fatigue-testing machines, creep testing, methods of strain measurement, non-destructive testing of metals, as well as other means available in engineering for testing materials and structures. The fourth part of the book deals with the properties of metals, their behaviour at high and low temperatures and their weldability being amongst the points discussed.

The book is profusely illustrated with informative tables and diagrams, as well as excellent photographs. The material presented is up to date, and the authors have an unusually pleasant and simple way of telling their story. English readers would find one possible shortcoming in the absence of a list of references, which would enable students to gather more detailed and profound information on the very many subjects which are merely enumerated and briefly defined in this short, but otherwise very valuable, manual.

Nicol Gross.

Aluminium-Taschenbuch. Zehnte Auflage. Bearbeitet von Johannes Reiprich und Wilhelm v. Zwehl. 18 × 13 cm. Pp. xviii + 628, with 300 illustrations and 130 tables. 1951. Düsseldorf: Aluminium-Zentrale e.V., Alleestrasse 31. (DM 12.60.)

This, the latest edition of a familiar work, has been compiled by 24 authors, well known in this country as being experts in their particular fields. It is therefore not surprising to find that the information it contains is accurate and up to date. A short introduction dealing with the history and process metallurgy of aluminium is followed by an account of the metallography, physical properties, and chemical properties of the industrial alloy groups. A section on testing follows, covering chiefly mechanical and physical methods. Chemical analysis is not dealt with. Next come lengthy sections dealing with casting, working, fabrication, machining, jointing, and various methods of protection. The remainder of the book covers technical applications of aluminium and various alloys.

The text is concise and readable, and is illustrated by numerous diagrams. German specifications are quoted in full, and much additional information is given in tabular form. The bibliography covers French, English, American, and Italian sources as well as German, and is up to date.

H. W. L. Phillips.

Aluminium-Merkblätter. 21 × 15 cm. [1950.] Düsseldorf: Aluminium-Zentrale e.V., Alleestrasse 31.

These are data sheets similar to those issued by some industrial organizations in this country. The information varies from sheet to sheet, but generally covers properties, specification data where available, and hints on fabrication or assembly. Those already issued are as follows: A1.—Aluminium Roofing (Introduction). B1.—Bending of Semi-Finished Aluminium Products. E1.—Aluminium in Switchgear Assemblies. E2.—Aluminium Cables (Jointing Methods). G1.—Melting and Casting of Aluminium. K1.—Designing Aluminium Sand and Gravity-Die Castings. K2.—Designing Aluminium Pressure Die-Castings. K4.—Anodized Aluminium for Decorative Purposes. O1.—Maintenance of Architectural Members and Ornaments. O2.—Chemical Oxidation (M.B.V. and E.W. Process). O5.—Grinding and Polishing. O6.—Pickling and Deep Etching of Aluminium. O7.—Care of Aluminium Articles in the Household and Commercially. V1.—Autogenous Welding. W2.—Wrought Aluminium Alloys (Survey). W4.—Chemical Resistance of Aluminium Materials (Tables). W5.—Simple Means of Distinguishing Between Aluminium Alloys. W10.—Pure Aluminium.

H. W. L. Phillips.

Notes on the Prehistoric Metallurgy of Copper and Bronze in the Old World. By H. H. Coghlan. (Occasional Papers on Technology, No. 4). 9¾ × 7½ in. Pp. 131, with 21 illustrations in the text and 16 plates. 1951. Oxford: Pitt Rivers Museum. (15*s*.)

This admirably illustrated monograph is the fourth of a series of "Occasional Papers on Technology" published under the auspices of the Pitt Rivers Museum at the University of Oxford. The object of those concerned with its production has been to indicate the means which are now available for a more scientific and systematic approach to a better understanding of the level of technique that had been reached by primitive workers in metals at various periods to which objects of copper and bronze have been attributed as the result of archæological research. The author of the major portion of this monograph (Mr. Coghlan) has rightly taken the line that it is only by having some knowledge of the means by which such objects would be made at the present day that any real attempt can be made to give a satisfactory explanation of the procedures that were followed in early times. This he has succeeded in doing by providing the archæologist and the general reader with an appropriate summary of the elements of metallurgy, much of which must have been known to early workers, though, perhaps, somewhat vaguely, but without which success in their efforts could not have been achieved at all. He has blended this useful instruction with running references to actual objects that have survived, and has thus emphasized how much was known and how considerable was the skill attained, even with the primitive equipment then available. It is, perhaps, in Chapters V and VI, which deal with the actual working and manipulation of copper and bronze, that the author has brought his own technical knowledge as an engineer to bear most successfully on those problems which, as curator of a museum himself, he knows to be of interest and value to those engaged in archæological research. He gives many examples in which an acquaintance with the basic craftsmanship of to-day is essential to those who set out to classify objects of antiquity without falling into error, which, as he says, (p. 99), "has rendered so many museum specimens misleading" in the past.

The valuable aid that metallographic and spectrographic techniques give to investigations of this kind is abundantly evident in the admirable contributions made by the British Non-Ferrous Metals Research Association to this monograph. The author is fully aware, however, that while the tendency is to rely more and more on such physical methods of examination, there is, at the same time, a tendency to neglect the older technique of chemical analysis. It is indeed a pity that, at the present time, there are so few available who have the time, the inclination, or even the skill to fill this gap. He says, (p. 31) "the ideal way in which to establish a case would be to have a comprehensive series of analyses and metallographic reports of early copper objects . . . unfortunately nothing like this desirable state of affairs exists . . . analyses are all too rare". He draws timely attention (p. 35) to the limitations of the spectrographic method. The fear that the removal of portions sufficient for analysis may damage the specimen, can, in many cases, be overcome by judicious selection from the less prominent positions. The value of the analysis may outweigh any objections. Moreover, the possibilities presented by modern micro-analytical methods should help to obviate these objections.

Although this monograph is intended to deal only with copper and bronze in antiquity, it is to be hoped that, in time, a closer approach to the problems of primitive metallurgy may be made by an integration of knowledge appertaining to the early procedures with regard to all the more common metals which were known and worked. It is possible that the working of the more easily fused metals, lead and tin, may have pointed the way to solving some of the difficulties in dealing with those of higher melting points like copper, gold, and silver; while the contribution of the "potter", with his knowledge of "refractories" and his ability to reach high temperatures, may, as the author clearly indicates, (p. 64), have been of considerable help to the primitive worker in metals.

A valuable part of the monograph is the long list of references—some 70 to 80 in number—to previous work bearing on the questions to which the author, and those who have collaborated with him, have directed their attention.

S. W. Smith.

Corrosion Guide. By Erich Rabald. 24 × 15·5 cm. Pp. vii + 629, with 16 illustrations. 1951. Amsterdam: Elsevier Publishing Co., Inc. (45.50 fl.); London: Cleaver-Hume Press, Ltd. (90*s.*)

The "Corrosion Guide" provides engineers, chemists, and metallurgists with a reference book that should prove a useful supplement to the text-books and other works on the corrosion of metals and the deterioration of other constructional materials.

The General Introduction consists of short chapters on the choice of materials, principles of corrosion, measurement of corrosion-resistance, and hints on using the tables. The main part of the book (pp. 49–605) consists of tables in which are summarized the orders of resistance of metals, alloys, and non-metals towards corrosive reagents. An index (pp. 621–629) facilitates reference to a material or a corrosive reagent, whilst a bibliography lists books on corrosion and journals in which corrosion papers are to be found.

The book should prove very useful, although readers should realize that the information can be given only in the broadest terms. A degree of resistance that is satisfactory to one user may be inadequate for another. The introductory chapters are generally sound, although more references might have been given to British and American work. The book is very well produced, and there appear to be few printing errors. The wording could be improved in places, but the meaning is quite clear.

F. Wormwell.

International Conference on Hot-Dip Galvanizing Held at Copenhagen, 17–21 July 1950. 9⅛ × 6¼ in. Pp. 161, illustrated. 1951. Oxford: Zinc Development Association, Lincoln House, Turl Street. (30*s.* net.)

An account of the proceedings at the Conference, organized by the Hot-Dip Galvanizers Association, together with revised texts of the papers presented and the discussions on them. Ten papers are included, and these have been abstracted individually under the appropriate headings.

G. T. Colegate.

Foundry Practice. By William H. Salmon and Eric N. Simons. 8½ × 5½ in. Pp. xiii + 384, with 192 illustrations. 1951. London: Sir Isaac Pitman and Sons, Ltd. (30*s.* net.)

In the preface the authors comment on the existing books on the subject of foundry practice in the following way: "Most of these books are, however, written by metallurgists or professors for the general information of metallurgical students or industrial workers. They are sometimes highly technical and cluttered up with a mass of theoretical material that the student does not immediately need. They are occasionally difficult to follow . . ." And yet, in the selected list of forty-odd references given on this subject by the authors, I was able to detect the names of three professors only. The number of metallurgists is more difficult to arrive at, as there is no clear way of defining always the qualifications for a metallurgist. In any case the number of metallurgists who have written books on foundry practice has been increased by one, as at least one of the two authors of the present volume admits his qualifications of this learned and esteemed profession. I would suggest on the other hand, that it is a great pity that so few professors have been inspired in the past to write on this topic. It is also a pity that at least one of the present authors was not a professor. Yet another chance of having at last a thoroughly good book on foundry practice has thus been lost.

For their prospective typical reader the present authors have assumed mainly a student proceeding to take a City and Guilds of London Institute Examination for intermediate students. They have assumed that he knows very little and that he needs guidance even in the problem of calculating the areas of simple geometrical figures. Now, that may be a right or wrong assumption in itself, but what is certainly not right for a good book on foundry practice is to "clutter it up" with all subjects and topics that are related to foundry practice and are normally taught and written about separately and could justifiably be included only in larger volumes of "Foundry Handbooks". In this way, many pages of the present book are wasted that could have been devoted to much more essential features of foundry practice. For example, the authors assume, without warning the reader in advance about this, that to them "Foundry Practice" is a term synonymous with "Sand Foundry Practice". In other words the book deals only with the question of production of shaped castings in sand moulds, and by the gravity process of pouring only. There is no treatment of any other methods of founding.

In dealing with the main subject of their book the authors have taken great pains and described in much detail all stages of manufacture of a shaped sand casting, especially when made in a ferrous alloy. The method of presentation adopted throughout is that of giving a very simplified account of how things are done in a sand foundry and even why they are done in a particular fashion. On the other hand, little effort has been made to explain why one gets a particular result. This is a serious weakness of a book intended for an educational purpose.

Strict adherence to presentation of the subject of foundry practice based on the empirical data only is not necessarily undesirable. But what is surely undesirable is to neglect such modest contributions made by science to founding as are available. No professor could, for example, allow the following statements to pass in an examination paper—even for an intermediate standard—without resorting to the red pencil: "in annealing . . . molecules or ions of a dissolved substance move freely through the solvent" (p. 295); "in non-ferrous melting the presence of hydrogen or sulphur dioxide in the metal . . . causes gas holes just below the surface of castings" (p. 259); "impurities such as silicon in copper alloys and a low tin content are factors causing tin and lead sweat" (p. 288). Examples of this type are particularly numerous on those pages of the book dealing with foundry metallurgy.

This latest addition to the literature on foundry practice is likely to prove useful to a beginner interested in the "know how" of making a sand casting. The presentation is logical, the writing is clear, and the illustrations are instructive. In addition numerous questions and examples are provided to stimulate true interest in the subject. The book could not be recommended, however, to a more advanced student of this subject, as it contains little new either in knowledge or ideas.

V. Kondic.

Centrifugal Casting by the Cire Perdue Process. By C. Rosen. $9\frac{1}{4} \times 6\frac{1}{4}$ in. Pp. 204, with 194 illustrations. [1950.] London: Contact Publications, Ltd. (for David Wallace, Ltd.) Distributed by The Bruce Publishing Co., Ltd., Watford, Herts. (52*s.* 6*d.*, post free.)

The jewellery trade has in recent years shown an increasing interest in the possibilities of the "lost-wax" process for making small—and often very intricate—castings in gold and silver alloys; and the supply houses, particularly in America, have put on the market several centrifugal machines for the manufacturing jeweller. Mr. Rosen's book is more limited in scope than the title might suggest and is intended to provide working instructions for jewellers undertaking the process for the first time.

It is obviously written by one technician for other technicians, and some of its faults of presentation will possibly not worry many of the readers for whom it is intended. They will, presumably, know the broad outlines of the process and will not be puzzled by the absence of any adequate introductory description, and they may be ready to accept irrelevant data, lengthy extracts from catalogues of diffusion pumps they will never use, or occasional uncouth expressions such as "vacuumizing". They may, however, complain with justification of the make-up of the volume, which is unforgiveably slipshod in the arrangement of illustrations and tables. The illustrations, most of which are clear, boldly drawn, and heavily shaded in a pleasing style by an architectural draughtsman, Mr. A. M. Tribich, add greatly to the understanding of the book, but their value is unfortunately reduced by the absence of captions or references to the text and by the illogical way they are distributed.

Technically, this account of the investment casting of jewellery is little more than adequate. No discussion of the metallurgical aspects of the process is attempted and there is no suggestion that some alloys may be easier to cast than others. Indeed, no mention at all is made of defective castings or of how even the commonest faults may be recognized and overcome. The small amount of technical data is not always either helpful or accurate and there are far too many odd statements such as "a concentrated solution of silica and water will rapidly gelatinize unless subjected to heat-treatment". Readers should be warned that success in investment casting demands a good deal more than manual dexterity in the operations described and illustrated in this volume.

J. C. Chaston.

Cold Working of Metals. A Seminar on the Cold Working of Metals Held During the Thirtieth National Metal Congress and Exposition, Philadelphia, 23–29 October 1948, Sponsored by the American Society for Metals. 9×6 in. Pp. [iv] + 364, with numerous illustrations. 1949. Cleveland 3, O.: The Society, 7301 Euclid Ave. ($5.00.)

The individual contributions to this "seminar on the cold working of metals" have been abstracted in *Met. Abs.* in some detail. It remains to commend this volume as a whole to the attention of all those interested in modern views on the effects of cold work on the structure and properties of metals. The aim of the organizers, as explained in the introduction by M. Gensamer, was that the seminar should be "of lively interest to research people and . . . should emphasize the fundamental and scientific features of the topic and avoid attempts to bring out discussions of commercial applications". The seventeen papers provide an excellent account of current American work and views on such matters as the structure of slip bands and cold-worked metals, twinning, banding, deformation and recrystallization textures, plastic aftereffects, the mechanisms involved in creep and fatigue, and the influence of cold work on precipitation-hardening and on the martensite reaction. Some contributions, such as the excellent address by W. Shockley on dislocation theory, provide a refreshingly personal insight, authoritative yet modest, into the philosophy of the metal physicists. Some of the others are inclined to be unnecessarily difficult and obscure, but all are conscientiously prepared reviews, all are well documented, and nearly all will present to most readers fresh points of view. The only regret is that there is no record of the discussions.

J. C. Chaston.

Werkstoff und Schweissung. Handbuch für die Werkstoff- und werkstoffbedingte Verfahrenstechnik der Schweissung. Bearbeitet und herausgegeben von Friedrich Erdmann-Jesnitzer. Band I. 25×18 cm. Pp. xxiv + 1002, with 1117 illustrations and 196 tables. 1951. Berlin NW7 (Ostberlin): Akademie-Verlag G.m.b.H., Schiffbauerdamm 19. (Geb., Band I + II DM 160.—.)

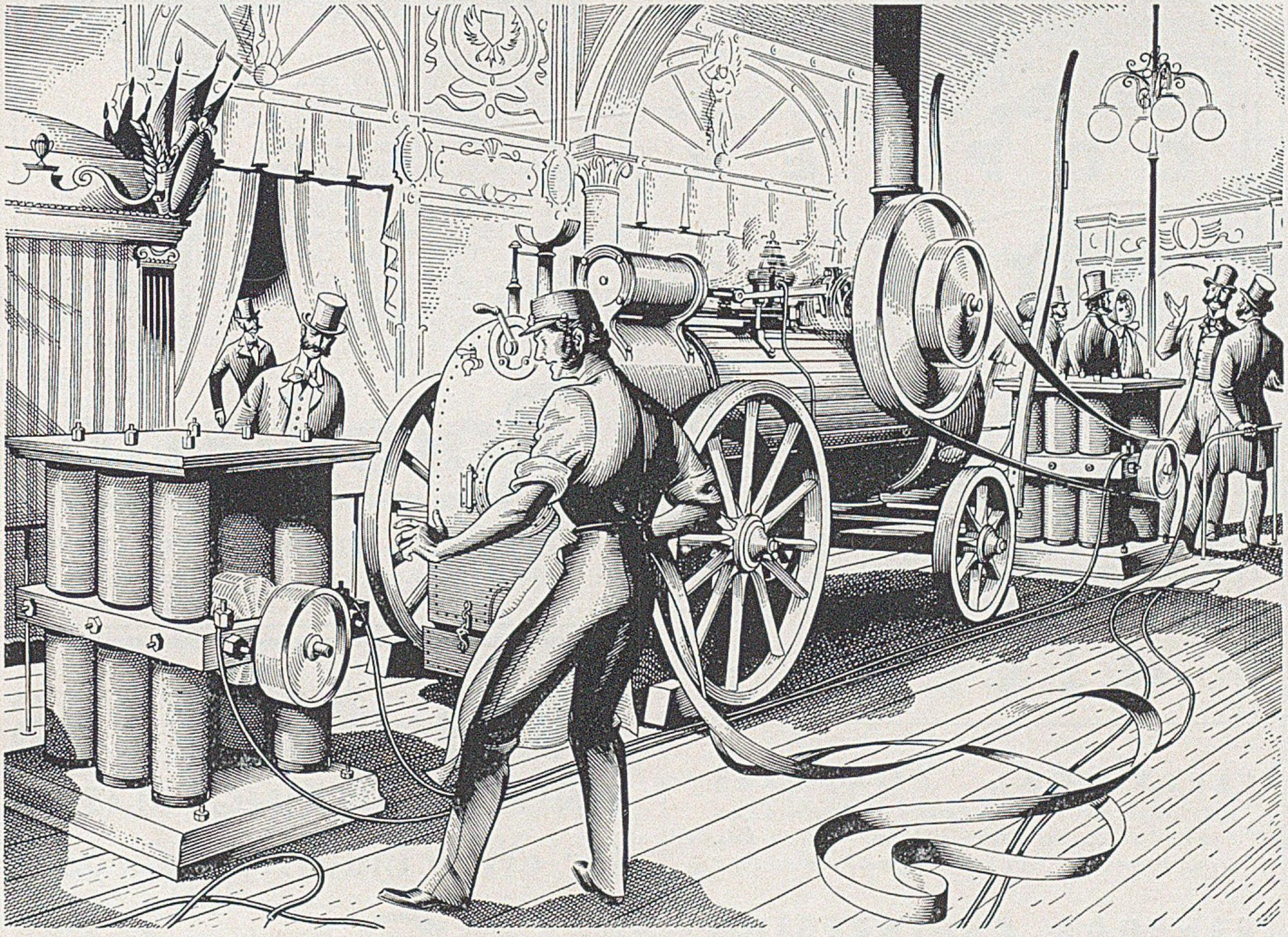
Most of the books on welding suffer from the disadvantage that they are written by a single author. In the past it may have been possible for an expert in the field to cover the entire matter satisfactorily, but nowadays the whole subject has become so complex that it is necessary to allocate the various chapters of a comprehensive book on welding to individual authors. The first book based on this idea is the present volume by Erdmann-Jesnitzer. It consists of two volumes, the first one of which deals mainly with ferrous metals and welding processes. There is, therefore, comparatively little scope for the reader who specializes in non-ferrous metals. Nevertheless any scientist or engineer whose work is connected with welding will find it invaluable as it contains a lot of unpublished work and ample literature references. Many of these references relate to Russian literature, which is unusual for most books on welding, and this may be a very attractive feature.

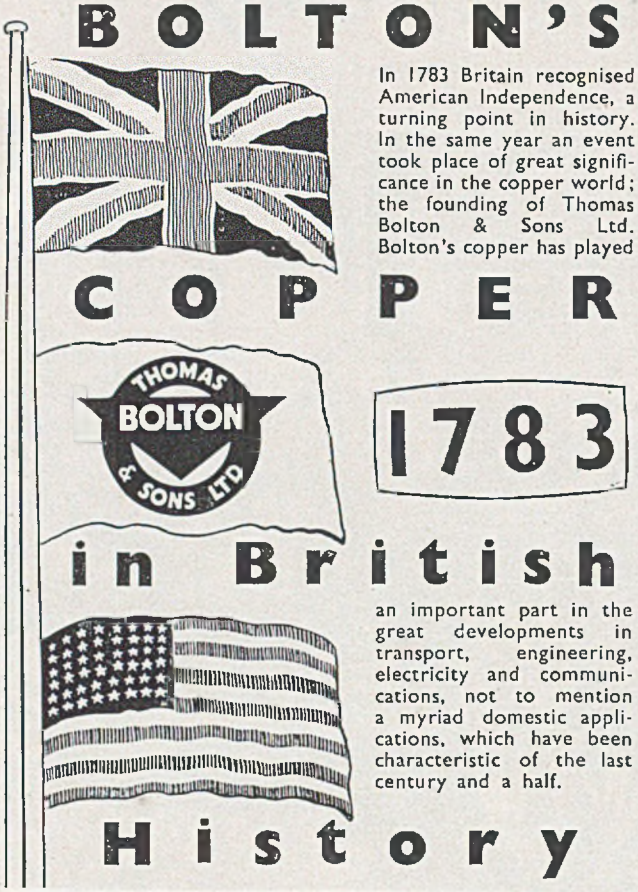
More than fifty experts have contributed to the book, and this partly accounts for the excellent presentation of every chapter. The co-ordination of such a standard book involves a great amount of work, and the editor notes, with German thoroughness, that the correspondence with the fifty-six contributors amounted to 8420 odd letters.

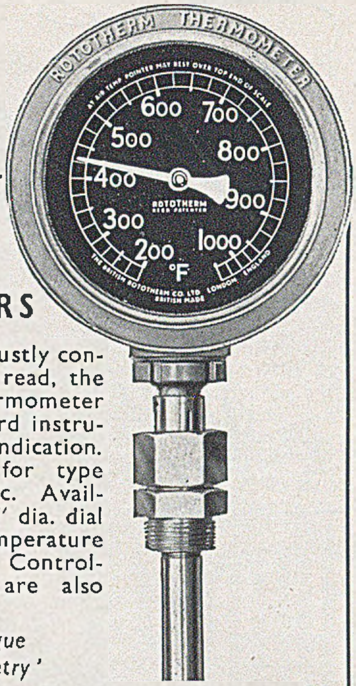
All graphs and schematic drawings are presented in such a way that they do not contain detailed information in German but only reference letters or figures, thus obviating the necessity for alterations of the blocks for any future translation.

It is regrettable that the indexes are to be contained in the second volume which has still to be published, and it is, therefore, rather difficult to find details without having to read through the whole chapter; it is suggested that the publishers should print a separate index which should be added to Volume I.

G. Haim.



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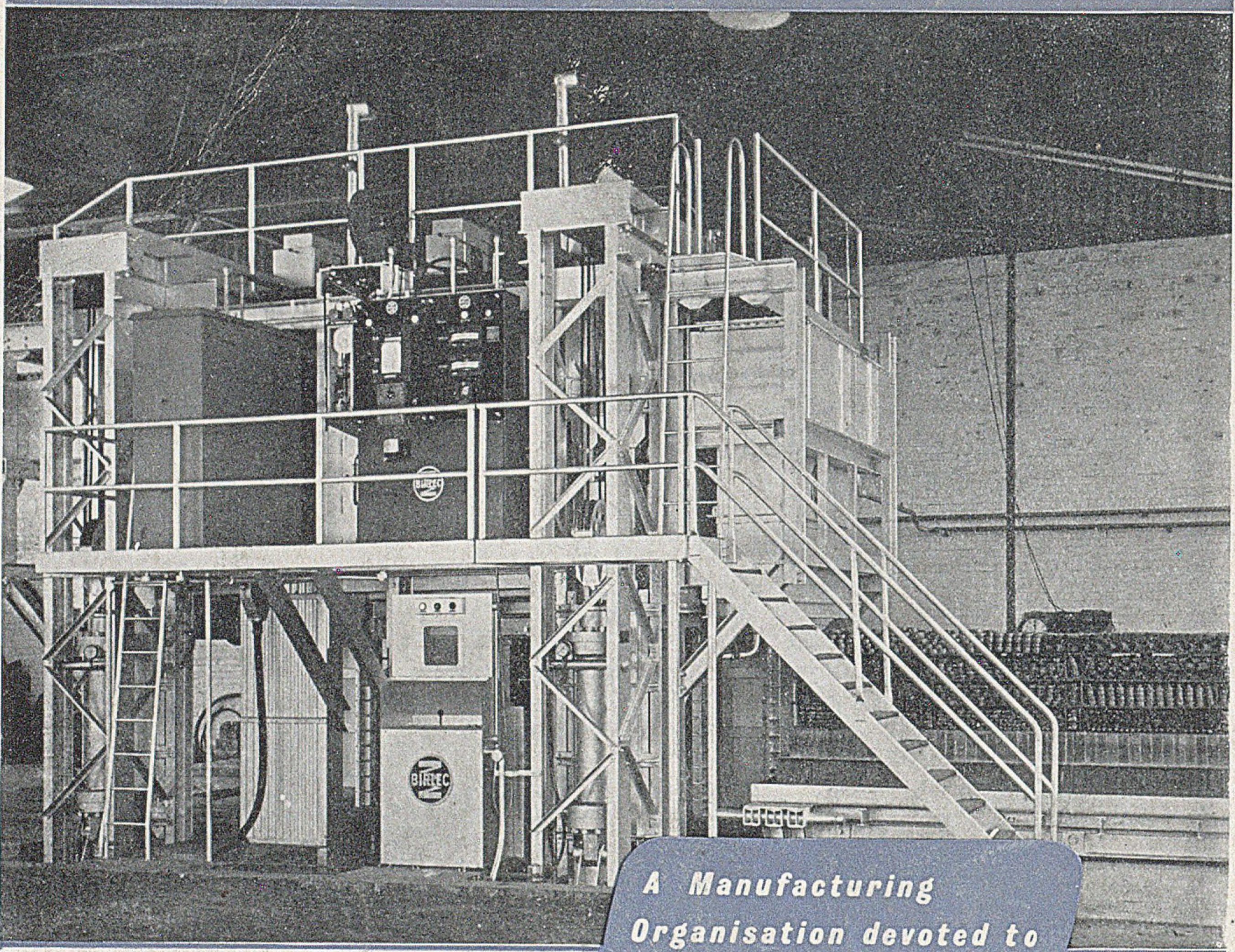

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