THE JOURNAL OF THE

INSTITUTE OF

Metals

WITH THE BULLETIN
AND METALLURGICAL ABSTRACTS

Principal Contents :

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DECEMBER 1951

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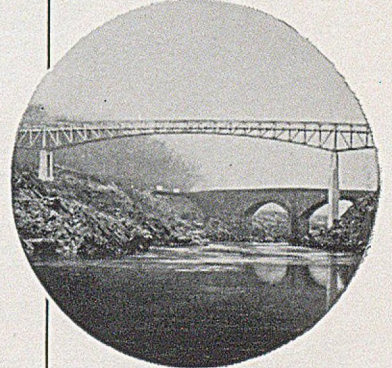
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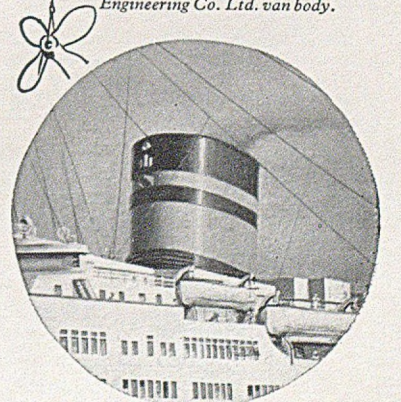
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N. B. VAUGHAN, M.Sc.

CONTENTS

BULLETIN

OF THE INSTITUTE OF METALS

INSTITUTE NEWS

The Autumn Meeting in Italy

About 250 members and their ladies (exclusive of those resident in Italy) took part in the recent Autumn Meeting which was held in Italy by invitation of the Associazione Italiana di Metallurgia. After four days in Venice, members proceeded to visit metallurgical works at Bolzano, Milan, Turin, and—from Florence—Fornaci di Barga. The meeting, a detailed programme of which was published on pp. 129-141 of the news section of the August issue of the monthly *Journal*, opened in the Doge's Palace at Venice, the opening ceremony being attended by the Minister of Education, the Mayor of Venice, the British Consul, the Presidents of the Industrial Associations of Marghera and Venice, the President of the Associazione Italiana di Metallurgia, many local notabilities, and over 500 members and ladies of the Associazione and the Institute.

In Venice, the scientific sessions of the Institute and of the Associazione were held, side-by-side, in the Ca' Foscari, on the Grand Canal. By the courtesy of their hosts, members received a publication containing, in English, summaries of the numerous papers presented to the Associazione, in addition to details of the meetings of both societies.

The meeting was one of the most successful that the Institute has held. On behalf of the Council and members, the President has expressed to the President of the Associazione di Metallurgia, and to all those who were concerned with the organization of the meeting, the Institute's warm sense of gratitude to its hosts for the most friendly and generous way in which the members and their ladies were received and entertained.

This meeting, it is hoped and believed, will have been the means of drawing more closely together those in Europe who are engaged in the science and practice of metallurgy. Old acquaintanceships were renewed and many new acquaintanceships and new friendships were made at this meeting, and members were privileged to see some very fine metallurgical plants in both the non-ferrous and ferrous industries.

A memorable meeting indeed, for which we offer our sincere thanks to our hosts and to Dr. Leno Matteoli, our Honorary Corresponding Member for Italy, who arranged it and played so large a part in its organization. We hope that, at a not too distant date, the Associazione Italiana di Metallurgia will be able to hold a meeting in Britain, side-by-side with a meeting of the Institute.

Membership Development

The Secretary has received instructions from the Council to take all possible steps to increase the Institute's membership in a big way during the present financial year. Unless specifically requested to do so by members of the Institute, he will not however invite possible members to join the Institute, but hopes to enlist the help of members themselves to introduce applicants for membership from their own Companies and laboratories or from among their own acquaintances in other Companies which, as yet, may not even be represented on the Institute's roll.

In many Companies in the non-ferrous metals industry there are no members at all; in others there are obviously many scientific and technical men who should be invited to join the Institute and to take part in its activities, and who would benefit from its publications and other services.

Will those members—both in the British Isles and in other countries—who are prepared to take steps to increase the Institute's membership in their own localities, or branches of the industry, please advise the Secretary to that effect? Copies of an attractive document giving brief details of the Institute's aims, work, and membership can be supplied, on application, and the Secretary will be glad, from his records, to give members the names of those who are already members of the Institute in the firms and other establishments in their own areas or branches of the industry.

With members' co-operation, it is believed that the membership of the Institute can be increased very largely in a short space of time.

Informal Discussion on " Tool and Die Materials for the Extrusion of Non-Ferrous Metals and Alloys "

As already briefly announced, a General Meeting of the Institute will be held at The University, Edgbaston, Birmingham 15, at 2.30 p.m. on Thursday, 3 January 1952, when there will be an informal discussion on "Tool and Die Materials for the Extrusion of Non-Ferrous Metals and Alloys". The Chair will be taken by Mr. Christopher Smith, Chairman of the Metallurgical Engineering Committee.

This is a new venture on the part of the Institute, and is designed to interest those who are actively concerned in production, both as engineers and metallurgists. If successful, it may lead to an extension of the Institute's activities which it is felt will be of considerable value to a wider range of members.

After short introductory talks by a user and by a manu-

facturer of tool and die materials, there will be a general discussion in which those engaged in extrusion operations will be encouraged to state their problems. It is hoped that representatives of extrusion-press makers and steel manufacturers will be present to take part in the discussion and to give advice where possible. Non-members who are interested in extrusion will be welcomed.

On the morning of 3 January the President (Professor A. J. Murphy, M.Sc.) has kindly arranged for the new Aitchison Laboratories of the Metallurgy Department at Birmingham University to be open for inspection by members at 10.30 a.m. Tickets of admission to the laboratories are obtainable from the Secretary of the Institute.

Lunch can be provided at the University (price 3s. 6d.), but those who wish to take advantage of this facility must notify the Secretary by 15 December, sending the necessary remittance.

" Non-Destructive Testing of Metals "

No. 10 in the Institute's Monograph and Report Series has now been published. It is entitled " Non-Destructive Testing of Metals ", and was specially written for the Series by Dr. R. F. Hanstock, Chief Physicist to High Duty Alloys, Ltd., Slough.

The published price is 21s. (or $3.50) post free, and—as a privilege of membership—each member of the Institute may purchase one copy at the reduced rate of 10s. 6d. (or $2.00) post free.

Representatives : General Board of the National Physical Laboratory

The Council has nominated Major C. J. P. Ball, D.S.O., M.C., F.R.Ae.S., Vice-President, to be a representative on the General Board of the National Physical Laboratory, in succession to Sir William Griffiths, D.Sc., F.R.I.C., F.I.M., whose term of duty ends on 31 December 1951.

The other representative of the Institute on the General Board is Professor A. J. Murphy, M.Sc., F.I.M.

Metal Physics Committee

Dr. C. S. Barrett has agreed to serve on the Metal Physics Committee while he is in this country.

Election of Members

The following 27 Ordinary Members, 2 Junior Members, and 9 Student Members were elected on 18 October 1951 :

As Ordinary Members

Ahmad, Halim Uddin, B.Sc., Assistant Sectional Officer, Royal Ordnance Factory, Radway Green, near Crewe, Cheshire.

Battin, Royston George, L. H. Dowling and Sons, Kingsbury Road, Minworth, near Birmingham.

Clark, Arthur, Works Manager, British Driver-Harris Company, Ltd., Cheadle Heath, Stockport, Cheshire.

Daniels, Neville Harold George, B.A., Senior Scientific Officer, Metallurgy Department, Royal Aircraft Establishment, Farnborough, Hants.

Darmara, Falih N., Ph.D., Assistant to the President, Utica Drop Forge and Tool Corporation, Utica, N.Y., U.S.A.

Davies, Daniel Evan, Metallurgist, Aluminium Wire and Cable Company, Ltd., Port Tennant, Swansea, Glam.

Doo, Betty, Senior Steward, Advanced Physical Metallurgy Laboratories, County Technical College, Wednesbury, Staffs.

Fisher, John C., A.B., Sc.D., Head, Physical Metallurgy and Cryogenics Section, General Electric Research Laboratory, The Knolls, Schenectady, N.Y., U.S.A.

Gawley, Thompson Leslie, B.Sc., Works Manager, Elkington and Company, Ltd., Goscote, near Walsall, Staffs.

Hameed, Muhammad Abdul, B.A., Assistant Sectional Officer, Royal Ordnance Factory, Radway Green, near Crewe, Cheshire.

Keefe, Joseph Marsden, Melting Shop Mechanical Engineer, Steel Company of Wales, Abbey Works, Port Talbot, Glam.

McKinnon, Neil A., M.Sc., Atomic Energy Research Establishment, Harwell, near Didcot, Berks.

Mehl, Ernst, Ph.D., Technical Adviser, McKechnie Brothers, Ltd., Widnes, Lancs.

Khand, Ghulam Mustafa, B.Sc., Assistant Sectional Officer, Royal Ordnance Factory, Radway Green, near Crewe, Cheshire.

Owen, William Edward, Chief Chemist, Richard Thomas and Baldwins, Ltd., Wern Works, Briton Ferry, Glam.

Parghi, H. S., B.Sc., Metallurgist, Kamani Metals and Alloys, Ltd., Agra Road, Kurla, Bombay, India.

Park, Alexander Gordon Paterson, Stewarts and Lloyds of South Africa, Ltd., P.O. Box 74, Vereeniging, Transvaal, South Africa.

Pienaar, Noel Pieter, B.Sc., Assistant Metallurgist, Stewarts and Lloyds of South Africa, Ltd., P.O. Box 74, Vereeniging, Transvaal, South Africa.

Rang, Edmund Jule, M.Sc., Head of Boiler Design Department, Merz and McLellan, Consulting Engineers, Carliol House, Newcastle-on-Tyne 1.

Schapiro, Leo, B.S., Ph.D., Chief Metallurgist, Douglas Aircraft Company, Inc., Santa Monica, Calif., U.S.A.

Sendorek, Andrew, Dipl.Ing., Metallurgist, Research Laboratories, The British Aluminium Company, Ltd., Chalfont Park, Gerrards Cross, Bucks.

Sperring, Elbert D., Assistant to Continuous Strip Mill Superintendent, Scovill Manufacturing Company, Waterbury, Conn., U.S.A.

Spooner, Edwin George, Director, Martin Burn, Ltd., 12 Mission Row, Calcutta, India.

Sullivan, John W. W., Sc.D., American Iron and Steel Institute, 350 Fifth Avenue, New York 1, N.Y., U.S.A.

Thacker, M. S., B.Sc., Director, Indian Institute of Science, Bangalore 3, India.

Werdenberg, Wilhelm, Dipl.Ing., Director, S.A. des Câbleries et Tréfileries de Cossonay-Gare, Switzerland.

Wright, Eleanor V., A.B., Engineering Librarian, Chrysler Corporation, P.O. Box 1118, Detroit 31, Mich., U.S.A.

As Junior Members

Dodd, Peter Graham, Head of Works Laboratory, Metal and Ores Pty., Ltd., 116 Bourke Road, Alexandria, N.S.W., Australia.

Wilkinson, Henry, B.Sc., Metallurgist, Northern Aluminium Company, Ltd., Banbury, Oxon.

As Student Members

Banerji, Salil Kumar, B.Sc., National Metallurgical Laboratory, Jamshedpur, India.

Bruce, Douglas Swirney, B.Sc., A Squadron, 3 Training Regiment, Royal Engineers, Cove, Hants.

A Group of Members and Ladies, photographed on the Piazzale Michelangelo, Florence, during the Autumn Meeting in Italy.   The statue is a copy of Michelangelo's "David".

CHAPPELL, John Brian, Metallurgist, Richard Thomas and Baldwins, Ltd., Section Research Department, Wern Works, Briton Ferry, Glam.

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

HOWARD, WILLIAM James, Mechanical Testing Officer, Australian Aluminium Company Pty., Ltd., P.O. Box 12, Granville, Sydney, N.S.W., Australia.

KHAN, Mohammed Idrees, Student of Engineering, J. I. Thornycroft and Company, Ltd., Woolston, Southampton.

LATIMER, Keith Graham, Student of Metallurgy, Battersea Polytechnic, London, S.W.11.

STILES, Denis, Student of Metallurgy, British Non-Ferrous Metals Research Association, Euston Street, London, N.W.1.

WOOD, Derek Statham, B.Sc., Research Student, Department of Metallurgy, The University, Manchester 13.

PERSONALITIES

Mr. N. I. Bond-Williams

Mr. Noel Ignace Bond-Williams, who has recently been appointed to fill a casual vacancy on the Council of the Institute, was born in 1914 and educated at West House School and Oundle. He entered Birmingham University in 1932 and studied metallurgy under Professor Hanson, obtaining his B.Sc. degree with Honours in 1935. For a period he carried out research at the University, but abandoned it in 1937 on the death of his father, Mr. W. H. Williams, whom he succeeded as Managing Director of The Aston Chain and Hook Co., Ltd., Birmingham.
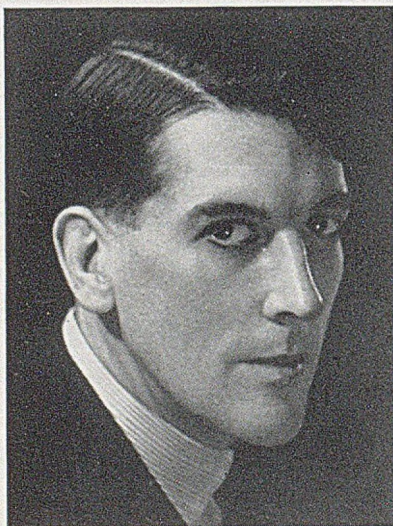
Mr. Bond-Williams was President of the Guild of Undergraduates of Birmingham University in 1936-37, and President of the Guild of Graduates in 1947-50. He is a member of the American Society for Metals, and the American Institute of Mining and Metallurgical Engineers, and an Associate of the Institution of Metallurgists. He also serves on the Council of the Industrial Welfare Society.

PERSONAL NOTES

MR. R. W. BAILEY, of the Zinc Development Association, was a member of an O.E.E.C. Technical Mission which has just visited the United States to study galvanizing techniques.

MR. T. G. BAMFORD is retiring at the end of the year from the position of Principal of the County Technical College, Wednesbury, which he has held for the last 23 years.

SIR LAWRENCE BRAGG has been awarded the Reginald Mitchell Memorial Gold Medal of the Stoke-on-Trent Association of Engineers.

MR. D. H. BUTLER has left Copper and Alloys, Ltd., West Bromwich, and taken a post as Development Metallurgist to the Phosphor Bronze Co., Ltd., Bradford Street, Birmingham.

MR. R. A. BUTLER has taken up an appointment as metallurgist at the Rugby works of the English Electric Co., Ltd.

DR. A. T. CHURCHMAN has resigned his fellowship at Birmingham University and is now a research metallurgist in the laboratory of Associated Electrical Industries, Ltd., Aldermaston, Berks.

MR. L. M. CLAREBROUGH, a Research Officer in the Division of Tribophysics of the C.S.I.R.O., Melbourne, is working for a year at Birmingham University.

MR. C. L. M. COWLEY, of Imperial Chemical Industries, Ltd., Metals Division, has just visited the United States as a member of an O.E.E.C. Technical Mission which studied non-ferrous heavy metal fabrication.

MR. C. F. DAVEY, of Thomas Bolton and Sons, Ltd., has just visited the United States as a member of an O.E.E.C. Technical Mission which studied non-ferrous heavy metal fabrication.

MR. C. H. DAVY has been appointed Works Director at the Renfrew and Dumbarton Works of Babcock and Wilcox, Ltd.

MR. E. GREGORY has now left Cambridge University, where he has been engaged on research for the last three years, having been awarded an E.C.A. Scholarship for the study of production technology and management in the field of high-temperature materials. His address is 933 Forest Avenue, Ann Arbor, Mich.

MR. V. GRIFFITHS has completed his post-graduate research at University College, Swansea, and is now a half-time assistant in the Department of Metallurgy at the Massachusetts Institute of Technology.

MR. K. W. GRIMWOOD has recently obtained the degree of B.Sc.(Eng.) at London University.

MR. F. HARGREAVES has been appointed Regional Works Metallurgist, Southern Region, British Railways, with headquarters at Brighton.

MR. R. HAYNES has recently left the University of Sheffield and has been appointed assistant lecturer in metallurgy in the University of Leeds.

MR. H. HERNE has left the British Iron and Steel Research Association and joined the Research Association of British Rubber Manufacturers, Croydon.

MR. J. E. HILLIARD has taken a temporary appointment as Research Associate in the Metallurgy Department of the Massachusetts Institute of Technology, Cambridge, Mass.

MR. E. J. HOOKER has left the Research Laboratories of British Insulated Callender's Cables, Ltd., and has been appointed metallurgist at Southern United Telephone Cables, Ltd.

MR. H. A. MACCOLL, at present Head of the Department of Metallurgy, has been appointed Principal of the County Technical College, Wednesbury, in succession to Mr. T. G. Bamford.

DR. H. L. MAXWELL, Supervisor of General Consultants, E.I. du Pont de Nemours and Co., Inc., Wilmington, Del., has been elected a Vice-President of the American Society for Testing Materials.

MR. H. W. MEAD is now working in the Metal Research Laboratory of the Carnegie Institute of Technology, Pittsburgh, Pa.

MR. H. J. MILLER, of British Insulated Callender's Cables, Ltd., has just visited the United States as a member of an O.E.E.C. Technical Mission which studied non-ferrous heavy metal fabrication.

MR. J. MORGAN, formerly Chief Metallurgist to Magnal Products, Ltd., has left for Canada to take up a post as Chief Metallurgist to Foundry Services (Canada), Ltd., Guelph, Ont.

MR. R. PRASAD is now a research student in the Department of Metallurgy at Cambridge University.

MR. C. W. RIDGE has recently returned to Southern Rhodesia, where he has been appointed consultant to several firms.

MR. N. H. TOWNEND has left Thomas Bolton and Sons, Ltd., Widnes, to join the staff of the Research Laboratories of the General Electric Co., Ltd., Wembley.

MR. L. WALKER has left the English Electric Co., Ltd., and is now a development metallurgist in the laboratories of Plessey Co., Ltd., Towcester, Northants.

Obituary

The Editor regrets to announce the death of:

MR. GEORGE CHELIOTI, a Director of the General Electric Co., Ltd., and Associated Companies, and General Manager of the Osram–General Electric Company group of works, who died on 27 September at sea while returning from South Africa.

JOHN DEVONSHIRE ELLIS at his home at Lavant, Sussex, on 31 August, in his 68th year. Until his retirement last year, Mr. Ellis had been since 1929 London Manager of The Birmingham Battery and Metal Co., Ltd.

MR. ARTHUR WESLEY HOTHERSALL, M.Sc., F.I.M., Deputy Chief Scientific Officer, Armament Research Establishment, Woolwich, at his home at Eltham on 20 October, aged 55.

MR. FRANK WALKER, a metallurgist with British Insulated Callender's Cables, Ltd., Liverpool, at the age of 28. Mr. Walker was one of the victims of the accident to the Liverpool–London express which occurred at Weedon, Northants, on 21 September.

NEWS OF LOCAL SECTIONS AND ASSOCIATED SOCIETIES

Birmingham Local Section

The Birmingham Local Section is again following its practice in recent years by holding an All-Day Meeting in the Spring of 1952. This will take place on Friday, 29 February, at the College of Technology, Suffolk Street, Birmingham, and will be concerned with " New Techniques of Metallurgical

Research ". Papers have been promised on a variety of aspects of metallographic and X-ray technique, on surface examination of metals, on mechanical testing, and on the contribution which powder metallurgy is able to make to metallurgical research. An Introductory Address will be given by Professor A. G. Quarrell and, in addition to discussions on the individual papers, there will be an Open Discussion at the end of the afternoon session. Short contributions to this (" Supplementary Communications ") will be welcome and will be circulated, with the synopses of the papers, to those who register for the meeting. (Registration fee 7s. 6d.; lunch 8s. 6d.) Communications and enquiries should be addressed to the Hon. Secretary, Mr. E. H. Bucknall, M.Sc., F.I.M., 264 Harborne Park Road, Harborne, Birmingham 17.

The Debate: " That in the opinion of this House further encroachment of Metal Physics into the sources of instruction would be detrimental to training in Metallurgy," originally arranged for 25 October, has been postponed until 7 February 1952.

All meetings of the Section this session will begin at 6.30 p.m., and not at 7 p.m. as stated in the programme.

South Wales Local Section

The President, accompanied by the Secretary, will pay an official visit to the South Wales Local Section on Tuesday, 15 January 1952.

JOINT ACTIVITIES

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee announces the following awards for 1951:

MR. J. PRESTON (British Non-Ferrous Metals Research Association) to study, in Great Britain, the United States, and Canada, specialized methods of production and fabrication of metals, with particular reference to powder-metallurgical techniques.

MR. P. E. WHITE (J. B. and S. Lees, Ltd.) to study the metallurgy and detailed production technique of high-quality strip steels in the United Kingdom, Scandinavia, the United States, and Canada, with particular reference to hardened and tempered steel strip, stainless steels, and silicon steels.

NEWS OF KINDRED SOCIETIES

The Institution of Metallurgists

The following candidates have been successful in the 1951 examinations:

Fellowship

G. R. Morton (Coventry).

Associateship

★ R. J. Beishon (Wembley), ★ D. Bindley (Allesley), K. G. Blackmore (Liverpool), P. Bradley (Aylesbury), ★ H. Brunner (London), D. Cemm (Lewisham), J. Cowie (Kilbirnie), ★ A. R. Daniel (Morden), D. Davies (Birmingham), J. Donaldson (Glasgow), C. Ford (Glasgow), J. F. Giblin (St. Helens), C. C. Hanson (Walsall), J. Hines (Nuneaton), S. R. A. Langford (Brockley), R. C. Lloyd (Birmingham), R. M. Loach (Bilston), ★ P. H. Meade (Wirral), R. Mills

(Streatham Common), * R. C. Perriton (Worcester Park), M. W. Randall (Coventry), * R. W. Shapton (Danygraig), N. F. Sheppard (Hornsey), E. Ward (Doncaster), T. Williams (Ruddington).

* Election to the appropriate grade of membership deferred until regulations in respect of age and experience shall have been complied with.

Licentiateship

R. G. Bretherton (Croston), N. T. Burgess (Merstham), J. W. Campbell (London), W. H. Casling (Heston), J. D. Chadwick (Clapham Common), D. J. Fricker (Slough), P. Gregory (Farnborough), E. G. Hall (Slough), J. R. Hawley (Leicester), A. Milligan (Rosyth), R. K. Mitra (India), C. Morris (Stoke Gifford), R. J. Page (Uxbridge), E. C. Sykes (Palmers Green), P. C. Thornton (Birmingham), W. R. Utteridge (Coventry).

Institute of Metal Finishing

The Council of the Electrodepositors' Technical Society announces that, in accordance with a Resolution passed at the last Annual General Meeting, application which was made some time ago for official incorporation has now been granted and as a result the Society is now an Incorporated Body.

At the same time application was made for permission to adopt a new title embracing metal finishing generally (excluding only vitreous enamelling). This is in accordance with the general policy of the Society to extend its field beyond electrodeposition processes to include metal finishing generally, with which the bulk of its members are vitally concerned.

The new title by which the Society will henceforth be known is The Institute of Metal Finishing incorporating Electrodepositors' Technical Society Limited. As a result of this change the original aims and objects of the Society—namely, to promote the study of, and disseminate information on, electrodeposition and cognate processes as widely as possible— are in no way affected and will, in fact, be pursued as vigorously as ever. The sphere of interest has now been significantly widened, and the Council is of the opinion that this will bring with it corresponding benefit to the membership.

The Executive Officers and Council of the new Institute remain unchanged. The present Executive Officers are: Mr. H. Silman, President; Dr. S. Wernick, Honorary Secretary; and Mr. F. L. James, Honorary Treasurer. The Institute's address is 27 Islington High Street, London, N.1 (TERminus 3251).

A.S.T.M. Symposium on Testing Metal Powders and Metal-Powder Products

A Symposium on Testing Metal Powders and Metal-Powder Products, consisting of several technical papers by leading authorities in this field, will be an important feature of the 1952 Spring Meeting of the American Society for Testing Materials, which will be held during the week beginning March 3 at the Hotel Statler in Cleveland, Ohio.

While many symposia have been held on the scientific and technical aspects of powder metallurgy, this is believed to be the first symposium which is specifically concerned with methods of testing. Powder metallurgy as a production method has made rapid progress during the last few years, and new and improved methods for testing powders and fabricated parts are constantly being evolved. The proposed symposium is intended to serve as a medium for the exchange of information between all who are interested in this field, and papers which are concerned with any phase of the testing of metal powders or of fabricated parts are invited. Anyone interested in contributing such a paper should write to Mr. F. V. Lenel, Rensselaer Polytechnic Institute, Troy, N.Y., who is the Secretary of A.S.T.M. Committee B-9, which is organizing the symposium.

OTHER NEWS

The Royal School of Mines

We offer our warmest congratulations to the Royal School of Mines on its Centenary, which it celebrated last month. There can be few members of the Institute who are not aware of the important part played by the School during the past hundred years, in particular in the fields of geology, mining, and extractive metallurgy, for its graduates have been found at work in the remotest parts of the world and have established a reputation for ability and integrity that reflects the greatest credit on the Royal School of Mines and the line of distinguished men who have directed the studies there.

It is now just 100 years ago that the Prince Consort actively assisted in the formation of the "Government School of Mines and of Science Applied to the Arts". This school started as an appendage to the Museum of Practical Geology, which had been formed by the efforts of Sir Henry De la Beche. The Museum and School were situated in Jermyn Street, London, on the site now occupied by Simpsons.

In 1853 the school was placed under the control of the Science and Art Department and the first staff included Lyon Playfair, Forbes, Hunt, Percy, A. C. Ramsay, and Warington Smyth. Later, Murchison succeeded De la Beche as Director and T. H. Huxley succeeded Forbes. In 1859 the name was altered to "The Government School of Mines" and four years later to "The Royal School of Mines".

In 1870 a Royal Commission advised the removal of the Royal School of Mines to a site near Exhibition Road in South Kensington, which Prince Albert had persuaded the Commissioners of the 1851 Exhibition to purchase, and during the next few years, in spite of stern opposition from Percy, this move took place. As a result Percy resigned and was succeeded by a former student, Chandler Roberts, later to be Sir William Roberts-Austen, who combined this post with that of Chemist and Assayer at the Royal Mint. During the next decade the staff included such men as Brough, Judd, Thorpe, and le Neve Foster. In 1902 Sir William Gowland, another of Percy's pupils, succeeded Roberts-Austen as Professor of Metallurgy, and the succession proceeded through Carlyle and Sir Harold Carpenter to the present holder of the chair, C. W. Dannatt. The courses in mining and metallurgy were in 1911 extended to four years, and later courses in oil technology and mining geology were included in the curriculum. Other professors in the School at present are J. A. S. Ritson (mining), David Williams (mining geology), and V. C. Illing (oil technology).

The Central Technical College of the City and Guilds Institute, founded in 1884, was already housed in Exhibition Road, and in 1907 a Charter established the Imperial College of Science and Technology, comprising the Royal College of Science, the Royal School of Mines, and the City and Guilds Engineering College. The present building in Prince Consort Road was occupied in 1913.

The Centenary celebrations have included a banquet at the Drapers' Hall attended by the Chancellor of London Univer-

sity, the Earl of Athlone, and H. R. H. Princess Alice; a Conversazione in the Royal School of Mines building; and a Commemoration Ball.

Leverhulme Research Fellowships, 1952

Application is invited for Fellowships and Grants in aid of research. The Fellowships and Grants are intended for senior workers who are prevented by routine duties or pressure of other work from carrying out research. They are limited to British-born subjects normally resident in the United Kingdom. In exceptional circumstances the Trustees may waive the condition as to residence.

The Trustees are also prepared to consider applications from groups of workers engaged upon co-operative programmes of research particularly from those engaged upon long-distance programmes.

The duration of the awards will not normally extend over more than two years or less than three months and the amount will depend on the nature of the research and the circumstances of the applicant.

Forms of application may be obtained from the Secretary, Leverhulme Research Fellowships, 3/5, Salisbury Square, London, E.C.4. Telephone: City 1910.

Applicants must be received on or before 31 December 1951. Awards will be announced in May and will date from 1 September 1952.

DIARY

Local Section Meetings

4 December. **Oxford.** Discussion on Selected Papers Published in the *Journal*. (7.0 p.m.)

4 December. **South Wales.** "The Solidification of Castings", by R. W. Ruddle. (University College, Metallurgy Department, Singleton Park, Swansea, at 6.30 p.m.)

6 December. **Birmingham.** Student Evening. (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

7 December. **Sheffield.** "Precious Metals in Industry", by H. G. Dale. Joint Meeting with the Sheffield Centre of the Institute of Metal Finishing. (The University, St. George's Square, Sheffield, at 6.30 p.m.)

10 December. **Scottish.** Symposium on "Materials Testing". (39, Elmbank Crescent, Glasgow, C.2, at 6.30 p.m.)

12 December. **London.** "The Development of Aluminium Alloy Casting", by A. P. Fenn. Joint Meeting with the Institute of British Foundrymen, London Branch. (Waldorf Hotel, Aldwych, W.C.2, at 7.0 p.m.)

Other Societies

4 December. **Institute of Metal Finishing, Midlands Centre.** "Paint Application by Automatic Means", by Dr. S. Wernick. (James Watt Memorial Institute, Great Charles Street, Birmingham 3, at 6.30 p.m.)

5 December. **Institution of Heating and Ventilating Engineers, East Midlands Branch.** "Internal Stresses in Metals", by Professor J. A. Pope. (School of Technology, Leicester, at 6.30 p.m.)

5 December. **Joint Committee on Materials and Their Testing.** Symposium on "Recent Developments in the Notch-Bar Testing of Materials and Their Relation to Welded Construction". (Institution of Civil Engineers, Great George Street, London, S.W.1, at 10 a.m. and 2.30 p.m.)

5 December. **Manchester Metallurgical Society.** Film Show. (Engineers' Club, Albert Square, Manchester, at 6.30 p.m.)

6 December. **Institution of Production Engineers, London Section.** "High-Speed Press Work", by J. A. Grainger. (Royal Empire Society, Northumberland Avenue, London, W.C.2, at 7.0 p.m.)

6 December. **Leeds Metallurgical Society.** Films on Process Heating Equipment. (Chemistry Department, The University, Leeds 2, at 7 p.m.)

APPOINTMENTS VACANT

ANALYTICAL CHEMIST assistant required by London firm engaged in manufacturing non-ferrous metal alloys. The job holds good prospects for the right man. Box No. 314, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

BIRLEC LIMITED, Tyburn Road, Erdington, Birmingham 24, require a trainee Service Engineer for starting up, servicing, and field research on large arc and induction melting furnaces, ferrous and non-ferrous. Qualifications are H.N.C. or equivalent in electrical or mechanical engineering and practical experience in foundry work. The position offers excellent prospects, with opportunities for travelling abroad. Applicants should give full details of their education and experience and state the salary required. Apply to Personnel Officer.

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 for investigations on non-ferrous alloys in Research Division of Laboratory attached to Founders in the London Area. Applicants should have a good degree in metallurgy or equivalent qualifications, or alternatively, a degree in physics or

chemistry with metallurgical experience. Salary according to qualifications and experience. Write to Box No. 316, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

PHYSICIST, Physical Chemist or Metallurgist required by research laboratory (London 50 miles) to study fundamental aspects of high-temperature surface reactions of the rarer metals (titanium, zirconium, &c.). Ph.D. or equivalent research experience essential. Salary up to £900 p.a. Write giving details of age, qualifications, and experience to Box No. 315, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

SENIOR METALLURGIST required to take general charge of research on bearing metals, bronze and other alloys containing tin. Industrial experience an advantage. The post is one of responsibility and will be suitably remunerated. F.S.S.U. superannuation. Applications to be sent to The Director, Tin Research Institute, Fraser Road, Perivale, Greenford, Middlesex.

YOUNG METALLURGIST with energy and ingenuity required for interesting work on the development and production of welding electrodes. Degree, L.I.M., or equivalent. Some knowledge of electrical and mechanical engineering an advantage, but previous experience not essential. Very good prospects for man with the necessary lively outlook. South London area. Salary up to £650 per annum. Please reply stating age, qualifications, and experience to Box No. 317, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

THERMOELASTIC ANALYSIS OF TRANSFORMATIONS IN COPPER ALLOYS *

By R. CABARAT,† P. GENCE,‡ Professor L. GUILLET,§ and R. Le ROUX ‖

SYNOPSIS

The purpose of the investigation was to demonstrate the effect of allotropic transformations on the elastic modulus and logarithmic decrement of copper alloys. To determine these two properties a specimen was subjected to forced longitudinal vibrations of small amplitude and high frequency under reduced pressure. The formation of body-centred cubic phases on heating eutectoid copper–aluminium and copper–tin alloys results in a marked decrease in the elastic modulus, while the logarithmic decrement assumes large values before and during the transformation. The results are discussed in the light of present knowledge of the structure of alloys.

I.—INTRODUCTION

THERMOELASTIC analysis is one of the physico-thermal methods that are often used to reveal transformations in alloys through the changes taking place in physical properties with temperature. Sudden changes in these properties occur at certain temperatures, corresponding to changes in lattice structure, the appearance or disappearance of a phase, &c. Physico-thermal methods have the advantage that their sensitivity is independent of the rate of heating and cooling.

Among the properties most frequently studied in this way are the coefficient of thermal expansion, electrical resistivity, thermoelectric power, specific heat, &c. The modulus of elasticity has been used to reveal anomalies in ferro-nickels,[1] cobalt,[2] and Monel metal,[3] allotropic transformations in steels [4, 5] and non-ferrous alloys,[6, 7] order–disorder transformations in gold–copper and copper–platinum alloys,[7, 8] as well as recrystallization phenomena in cold-worked metals.[9, 10] Variations in logarithmic decrement have also been determined in the case of order–disorder transformations in gold–copper and copper–platinum alloys and allotropic transformations in silver–zinc and silver–cadmium alloys.[7]

In order to measure the elastic properties accurately at high temperatures, viscous relaxation phenomena must be eliminated as far as possible, and dynamic methods are therefore particularly suitable. The present authors have already described [11] a new apparatus in which the specimen is subjected to forced longitudinal vibrations by means of an electrostatic field. This apparatus has since been improved so that it may be used at high temperatures, and the purpose of the present paper is to demonstrate its suitability for revealing transformations in alloys. Copper–tin and copper–aluminium alloys were chosen for the work because their equilibrium diagrams are now well established; in addition, the transformation temperatures are not very high, so that there are no experimental difficulties. A slow rate of heating (1° C./min.) is necessary, to afford time to make the temperature measurements with the desired precision.

II.—EXPERIMENTAL METHOD

The Cabarat elasticimeter depends on exciting longitudinal vibrations in a specimen electrostatically. The specimen, about 150–250 mm. long and 8–12 mm. in dia., is held in position by three hardened steel needles, between two adjustable electrodes A and B (Fig. 1).

A voltage amplified from a low-frequency oscillator is applied to electrode A and exerts a periodic electrostatic attraction on the specimen, thereby exciting longitudinal vibrations in it. Electrode B acts as an electrostatic microphone; the voltages produced in it by movements of the specimen are recorded after amplification on a cathode-ray oscilloscope, and can be measured by means of a valve voltmeter. The specimen is at earth potential. The amplitude of vibration is observable only if the frequency of the electrical vibrations coincides with the natural frequency of the specimen. Resonance is obtained by varying the frequency of the voltage between the electrode and earth by means of a variable condenser at the oscillator; the amplitude of the vibrations recorded on the oscillograph then passes through a very sharp maximum. The frequency read from the dial of the low-frequency oscillator (previously calibrated) is equal to the natural frequency (F_0) of the specimen.
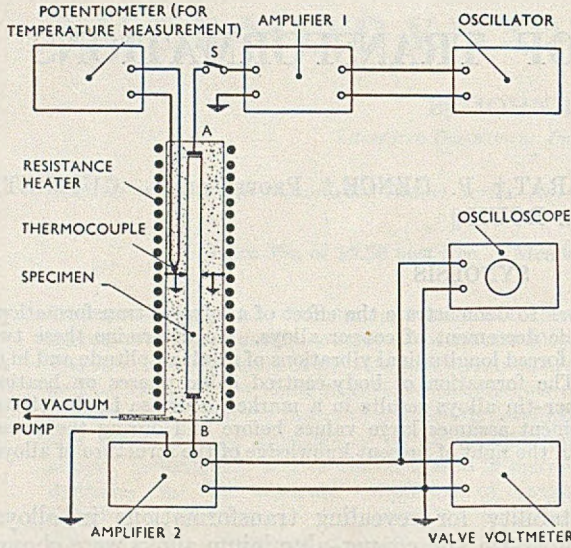
Fig. 1.—Schematic Diagram of the Apparatus.

If the density of the specimen is d and its length L, the elastic modulus (E) is given by :

$$E = (2F_0 L)^2 d \quad . \quad . \quad . \quad . \quad (1)$$

The internal friction, represented by the logarithmic decrement (δ) of the vibrations, can be obtained from the resonance curve by the formula :

$$\delta = \pi \Delta F / F_0 \quad . \quad . \quad . \quad (2)$$

ΔF being the frequency interval separating two points on the resonance curve at which the amplitude is reduced to $A_0/\sqrt{2}$ (A_0 = peak amplitude).*

The curve showing the exponential decay of the amplitude of vibration with time can also be recorded. The excitation of the specimen is cut off by the switch S, while leaving the microphone in action. To avoid errors due to the detection being no longer uniform at low microphone voltages, an amplifier having a logarithmic response is used. A straight line is thus obtained from which the value of the decrement can be calculated (Fig. 2). If t_s is the time in which the
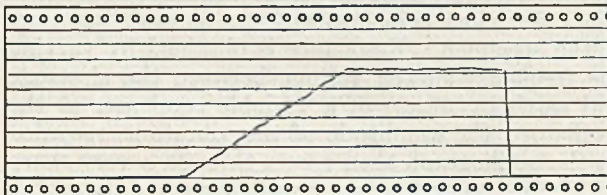
Fig. 2.—Oscillograph Record of Decay of Vibrations.

initial amplitude (A_0) of vibration of the specimen is reduced to $A_0/1000$, then :

$$\delta = \frac{\log_e 1000}{F_0 \times t_s} = \frac{6 \cdot 9}{F_0 \times t_s} . \quad . \quad . \quad (3)$$

These two methods complement one another perfectly. For a material of high internal friction the resonance curve can conveniently be used ; for a material with a very small damping capacity, however, the resonance curve is very sharp and the maximum amplitude is difficult to determine. In consequence ΔF cannot be obtained accurately.

Acoustic radiation from the ends of the specimen can prove a serious source of error in measurements of internal friction, and for this reason the equipment for supporting the specimen and the electrodes is placed in a sealed tube in which the pressure is of the order of 0·1 mm. Hg. A systematic study of the damping of a steel specimen (250 mm. long × 10 mm. dia.) as a function of the degree of vacuum showed that the logarithmic decrement of the vibrations became constant at pressures below 1 mm. Hg (Fig. 3). The damping times corresponding to a reduction of the initial amplitude A_0 to a value $A_0/1000$ are reproducible to within about 5%.

This method has the advantage of being convenient for the study of the changes of elastic modulus and internal friction with temperature (θ) ; all that is necessary is to place the apparatus previously
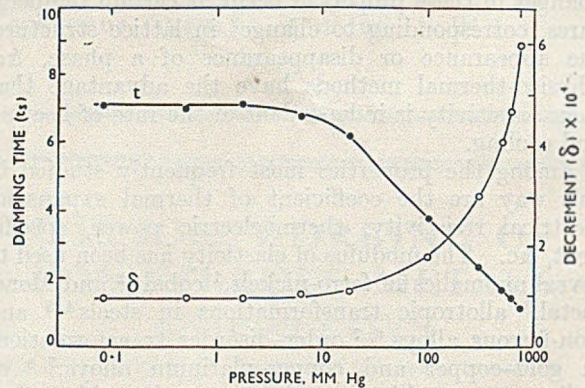
Fig. 3.—Time of Decay of Vibrations and Logarithmic Decrement of Steel Specimen as a Function of Pressure.

described in an electric furnace. To allow for the alteration in density and length of the specimen caused by the rise in temperature, a correction must be applied to the value of the modulus given by equation (1). If the density and length are measured at 20° C. and if α is the mean coefficient of expansion of the metal between 20° C. and θ, then

$$E_{\text{corr.}} = \frac{E_{\text{obs.}}}{1 + \alpha (\theta - 20)} . \quad . \quad . \quad (4)$$

Chevenard's apparatus is used to obtain α.

Fig. 4 shows the Cabarat elasticimeter, which is in the form of a desk. In the centre is the furnace D in which is situated the device for holding the specimen and the electrodes. Above the desk, on the left, is the cathode-ray oscilloscope C by means of which the electrodes are controlled, and the resonance of the specimen is observed when the condenser of the low-
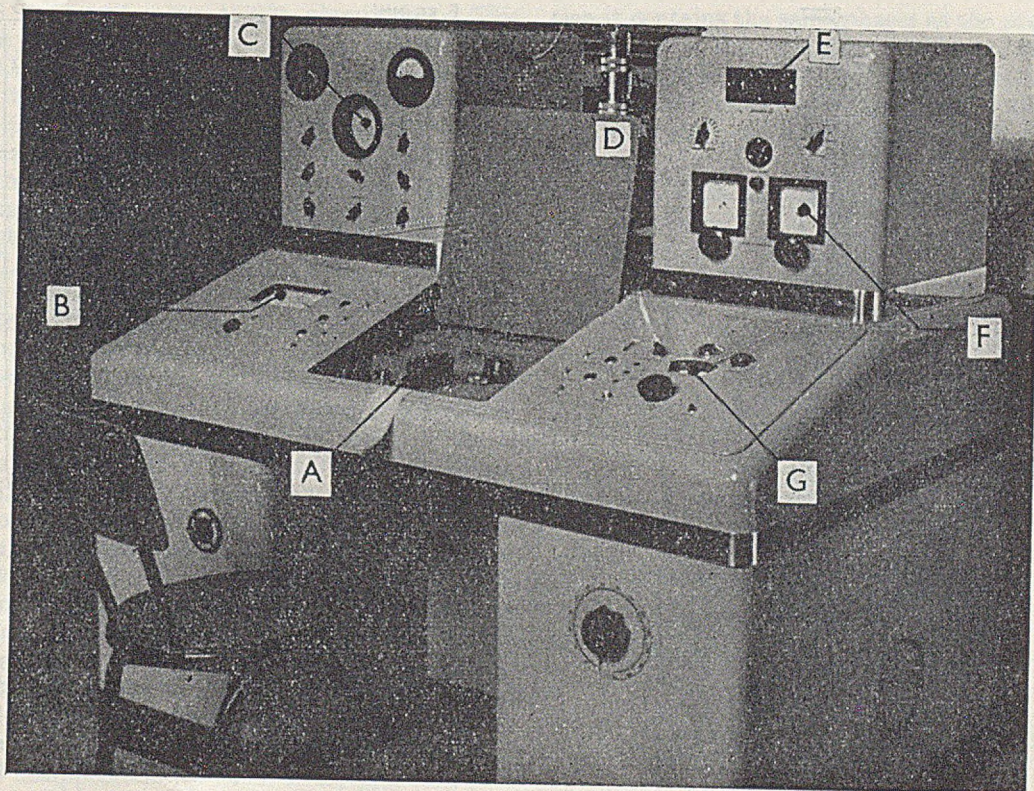
FIG. 4.—General View of the Apparatus.

frequency oscillator F on the right of the furnace is varied. On the desk itself, from left to right, are mounted the valve voltmeter B, the logarithmic recorder A, and the potentiometer G for measuring the temperature inside the furnace. E is the vacuum gauge.

With this apparatus the frequency can be measured to ± 1 cycle in the range 5000—15,000 cycles/sec. To obtain this accuracy the low-frequency oscillator is calibrated against a tuning-fork having a frequency of 1000 cycles/sec. Particular care is taken to filter the supply current to the amplifiers, and the circuits are carefully shielded to avoid parasitic currents. Variations in the supply current have practically no effect on either the amplified microphone voltage or the frequency of the oscillator, as all the electrical equipment is fed through a voltage regulator.

The details of the furnace and of the specimen support are shown in Fig. 6. The furnace consists essentially of a steatite tube enclosing the well-lagged winding. Inside the furnace is a heat-resisting steel tube which can be evacuated at the lower end. A special tap cuts off the tube from the pump. The support for the specimen and the electrodes can easily be withdrawn from the tube, which is sealed by means of bolts and rubber joints. Finally, cooling fins reduce the heating of the upper part of the support, which comprises the rubber-sealed joints, soldered parts, and the milled knobs for controlling the electrodes.

III.—EXPERIMENTAL RESULTS

1. COPPER–ALUMINIUM ALLOYS

The alloy investigated was made from electrolytic copper and commercially pure aluminium (99·96%) and contained 11·95 wt.-% aluminium, which corresponds closely to the eutectoid composition (11·8%). It had been found that if a high-purity
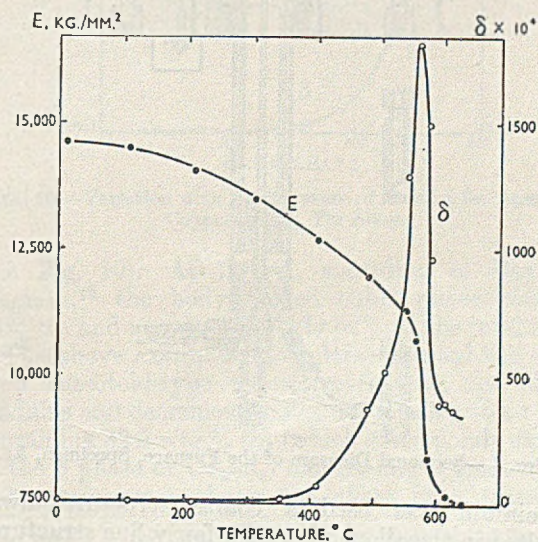
FIG. 5.—Variation with Temperature of E and δ for Annealed Copper–11·95% Aluminium Alloy.
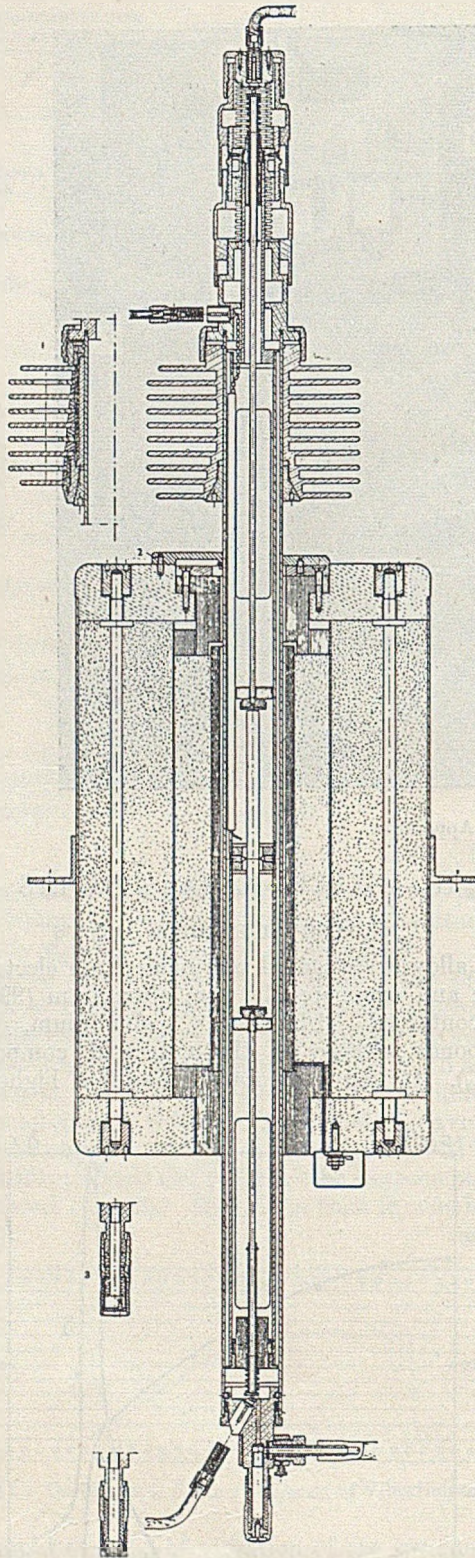
Fɪɢ. 6.—Sectional Diagram of the Furnace, Specimen, &c.

Fɪɢ. 7.—Variation with Temperature of E and δ for Annealed Copper–11·40% Aluminium Alloy.
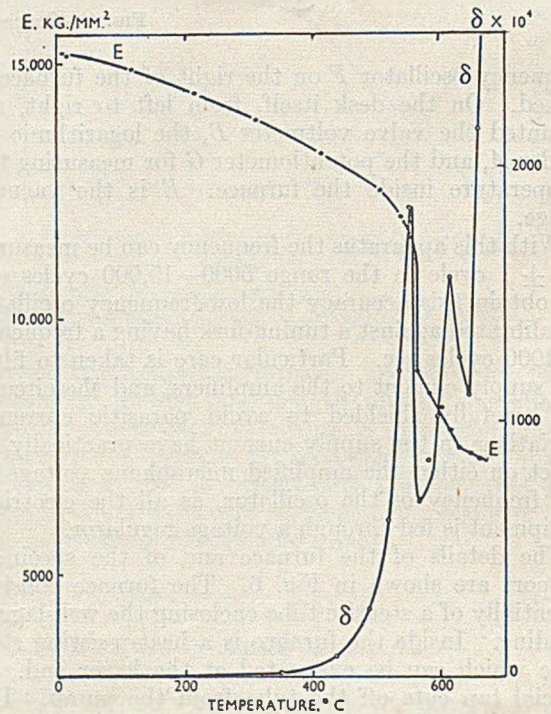
Fɪɢ. 8.—Variation with Temperature of E and δ for Annealed Copper–12·37% Aluminium Alloy.

aluminium was used a coarse structure always resulted in the alloy, whereas a fairly fine structure is desirable if adequate pseudo-isotropy is to be obtained. Ingots, 8 mm. in dia. and 150 mm. long, with a grain-size of 1150 grains/cm². , were made by casting into a graphite mould. The ingots were subsequently annealed for 100 hr. at 750° C.

Below 565° C., this alloy, when in equilibrium, consists of a mixture of α (face-centred cubic, containing 9·4% aluminium) and γ (cubic with 52 atoms

per unit cell, containing 15·6% aluminium). This condition is obtained only if the rate of cooling is less than $\frac{1}{2}°$ C./min. If an alloy annealed as described is heated, it undergoes at 565° C. the transformation $\alpha + \gamma \rightarrow \beta$, accompanied by a very marked reduction in the elastic modulus E (see Fig. 5), owing to the disappearance of the γ phase, whose modulus is 20,400 kg./mm.² at 20° C.[11] The modulus of the body-centred cubic β phase obeys the law of mixtures at 600° C., so that it can be calculated to within 5% from the moduli of copper and aluminium at that temperature. The logarithmic decrement δ increases continuously as the temperature is raised and attains a considerable value *before* and during the transformation in spite of the fact that the latter is isothermal (see Fig. 5). The decrement is as small in the β condition as it is in the $\alpha + \gamma$ state, but it reaches a value four times greater during the transformation.[12]

A hypo-eutectoid alloy and a hyper-eutectoid alloy, containing 11·40 and 12·37% aluminium respectively, were also studied. They undergo the $\alpha + \gamma \rightarrow \beta$ transformation at 565° C., but above this temperature the α phase in the first case and the γ phase in the second have still not entirely disappeared, and the β phase field is reached at 632° and 638° C., respectively.

The elastic-modulus/temperature and decrement/temperature curves therefore show two anomalies. The total disappearance of the α and γ phases is preceded by a marked increase in internal friction, which reaches a maximum during the transformation; this is also revealed by a break in the modulus curve (see Figs. 7 and 8).

2. Copper–Tin Alloys

The alloy investigated contained 24·9 wt.-% tin, whereas the eutectoid composition corresponds to 27 wt.-%. It was made from electrolytic copper and Union Minière du Haut Katanga tin and cast similarly to the copper–aluminium alloy; the ingots were subsequently annealed for 100 hr. at 750° C.

Below 520° C., the alloy when cooled at a rate of 1° C./min. consists of two phases: α, which has a face-centred cubic lattice and contains 15% tin at 500° C., and δ, which has a cubic lattice of the γ-brass type and contains 32% tin.* On heating, it undergoes a transformation similar to that in the aluminium bronzes, and above 520° C. it forms a body-centred cubic phase.

The disappearance of the δ phase, with an elastic modulus of 14,000 kg./mm.² at 20° C.,[13] results in a marked decrease in the modulus of the alloy. The decrement attains considerable values before the transformation and the decrement/temperature curve shows a relative maximum during the transformation (see Fig. 9).

Another hypo-eutectoid alloy containing only 20·2 wt.-% tin was also studied. In the annealed condi-

tion it contains the same phases as the previous alloy, but at 20° C. α is in excess.† The anomalies in the properties in the region of the transformation temperature are less marked, however, since above 520° C. this alloy still contains a considerable amount of α
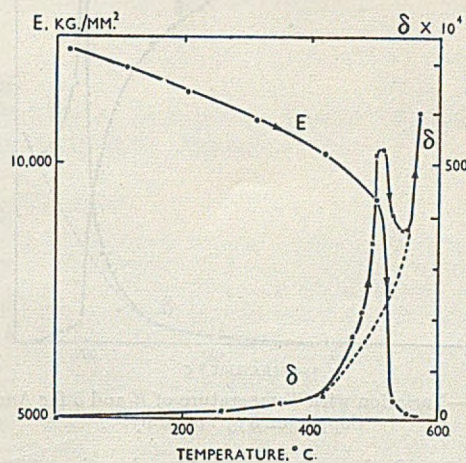
Fig. 9.—Variation with Temperature of E and δ for Annealed Copper–24·9% Tin Alloy.
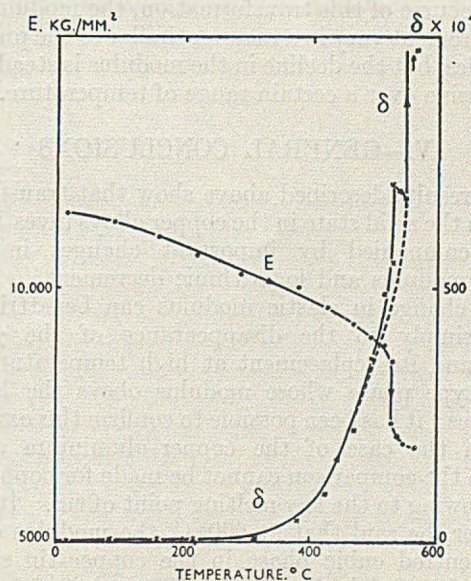
Fig. 10.—Variation with Temperature of E and δ for Annealed Copper–20·2% Tin Alloy.

(see Fig. 10). At 700° C., according to Raynor's diagram,[14] the body-centred cubic phase contains 23% tin and accounts for only 62% of the total mass.

The above examples have demonstrated the effect of a eutectoid-type transformation on the elastic modulus and decrement. Fig. 11, relating to an alloy containing 32·5 wt.-% tin (which corresponds exactly to the δ phase or Cu_3Sn_8) and having a grain-size of 2300 grains/cm.², shows the effect of a transition

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* The δ phase is not stable below 350° C., but the rate of decomposition is very slow, and the present authors have not determined it.

† The modulus of elasticity of this alloy at 20° C. was less

than would be predicted from the curve already published [13]; this was due to the presence of some micro-blowholes in the specimen, but does not alter the form of the curves in Fig. 10.
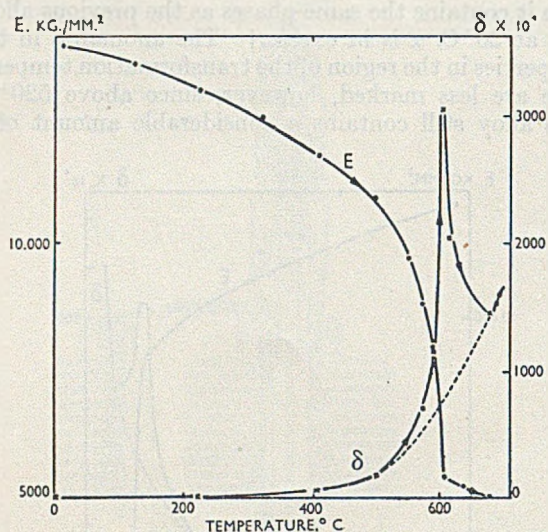
FIG. 11.—Variation with Temperature of E and δ for Annealed Copper–32.5% Tin Alloy.

point. The δ phase decomposes on heating to 590° C. into ζ and β^*, ζ having an hexagonal lattice with 26 atoms per unit cell and β a body-centred cubic lattice. In the course of this transformation, the modulus and the decrement vary in a manner similar to the previous examples, but the decline in the modulus is steady and progressive over a certain range of temperature.

IV.—GENERAL CONCLUSIONS

The results described above show that transformations in the solid state in the copper alloys investigated are accompanied by important changes in their elastic modulus and logarithmic decrement.

The change in elastic modulus can be attributed quite simply to the disappearance of the γ-type phase and its replacement at high temperatures by the β-type phase whose modulus obeys the law of mixtures; it has been possible to confirm this explanation in the case of the copper–aluminium alloys, though the comparison cannot be made for copper–tin alloys owing to the low melting point of tin. It may, however, be said that at 600° C. the modulus of the body-centred cubic phase in the copper–tin system lies on the straight line joining the modulus of copper and that of the ϵ phase (corresponding closely to Cu_3Sn) at this temperature. At 20° C., the modulus of ϵ obeys the law of mixtures.[11]

As regards the hypo-eutectoid alloy containing 20.2% tin, if the change in the modulus on passing through the transformation point is less marked, this is because the alloy contains an excess of α which does not take part in the transformation. Above the transformation temperature the modulus remains higher than that of the alloy containing 24.9% tin, since the α phase in it, being less rich in tin, has a greater modulus. It is known that the decrease in the elastic modulus of copper alloys is roughly

proportional to the content of added element in the region of the face- and body-centred cubic phases.[11]

The results for the logarithmic decrement are particularly interesting. From the high values obtained and the frequencies employed, it appears that the internal friction is here due for the most part to irreversible phenomena initiated by the vibrations. The dislocations set up in the crystals may be caused by the transformation or by the process of preparation for it; they may also be in existence before the transformation begins at all. The case of the copper–aluminium alloy, in which the transformation is considered to be isothermal and the constitution of which undergoes no change before reaching 565° C., is especially significant, for if the logarithmic decrement attains considerable values before reaching this temperature, it means that the isothermal transformation is preceded by a period of preparation, a phenomenon that appears to be revealed only by so sensitive a property as the internal friction. The same applies to the alloy containing 32.5% tin, consisting solely of δ. It is during the isothermal transformation itself that the decrement passes through a maximum, and this indicates that the instability of the lattice structure is the cause of an energy loss which ceases when the transformation is complete because the state is then more stable.

ACKNOWLEDGEMENTS

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* In Raynor's diagram [14] the body-centred cubic phase is designated γ, but the present authors prefer to use β for phases having this structure.

THE STRUCTURE AND SOME PROPERTIES OF TITANIUM–OXYGEN ALLOYS CONTAINING 0–5 AT.-% OXYGEN *

By A. E. JENKINS,† M.Eng.Sc., Junior Member, and
H. W. WORNER,‡ M.Sc., Member

SYNOPSIS

By means of measurements of thermoelectric power at various temperatures and by quenching experiments on alloys made with refined titanium, the limits of the (α + β) region up to 5 at.-% oxygen have been established. Alloys based on commercially pure titanium have also been studied, and it has been demonstrated that the impurities present in the commercial grade of metal cause a marked broadening of the α ⇌ β transformation range.

The mechanical working and annealing of commercially pure alloys have been examined, and it has proved possible to develop techniques for forging and swaging alloys containing as much as 3·5 at.-% oxygen. Alloys containing more than 1·5 at.-% oxygen may be hot worked, but they are more or less brittle at normal temperatures. Special attention has been given to some of the main factors affecting the cold drawing and annealing of alloy wires. The proof stress, nominal ultimate stress, plastic elongation, and hardness of annealed alloys have been determined, chiefly with a view to providing some quantitative concept of the variation of these mechanical properties with oxygen content.

I.—INTRODUCTION

Until recently, many metallurgists have regarded oxygen as an undesirable impurity in titanium, particularly in relation to its mechanical and working properties. However, recent investigations have shown that concentrations of oxygen in the region of 1·5 at.-% (0·5 wt.-%) do not seriously impair the working characteristics of the metal. Such small concentrations of oxygen do in fact cause a marked increase in the proof stress and the ultimate tensile stress of titanium, and in this connection, oxygen can now be regarded as an alloying element which may be very desirable for many practical purposes. The investigation to be described in this paper was undertaken with this idea in mind.

The early work of de Boer, Burgers, and Fast [1] on the electrical resistance of titanium indicated that oxygen can be held in solid solution in titanium and that the α ⇌ β transformation temperature is elevated by the introduction of oxygen. Subsequently, Ehrlich's [2] X-ray-diffraction investigations revealed that the maximum solid solubility of oxygen in α-titanium is approximately 30 at.-%. Recently Clark [3] has reported more precise values of the lattice parameters of α-titanium–oxygen solid solutions in the range up to 1·5 at.-%. Jaffee and Campbell [4] have investigated the effects of oxygen (up to 1 at.-%) on the electrical resistivity, microstructure, and cold-working characteristics of titanium, while Finlay and

Snyder [5] have studied in some detail the mechanical properties of titanium containing oxygen up to 0·75 at.-%. Since the completion of the present work, Jaffee, Ogden, and Maykuth [6] have published an account of their determination of part of the titanium–oxygen phase diagram. These workers have also reported some mechanical properties of alloys containing up to about 3 at.-% oxygen.

In the present investigation the range of oxygen contents studied was 0–5 at.-%. The first part of the work was aimed at establishing the α ⇌ β transformation range up to 5 at.-% oxygen. A knowledge of this portion of the titanium–oxygen phase diagram proved useful in connection with high-temperature annealing experiments. The second part of the investigation covered the working and mechanical properties of the alloys, including the effects of cold working and the determination of annealing ranges.

II.—EFFECT OF OXYGEN ON THE α ⇌ β TRANSFORMATION IN TITANIUM

1. Materials Employed and Preparation of Alloys

Titanium refined by the iodide process in the Philips Laboratories, Eindhoven, was used as the basis metal, the impurities present being : nitrogen, 0·03; oxygen, 0·02; iron, 0·02 at.-%; tin, strong

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trace; antimony, strong trace; and faint traces of copper, silicon, vanadium, magnesium, and manganese. The nitrogen and oxygen content were those reported by the Philips Laboratories.

The metal and the titanium–oxygen alloys were required in the form of wire or ribbon because it was decided to determine the $(\alpha + \beta)$ range by measuring the thermoelectric power/temperature relationship for each composition. The bar of refined titanium was therefore rolled down to ribbon 0·08 mm. thick, during which process two intermediate annealing treatments of $\frac{1}{2}$ min. each at about 550° C. were required. This heating was done in the air, and the resulting thin oxide film was removed by abrasion with waterproof carborundum paper.

The alloys were made by introducing oxygen into the titanium ribbon in the following way. A convenient length, usually about 8 cm., of the ribbon was degreased with carbon tetrachloride. After drying and weighing, the ribbon was heated uniformly in air to a temperature in the range 500°–750° C. for several minutes, the actual temperature and heating period depending on the amount of oxidation required. Then the oxide-coated specimen was heated *in vacuo* to 850°–950° C. for about 30 hr. to enable inter-diffusion between the oxide coating and the underlying metal to take place. To minimize the risk of the specimen becoming contaminated with carbon from oil vapours back-streaming from the diffusion pump or with silicon from the walls of the silica vacuum envelope, the ribbon was held within a long, hollow cylinder of titanium during the prolonged heating. The titanium receptacle was provided with a loosely-fitting titanium stopper, so that gases and vapours could not enter the inner chamber without having passed through the fine space between the stopper and the mouth of the cylinder. In this way, a gettered atmosphere was maintained around the specimen. It was considered that this precaution, together with the maintenance of a pressure in the range 10^{-4}–10^{-5} mm. Hg within the vacuum envelope, provided adequate protection for the specimen. At the end of the prolonged heating, the specimen was cooled and re-weighed, and the increase in weight was taken as the amount of oxygen introduced into the specimen. Careful chemical analyses revealed scarcely any detectable entry of nitrogen into the titanium, despite the fact that the metal was heated in air to produce the oxide coating. This was in agreement with the results of Carpenter and Reavell,[7] who have reported that the reaction between titanium and nitrogen is slow as compared with that between titanium and oxygen at temperatures up to 1000° C.

In most cases the amount of oxygen absorbed was too small to be determined accurately by chemical analysis. However, a specimen containing 4·9 at.-% oxygen (nominal composition) was analysed by the differential gravimetric method involving conversion of a weighed sample to TiO_2 and determination of the increase in mass caused by the conversion: $TiO_x \rightarrow TiO_2$. The analytical result was 4·7 $\pm$ 0·1

at.-% oxygen. In view of the small difference between the two figures, the nominal compositions were accepted in all cases.

In the early stages of the work, it was necessary to ascertain what period of heating at 850°–950° C. would be necessary to produce a practically uniform concentration of oxygen within the ribbon. Some exploratory work had indicated very approximately the location of the boundary between the α and $(\alpha + \beta)$ fields, and it had been shown that alloys quenched from the $(\alpha + \beta)$ field exhibited a mixture of well-defined dark and light regions on etching in an aqueous solution of 3% HF and 2% H_2O_2. It became evident that the dark-etching regions had consisted of β solid solution at the quenching temperature, but that the β phase had very rapidly changed (presumably to α) during quenching. Now it will be evident that a specimen of uniform oxygen content would exhibit an even distribution of the light- and dark-etching areas after quenching from the $(\alpha + \beta)$ range. Hence, to determine whether or not a uniform concentration of oxygen had been produced in any given prolonged period of heating, it was simply necessary to examine a polished and etched cross-section of the ribbon after quenching from any temperature in the $(\alpha + \beta)$ range. In this way, it was found that 25–30 hr. heating at 850°–950° C. produced a close approach to equilibrium conditions in specimens containing up to 4·9 at.-% oxygen. It was realized that if the prolonged heating was carried out at temperatures in the $(\alpha + \beta)$ region, then there was the risk of the specimen finally consisting of an outer rim of α encasing a core of β, each phase having its own equilibrium oxygen concentration at the temperature concerned. While such a specimen would be in an equilibrium state at the treatment temperature, it would not, of course, possess a uniform distribution of oxygen. Therefore, during the last 10–12 hr. of the homogenizing treatment, the temperature was kept in the range 850°–880° C., so that the specimen had an opportunity to approach a uniform concentration over the wholly α field.

In view of the decision to employ a thermoelectric property in determining the $(\alpha + \beta)$ range, another aspect of specimen preparation had to be kept in mind, namely, the possibility of there being a preferred orientation in the samples. The specimens were made from rolled titanium ribbon, which in its original state exhibited a preferred orientation, the (0001) planes tending to lie parallel to the plane of rolling. However, one of the authors (H. W. W.) had previously found [8] that the preferred orientation in refined metal could be practically eliminated by heating just above the $(\alpha + \beta)$ range (883°–887° C.). Therefore it was decided to begin the homogenizing treatment by heating rapidly up to 900°–950° C., in an endeavour to eliminate the preferred orientation before the onset of the interdiffusion between the oxide layer and the underlying metal. This measure appeared to be successful in all cases except that of the 4·9 at.-% alloy. The results pertaining to this alloy will be discussed in Section II, 2.
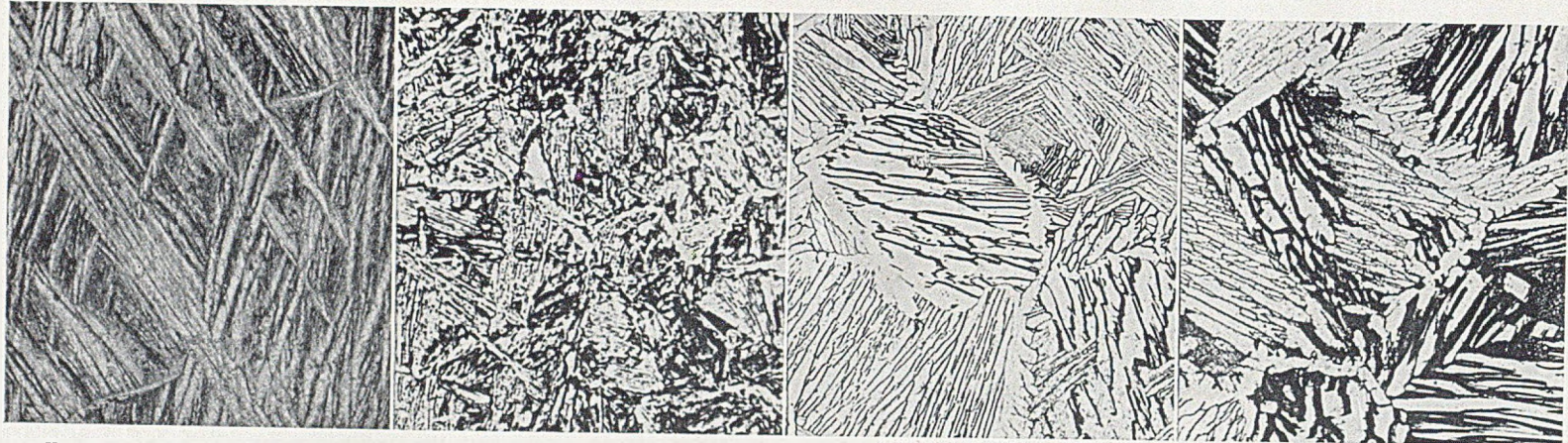
FIG. 7.—As Arc-Melted.　× 400.　　FIG. 8.—As Forged in Range 600°–850° C.　× 400.　　FIG. 9.—As Forged and Swaged at 600°–850° C., Heated at 900° C. for 4 Hr., and Cooled at 5° C./min. Single phase, α-titanium; dark-etching areas formerly β.　× 100.　　FIG. 10.—As Forged and Swaged at 900°–1000° C., Heated at 1000° C. for 4 Hr., and Cooled at 5° C./min. Single phase, α-titanium; dark-etching areas formerly β.　× 100.

TITANIUM–OXYGEN ALLOYS.

FIG. 11.—3 at.-% Oxygen Alloy Swaged at 900°–1000° C., Heated at 1000° C. for 5 Hr., and Cooled at 5° C./min.　× 100.　　FIG. 12.—1 at.-% Oxygen Alloy Cold Drawn to 73% Reduction in Area.　× 200.　　FIG. 13.—1 at.-% Oxygen Alloy Annealed for 3 Min. at 700° C.　× 100.

All specimens were etched in an aqueous solution of 3% hydrofluoric acid and 2% hydrogen peroxide.

PLATE XXIX.

2. DETERMINATION OF (α + β) RANGE BY THERMOELECTRIC-POWER MEASUREMENTS

In an earlier investigation, one of the authors (H. W. W.) had found that the α ⇌ β transformation in refined titanium was accompanied by an abrupt change in thermoelectric properties.[8] The curve relating the thermoelectric power of a titanium/platinum couple with temperature exhibited two marked discontinuities which corresponded with the extremities of the (α + β) range in the refined titanium. It was therefore considered that the limits of the (α + β) region in an alloy could be determined by observing the temperatures at which the discontinuities occurred in the thermoelectric power/temperature curve for that alloy.

The equipment and experimental technique employed in the thermoelectric measurements were essentially identical with those used in earlier work on pure titanium.[8] It was expected that the α ⇌ β transformation would be much slower in the alloys than in the pure metal because of the necessity for a considerable amount of diffusion to occur in them. The criterion for the establishment of equilibrium, or of a close approach to it, at a given temperature was the attainment of a steady value of the thermoelectric power. This required 20–30 min. at any temperature within the (α + β) region in the case of
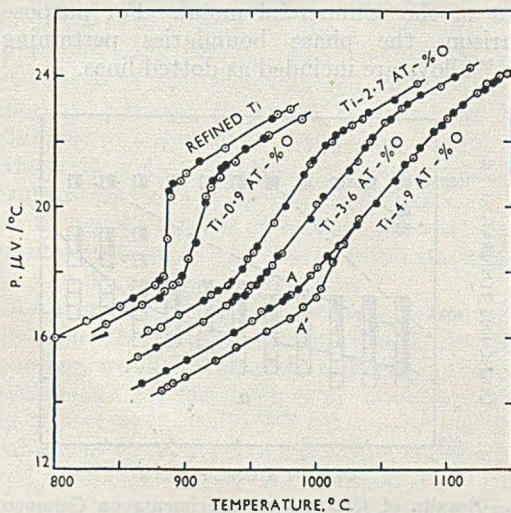
FIG. 1.—Temperature Dependence of the Thermoelectric Power (*P*) of Couples Comprising Pure Platinum and Titanium with 0–4·9 At.-% Oxygen.

KEY.

A′ represents first heating experiment with the Pt/Ti–4·9% O couple.
A represents subsequent cooling and heating experiments with the Pt/Ti–4·9% O couple.
○ Heating Curve.   ● Cooling Curve.

the 4·9 at.-% oxygen alloy, which was the most sluggish of the alloys studied.

The thermoelectric-power determinations are set out in Fig. 1, in which points determined during both heating and cooling are included. The degree of

correspondence between results obtained on heating and cooling is of value in revealing the extent to which equilibrium conditions had been achieved during the experiments. In this connection, the results for the 0·9, 2·7, and 3·6 at.-% oxygen alloys were satisfactory,
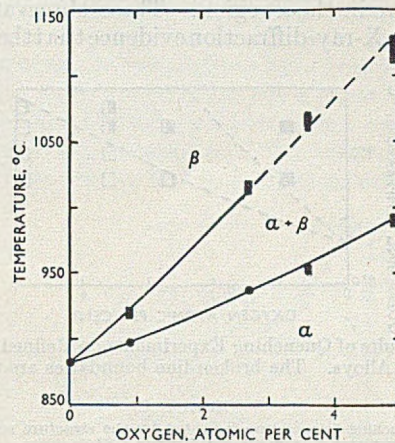
FIG. 2.—Phase Boundaries Deduced from Data in Fig. 1.

there being no definite signs of any hysteresis effects. In the case of the 4·9% oxygen alloy, the results pertaining to the first heating (curve A′ in Fig. 1) did not coincide with those obtained in subsequent cooling and heating experiments. However, the upward turn in curve A′ indicating the beginning of the α → β transformation occurred at the same temperature (990° C.) as that shown during subsequent cooling and heating. The same phenomenon had previously been noted in experiments on refined titanium which, in its initial state, had been annealed below the (α + β) range and which possessed residual preferred orientation until heated above the (α + β) region (cf. Worner[8]). In other words, the curve marked A′ in Fig. 1 almost certainly pertains to an alloy which still possessed some degree of preferred orientation, whereas curve A represents the 4·9% alloy free from preferred orientation.

The (α + β) ranges deduced from the curves in Fig. 1 are presented in Fig. 2. The extremities of the (α + β) region cannot be accurately determined from the curves for 3·6 and 4·9 at.-% alloys in Fig. 1, since the changes in gradient, dP/dT, on passing from the α or β field into the (α + β) region are not very sharply defined. This is especially true of the upper limit of the (α + β) range in each case. It is considered that this lack of sharpness is attributable to slight superficial contamination of the specimens during the experiments. Here it might be noted that exactly the same type of effect had been observed in earlier experiments on refined titanium which had suffered superficial contamination (cf. Worner[8]). In view of these observations, there is some uncertainty concerning the points denoting the high-oxygen portion of the β/(α + β) boundary; hence the latter is shown as a dotted line in Fig. 2.

3. Some Metallographic and X-Ray-Diffraction Observations on Quenched Alloys

When quenched from the β region and etched in an aqueous solution of 3% HF and 2% H_2O_2, refined titanium exhibited a Widmanstätten pattern. Such a structure was not produced by quenching from any temperature in the α region. These observations, together with X-ray-diffraction evidence that the β form of
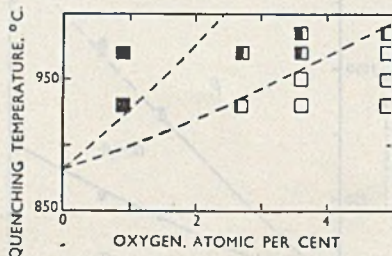
Fig. 3.—Results of Quenching Experiments on Refined Titanium–Oxygen Alloys. The broken-line boundaries are taken from Fig. 2.

KEY.

■ = β at quenching temperature indicated by the structure resembling fine Widmanstätten pattern (cf. Fig. 7, Plate XXIX).
□ = α at quenching temperature.
▣ = (α + β) at quenching temperature.

titanium could not be retained by quenching, indicated that the Widmanstätten pattern was a manifestation of a rapid β→α change having occurred during quenching. Similar, but more marked Widmanstätten structures, rather like that illustrated in Fig. 7 (Plate XXIX), were found in oxygen-bearing titanium quenched from the β field shown in Fig. 2. As in the case of refined titanium, X-ray-diffraction tests revealed no evidence of retained β phase in oxygen alloys quenched from the β field. Hence the presence of a Widmanstätten structure in an etched as-quenched alloy could be taken as evidence of the β phase having been present at the quenching temperature, and of rapid transformation of the β phase during quenching.

The results of some quenching experiments are presented graphically in Fig. 3, which also contains the boundaries deduced from the thermoelectric-power experiments. In general, the results of the quenching tests are in concordance with those shown in Fig. 2. The quenching experiments on the 4·9 at.-% oxygen alloy would suggest that at this oxygen concentration the α/(α + β) boundary lies a little below that estimated from the thermoelectric-power results. This may mean that the thermoelectric power/temperature relationship does not reveal the very first appearance of β phase on passing from the α into the (α + β) region. However, at lower oxygen concentrations, there is no marked discrepancy between the quenching-test results and those derived from the thermoelectric-power experiments.

4. Effect of Oxygen in Commercially Pure Titanium

All of the results presented above pertain to titanium–oxygen alloys made with refined titanium. However, it was the authors' intention to study the

working and mechanical properties of alloys based on commercially pure titanium, i.e. metal made by the reduction of titanium tetrachloride with magnesium. Hence it was necessary to ascertain the transformation range in the alloys of commercial grade. Special interest was centred on the α/(α + β) boundary. It was hoped that most of the hot-working and annealing treatments could be carried out wholly in the α field so that complications due to phase changes would not enter. For this purpose it was obviously necessary to determine the upper limit of the α phase field.

In an earlier investigation, one of the authors (H. W. W.[8]) had found from thermoelectric power/temperature studies that the α ⇌ β transformation in the commercial metal occurs over a broad temperature range. The transformation range could not be precisely ascertained, but it proved to be approximately 865°–970° C. Quenching experiments, similar to those outlined in the previous section, indicated that the range was from 860° C. up to some temperature between 965° and 980° C.

The α/(α + β) boundary in the system based on commercially pure titanium was determined by quenching experiments. The alloys were prepared by arc melting mixtures of titanium and titanium dioxide, Some details concerning the preparation and purity of the commercial alloys are given in Section III, 1 of the paper. Fig. 4 presents the results of the quenching tests. The results for the basis titanium are shown at 0·5 at.-% oxygen, this being the oxygen content of the commercial metal. For purposes of comparison, the phase boundaries pertaining to "pure" alloys are included as dotted lines.
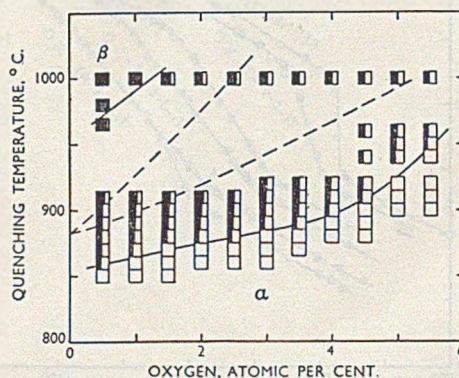
Fig. 4.—Results of Quenching Experiments on Commercially Pure Titanium–Oxygen Alloys. The broken-line boundaries refer to pure titanium alloys.

KEY.

■ = β at quenching temperature.
□ = α at quenching temperature.
▣ = (α + β) at quenching temperature.

It is evident that the impurities in the commercial grade of metal exert a marked influence on the transformation range of the alloys. The α/(α + β) boundary for alloys of commercial purity lies well below the α/(α + β) boundary of Fig. 2. It seems that the β/(α + β) boundary tends to rise rapidly to temperatures well above 1000° C., even at low oxygen

concentrations. The authors were not particularly concerned with the β region in this system; hence the investigation was not extended beyond 1000° C.

It will be noted that reference has been made to the (α + β) region in alloys of commercial purity, though it is not known for certain that it is in fact a simple (α + β) range. The commercial alloys contain several components, and it is possible that other phases may appear in the so-called (α + β) zone. However, for the sake of brevity, (α + β) has been used to denote the region between the α and β solid-solution fields.

5. Discussion on the Titanium–Oxygen Solid Solutions

Ehrlich[2] has shown that the titanium–oxygen α solid solutions are of the interstitial type, there being an increase in density and lattice parameters on passing from 0 to approximately 30 at.-% oxygen. The oxygen is almost certainly accommodated in the octahedral interstices in the α-titanium lattice, the bonding between the titanium and oxygen being essentially of the metallic-covalent type. If one adopts a "metallic" radius for oxygen, such as the value 0·80 Å. for co-ordination number 6 as deduced by Pauling,[9] then on the basis of the generally accepted rules pertaining to the lattice geometry of interstitial metallic phases, the extensive solubility of oxygen in α-titanium can readily be understood (cf. Worner[10]). On the other hand, considerations of lattice geometry suggest that serious distortion would occur if oxygen, having a "metallic" radius of approximately 0·80 Å., were introduced into the interstices of the body-centred cubic lattice of β-titanium. Hence it would be expected that the solubility of oxygen in β-titanium would be comparatively low. This idea is in agreement with the results obtained in the present investigation. In other words, considerations of lattice geometry explain in qualitative fashion why the addition of oxygen raises the α ⇌ β transformation range.

As a matter of general interest, it was decided to ascertain thermodynamically where the β/(α + β) boundary would lie using the following as bases for the calculation : (a) the α/(α + β) boundary shown in Fig. 2, and (b) the value 678 cal./g.-atom as the latent heat of the $Ti_\alpha \rightleftharpoons Ti_\beta$ transformation in the neighbourhood of the transformation temperature. This figure was determined by McQuillan[11] in an investigation of the titanium–hydrogen system.

The relationship employed was :

$$\left(\frac{d \ln x_\alpha/x_\beta}{dT}\right)_{\text{const. pressure}} = -\frac{\Delta_t H}{RT^2}$$

where x_α and x_β represent the atomic fractions of titanium in the α and β phases respectively, T is in °K., R is the gas constant, and $-\Delta_t H$ the latent heat of α ⇌ β transformation in pure titanium. It was assumed that the latent heat of transformation could be taken as constant in the range of temperature concerned.

Of course a boundary calculated in this way will be of interest only in the composition range in which the alloys can be described as dilute in the sense that the solvent (in this case titanium) obeys Raoult's law while the solute obeys Henry's law. There is some doubt as to the extent of the concentration range in which the titanium–oxygen solutions may, for practical purposes, be regarded as dilute in the above sense. In any case, the calculations have been restricted to the range 0–2 at.-% oxygen in the α phase. To facilitate comparison, the β/(α + β) boundary calculated as outlined above is included with the experimentally determined boundaries in Fig. 5. The calculated boundary
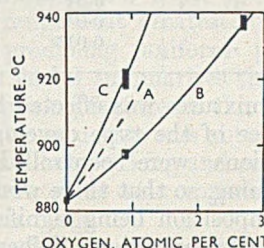
Fig. 5.—Comparison of the Calculated and Experimental β/(α + β) Boundaries. Curve *A* is a portion of the β/(α + β) boundary determined thermodynamically from the α/(α + β) boundary (curve *B*) assuming that both α and β solid solutions obey the simple laws of dilute solutions. Curve *C* is the experimentally determined boundary.

lies well below the experimentally determined curve. In view of the fact that oxygen cannot be accommodated very readily as an interstitial solute in β-titanium, it is not unlikely that the β solid solutions will deviate from the laws pertaining to dilute solutions even at concentrations of the order of 0·5–1 at.-%. This factor is probably an important one contributing to the discrepancy in Fig. 5.

The α/(α + β) boundary shown in Fig. 2 agrees reasonably well with that deduced from quenching experiments by Jaffee, Ogden, and Maykuth.[6] These authors have placed the β/(α + β) boundary somewhat higher than that obtained from thermoelectric power/temperature measurements in the present investigation. However, the discrepancy cannot be regarded as serious, especially when allowance is made for the experimental difficulties inherent in the determination of high-temperature phase boundaries in systems based on reactive metals such as titanium.

III.—THE WORKING AND MECHANICAL PROPERTIES OF COMMERCIALLY PURE TITANIUM–OXYGEN ALLOYS

1. Preparation of the Alloys

The alloys used in this part of the investigation were prepared by melting mixtures of chemically pure titanium dioxide powder and commercially pure titanium powder made by reducing titanium tetrachloride with magnesium. Impurities present in the basis titanium are listed in Table I. The value of

oxygen concentration was determined by vacuum-fusion analysis at the National Physical Laboratory, Teddington.

TABLE I.—*Impurities Present in Commercially Pure Titanium After Melting in an Argon-Arc Furnace.*

Impurity	Concentration	
	Wt.-%	At.-% (approx.)
Oxygen . . .	0·17	0·5
Nitrogen . . .	0·04	0·15
Carbon . . .	0·10	0·40
Iron . . .	0·20	0·18
Silicon . . .	0·04	0·07
Manganese . . .	0·04	0·04
Cobalt . . .	0·015	0·01

Melting of the mixtures was effected by means of an argon-arc furnace of the type developed by Kroll.[12] Melting conditions were controlled to minimize spitting and fuming, so that there would be little risk of the final composition being significantly different from the nominal composition. When the main part of the investigation was in progress, no facilities were available for having the alloys analysed. However, one alloy, a 2 at.-% oxygen alloy (nominal composition) was analysed by the chlorination method, the result obtained being 1·95 at.-%. Mr. J. A. Corbett carried out the analysis, the method being briefly as follows. A weighed sample was subjected to the action of a stream of pure, dry chlorine at 400° C. so that practically all the metal was converted to titanium tetrachloride, while the oxygen remained as titanium dioxide in the residue. The latter was weighed and analysed, and from this result, the oxygen content of the original sample was calculated. In a more recent investigation, some titanium–oxygen alloys prepared as outlined above were analysed, and it was found that the nominal compositions agreed with the analytical results within ± 5%. The use of nominal compositions therefore seems justified. The following list indicates the nominal oxygen contents of all the commercially pure alloys used in the present investigation :

At.-%	Wt.-%	At.-%	Wt.-%
0·5 *	0·17 *	3·5	1·20
1·0	0·33	4·0	1·37
1·5	0·51	4·5	1·55
2·0	0·67	5·0	1·73
2·5	0·85	5·5	1·91
3·0	1·02		

* Basis titanium.

The button-shaped specimens produced in the arc furnace were coarse grained, and on etching they exhibited a very marked Widmanstätten pattern, as shown in Fig. 7 (Plate XXIX). This structural feature was probably due to segregation during freezing of the alloys, and to the β → α transformation during cooling through the range 1000°–850° C. Early in the investigation it became apparent that alloys exhibiting structures like that shown in Fig. 7 were much less amenable to deformation than specimens possessing a normal fine polygonal grain structure. The coarse Widmanstätten structures could be eliminated and the usual polygonal grain texture produced only by subjecting the alloys to very considerable amounts of hot working followed by long heating. Section III, 3 deals with an investigation of the homogenization of the alloys, i.e. with the elimination of the cast structure.

2. WORKING AND MACHINING QUALITIES

During the forging and swaging of the button-shaped as-melted specimens into useful shapes such as rod and wire, numerous observations of working qualities were made. The essential features of these observations are summarized graphically in Fig. 6. At normal temperatures, alloys containing more than 2·5 at.-% oxygen were brittle, no matter what the
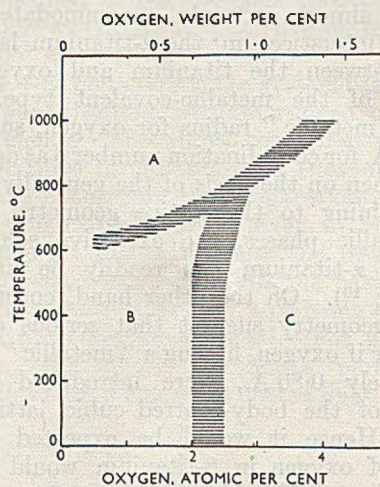
FIG. 6.—Working Qualities of Titanium–Oxygen Alloys at Various Temperatures.

KEY.

Region A : Alloys may be hot worked without intermediate annealing.
Region B : Alloys may be worked to a limited extent and require intermediate annealing at temperatures in the A region.
Region C : Alloys are, for all practical purposes, brittle in this region.

previous mechanical and thermal history of the specimens. In order to prevent the formation of cracks during forging or swaging of alloys containing more than 2·5 or 3 at.-% oxygen, working temperatures approaching 1000° C. had to be adopted. Even at 1000° C. alloys with 4% or more oxygen exhibited a marked tendency to crack during forging. Once a crack formed during hot forging, there was little chance of its healing during subsequent hot working, because of the rapidity of oxidation along the fissure. The transition from the hot-working temperature region (*A* in Fig. 6) to the range of brittleness (*C*) occurred over a comparatively small temperature interval, and this added to the difficulty of producing sound specimens of alloys containing more than 2·5 at.-% oxygen.

The production of alloy wires by drawing at room temperatures was investigated, and the results of this work are reported in Section III, 4.

The preparation of tensile-test specimens enabled the authors to compare the machining qualities of the alloys. High-speed steel tools were used, the rake and clearance angles being similar to those used in machining high-strength alloy steels. It proved advantageous to employ cutting fluids of the type used in machining steels. Table II summarizes the

TABLE II.—*Machining Properties of Annealed Titanium–Oxygen Alloys.*

Oxygen Content (nominal), at.-%	Comments
0·5 1·0 1·5 2·0	Machined easily, there being no difficulty in drilling or screw cutting.
2·5 3·0	Could be turned satisfactorily, but inclined to flake on screw cutting. Hard to drill.
3·5 4·0 4·5 5·0	Difficult to machine, screw-cutting and drilling being very unsatisfactory and practically impossible on 4·5% and 5·0% alloys.

observations made during turning, screw-cutting, and drilling operations. In general, it can be stated that machining alloys containing more than 2 at.-% oxygen proved difficult. In the range 0·5–2 at.-% oxygen, the machining qualities of the alloys were similar to those of high-strength structural alloy steels.

3. Homogenizing Treatments

In the early stages of the investigation some difficulty was experienced in producing alloy specimens with uniform polygonal grain structures. There was a marked tendency for the as-cast structure (Fig. 7, Plate XXIX) or a distorted form of it (Fig. 8, Plate XXIX) to persist despite the application of hot working and subsequent prolonged heat-treatments. It was therefore decided to make a systematic study of some of the main factors concerned in the elimination of the cast structure, viz. (a) the hot-working temperature, and (b) the subsequent heating conditions, i.e. temperature, duration of heating, and final cooling rate.

The amount of distortion effected during hot working was more or less constant, because all the buttons from the arc furnace had to be converted to specimens of rod form about $\frac{3}{16}$–$\frac{1}{4}$ in. in dia. Hence the total amount of hot working before the homogenizing heat-treatment did not enter as a variable factor. It should be noted that the buttons were shaped into rod by working in region A of Fig. 6, this being done largely to lighten the load on the working machinery. But the hot working alone did not suffice to eliminate completely the cast structure, even when temperatures as high as 900°–1000° C. were tried. The hot-working ranges could be broadly classified into: (a) within the α range (approximately 600°–850° C.) and (b) within the (α + β) range (900°–1000° C.).

Two ranges of homogenizing heat-treatment temperatures were tried: (a) within the α range (800°–850° C.), followed by cooling at about 100° C./min. and (b) within the (α + β) range (900°–1000° C.), followed by fairly slow cooling (about 1°–5° C./min.).

The heating periods ranged from 1 to 30 hr. In all cases the prolonged heat-treatments were carried out *in vacuo*, the pressure being kept below 10⁻⁴ mm. Hg. As an extra precaution against contamination, the specimens were held within a loosely stoppered hollow titanium cylinder during the heat-treatments.

Most of the study of homogenization was concerned with alloys containing 0·5–2 at.-% oxygen, because it was in this range that greatest difficulty was experienced in producing uniform polygonal grain structures. Table III summarizes the observations.

A rather surprising feature of the results was that the most desirable structure was obtained by carrying out both the final hot working and subsequent homogenizing heat-treatment at comparatively low temperatures in the α range.

It will be noted that coarse Widmanstätten structures persisted after slow cooling from the (α + β) range. This is considered to be largely due to the broadness of the (α + β) region in alloys of commercial purity (cf. Fig. 4). In other words, there is such a wide difference between the compositions of the α and β phases at any temperature in the (α + β) region,

TABLE III.—*Results of Investigation into Homogenization of Alloys Containing 0·5–2·0 At.-% Oxygen.*

Hot-Swaging Temp. Range		Homogenizing Temp. (periods up to 30 hr.)		Cooling Rate, °C./min.	Final Microstructure
Region	°C.	Region	°C.		
α	600–850 (depending on oxygen content)	α	800–850	100	Clean recrystallized α structure. Similar to Fig. 13 (Plate XXIX), but coarser. If worked consistently close to (α + β) region, tendency to leave unchanged the original broken-up cast structure (Fig. 8, Plate XXIX).
		(α + β)	900–1000	1–5	Very coarse grain structure. α formed in Widmanstätten pattern (Fig. 9, Plate XXIX).
(α + β)	900–1000 (depending on oxygen content)	α	800–850	100	Similar to Fig. 8 (Plate XXIX) but coarser. No change from swaged structure.
		(α + β)	900–1000	1–5	Very coarse grain structure (Fig. 10, Plate XXIX).

that diffusion during cooling at 1°–5° C./min. is unable to yield a uniform α structure.

From a full consideration of the results presented in Table III, the following forging and swaging procedure was developed for alloys in the range 0·5–2·0 at.-% oxygen. The button-shaped ingots from the arc furnace were forged down to 0·5-in.-dia. bar, which was then swaged in four consecutive steps to about 0·3 in. in dia. All these operations were performed with the alloy at 900°–1000° C. Subsequently, the bars were reduced to 0·2 in. in dia. in two steps, in the

range 600°–850° C. After pickling in an aqueous solution containing 10% hydrofluoric acid and 50% nitric acid, the bars were heated *in vacuo* to 800°–850° C. for 4 hr. in order to complete the homogenizing process. The above sequence of hot working and heating operations produced a clean, polygonal grain structure, somewhat coarser than that shown in Fig. 13 (Plate XXIX).

Homogenization of alloys containing 3–5 at.-% oxygen did not prove as difficult as with the more dilute alloys. It was simply necessary to hot forge and swage the high-oxygen alloys at 1000° C., heat *in vacuo* to 1000° C. for about 5 hr., and finally cool at 1°–5° C./min. After this treatment, some of the oxygen-rich alloys exhibited an interesting " block " type of microstructure, as shown in Fig. 11 (Plate XXIX).

Whenever close temperature control during hot working was desired, the specimens were heated to the temperature by immersion in a charcoal-covered tin bath. This method of heating minimized the superficial oxidation of the specimens, and the tin itself caused no serious contamination. In many cases, heating in air was satisfactory when close temperature control was not required. The oxide layers were removed by acid solutions after hot working, as described above.

4. Drawing and Annealing Alloy Wires

In this part of the investigation the composition range covered was 0·5–2·5 at.-% oxygen. The homogenized bars were hot swaged from 0·2 in. dia. down to 0·1 in. Table IV presents an outline of the essential features of working and annealing conditions, as well as the grain-size and hardness of the 0·1-in. wire stock.

TABLE IV.—*Preparation and Some Properties of Wire-Drawing Stock.*

	Oxygen Content (nominal), at.-%				
	0·5	1·0	1·5	2·0	2·5
Swaging temp., °C.	700	725	750	750	750
Stages at which annealed,* bar dia., in.	0·10	0·10	0·10	0·13 and 0·10	0·15 and 0·10
Approximate grain-size after final anneal, grains/mm.²	1000	1000	750	750	500
Vickers hardness number after final anneal	220	265	310	350	385

* 15 min. at 800° C. *in vacuo.*

Before beginning to draw the wires through tungsten carbide dies, they were anodized in 5% aqueous acetic acid solution, the potential difference across the cell being about 50V. The anodized wires

were warmed and coated with a heavy gear-box oil to ensure smooth motion during wire drawing, which was done at room temperature. The wires were pointed by dipping into the nitric–hydrofluoric acid pickling mixture mentioned above.

TABLE V.—*Summary of First Part of Wire-Drawing Investigation.*

	Oxygen Content (nominal), at.-%				
	0·5	1·0	1·5	2·0	2·5
Dia. after final draw, in.	0·020	0·020	0·050	0·090	No reduction possible on account of brittleness at room temperature.
Reduction in area, %	96	96	75	20	
Vickers hardness after final draw	320	405	415	360	
Vickers hardness of initial annealed 0·1-in. stock (cf. Table IV)	220	265	310	350	
Comments	No sign of brittleness at finish.	No sign of brittleness at finish.	Fractured after 75% reduction.	Fractured after 20% reduction.	

Table V presents a summary of the first part of the wire-drawing investigation.

In order to produce wires of fine gauge from the basis titanium and the 1·0 and 1·5 at.-% oxygen alloy specimens, intermediate annealing treatments had to be given. It was also necessary to limit the amount of reduction in cross-sectional area by cold drawing before these annealing treatments, the aim being to prevent over-working of the alloys. For this purpose reductions approximating to 75% of the values shown in Table V were considered to be safe.

With a view to gaining a general understanding of the annealing characteristics of the 1·0 and 1·5 alloys as well as of the basis titanium, some short-term annealing experiments were conducted. In each case, the previous reductions were in accordance with the limits mentioned in the previous paragraph, and the actual values were :

	Redn. of Area, %
Basis titanium	approx. 73
1·0 at.-% oxygen . . .	approx. 73
1·5 at.-% oxygen . . .	approx. 55

Two methods of annealing were employed, the object being to compare a simple 3-min. heating in air with a 15 min. anneal *in vacuo* (pressure approximately 10^{-4} mm. Hg). The anneal in air simply involved placing the specimen and a thermocouple in a small silica sheath which was dipped into a molten tin bath. A period of 3 min. was allowed after the attainment of a steady temperature, which generally took place within 30 sec. of immersion. The specimens annealed

in vacuo were heated for 15 min. at temperature and allowed to cool at a rate of about 75° C./min. Annealed specimens were set in Bakelite and a longi-

FIG. 14.—Results of Annealing Experiments on Cold-Drawn Wires. Amount of previous reduction :

Oxygen Content, %	Redn. of Area, %
0·5 (Basis Titanium)	73
1·0	73
1·5	55

KEY.

×	3 min. heating in air.	R_i	Beginning of recrystallization.
●	15 min. heating *in vacuo.*	R_c	End of recrystallization.

tudinal diametral section prepared for hardness testing and observation of the microstructure.

The results of the annealing investigation are presented graphically in Fig. 14., the hardness results being obtained from indentations made along the centre of the wire. In all cases the first signs of softening occurred on heating in the range 300°–400° C., even when the heating period was only 3 min. The hardness/temperature curves for the basis titanium and the 1·0% alloy suggest that there are two main stages in the softening process, one in the range 300°–450° C., the other from 500°–600° C. The first signs of recrystallization, as revealed by observation under the microscope, appeared in both the basis titanium and the 1% alloy after heating to 550° C. This suggests that recovery alone was occurring during heating at temperatures up to about 500° C., and that recrystallization became an important factor above 500° C. In the case of the 1·5% alloy, there was a suggestion of two stages in the annealing

process, but the effect was not as marked as in the other two instances. Here it might be added that the first signs of recrystallization in the 1·5% alloy were noted after annealing at 600° C.

It will be evident that the 3-min. annealing treatments were almost as effective as the 15-min. treatments. Both sets of curves exhibited the same general features.

Completely recrystallized structures were observed in the basis titanium and the 1% alloy after heating to 600° C. for either 3 or 15 min. A slightly higher temperature (650° C.) was required for the 1·5% alloy. In general, it may be said that for all three compositions the most uniform fine-grained structures were produced by heating for either 3 or 15 min. at 700° C. A typical structure produced during a 3-min. anneal at 700° C. is shown in Fig. 13 (Plate XXIX). This is a 1 at.-% oxygen alloy which, in the cold-drawn condition (73% reduction of area), exhibited the structure shown in Fig. 12 (Plate XXIX).

5. SOME MECHANICAL PROPERTIES OF ANNEALED ALLOYS

All alloys examined in this part of the investigation had been annealed to produce fully recrystallized structures with a grain-size of the order of 1000 grains/mm.². This was done so that variations in grain-size would not be a significant factor in the consideration of the results. The annealing temperatures used were in the range 700°–850° C., and the period of heating was usually 2–4 hr.

The modulus of elasticity of a 2 at.-% alloy was compared with that of the basis titanium. Specimens having a 0·2-in. dia. over a length of 4 in. were employed, and the elastic strain over a 2-in. gauge-length was determined by means of a Tuckerman
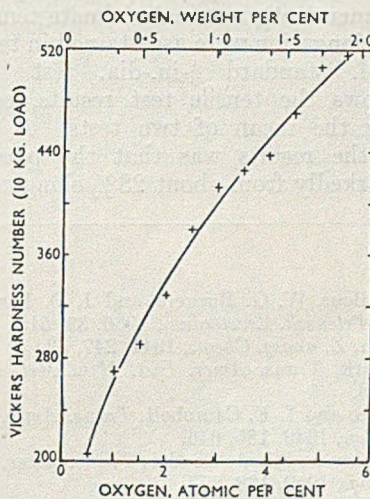
FIG. 15.—Hardness/Oxygen Concentration Relationship for Annealed Titanium–Oxygen Alloys.

optical strain-gauge with a sensitivity of $\pm 5 \times 10^{-6}$ in./in. Within the limits of measurement, the value of Young's modulus for the 2% alloy equalled that

for the basis titanium, the actual result being (1·66 ± 0·01) × 10[7] lb./in.².

From the results of numerous Vickers hardness tests, it was possible to establish with reasonable certainty the hardness/concentration relationship for annealed alloys containing up to 5·5 at.-% oxygen. In Fig. 15, each point represents the mean of about 10 determinations made on 3 or 4 specimens. The scatter in the case of each alloy was of the order of
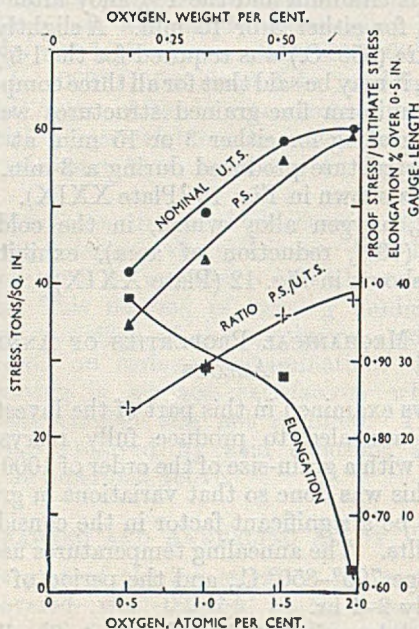
Fig. 16.—Mechanical Properties of Annealed Titanium–Oxygen Alloys. Proof stress is for approximately 0·2% permanent set.

± 5 Vickers units. It will be noted that the rise in hardness with increasing oxygen content is quite rapid.

Some values of the proof stress (approximately 0·2% permanent set), nominal ultimate tensile stress, and plastic elongation were ascertained in tensile tests on A.S.T.M. standard ⅛-in.-dia. test specimens. Fig. 16 shows the tensile test results, each point representing the mean of two tests. A particular feature of the results was that the plastic range dropped markedly from about 28% elongation in the case of the 1·5 at.-% oxygen alloy to just under 3% for the 2 at.-% oxygen alloy   Also, there was a marked rise in the ratio proof stress/nominal ultimate stress in the composition range studied.   Extrapolation of the curves in Fig. 16 would suggest that alloys containing more than approximately 2·5 at.-% oxygen would be brittle at normal temperatures.   Some very simple experiments on the 3 at.-% alloy confirmed this observation.

During the " necking down " stage in tensile tests, the basis titanium exhibited a non-uniformity of deformation near the point of fracture, resulting in the development of longitudinal ridges.   This phenomenon was not observed in the alloys.   The difference in behaviour between the basis metal and the alloys could not be attributed to grain-size, as all the specimens were similar in this respect.

6. Conclusion

The room-temperature properties of alloys containing up to 1·5 at.-% oxygen are such as to suggest that these alloys may be of value in certain engineering applications. A particular feature of the alloys is that they possess a comparatively high proof stress.

Finally, it is worth noting that the mechanical properties of the alloys could not be varied by quenching them from the β solid-solution field. It is thought that this was due to the great rapidity of the β → α transformation on cooling through the (α + β) range.

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SOME OBSERVATIONS ON THE DEFORMATION OF POLYCRYSTALLINE ZINC *

By J. A. RAMSEY,† M.Sc.

SYNOPSIS

It is shown that polycrystalline zinc, when deformed, behaves in a similar way to aluminium in that, both at elevated temperature and at slow strain rates, it tends to form a sub-grain or cell structure within the grains. Evidence is presented in support of the view that the cell structure is produced directly by the deformation and cannot be adequately explained by the same mechanism as that suggested for polygonization.

I.—INTRODUCTION

A NUMBER of papers [1,2] have appeared recently on the deformation of aluminium by creep at elevated temperatures, and it has been demonstrated that the grains break down into distinct elements or " cells ". Similar observations having been made by the present author [3] on the rapid deformation of aluminium at an elevated temperature, it was thought desirable to seek confirmation with another metal, and zinc was chosen for the purpose because it belongs to a different crystal system and was readily available in a very pure form.

II.—EXPERIMENTAL TECHNIQUE

Flat tensile specimens were prepared from strips of zinc produced by hot rolling cast ingots of purity 99·999%. Great difficulty was experienced in developing a uniform grain-size both within one specimen and from specimen to specimen after the same treatment. The method finally adopted was either to anneal the specimens at 270° C. after cutting from the rolled strip, or to anneal them at the same temperature after an elongation of 10%. Since the as-rolled strip was self-annealing, the former method depended upon grain growth and the latter upon recrystallization. The strain-anneal method was employed when a number of specimens of the same average grain-size were required, although it had the disadvantage of rumpling the surface in a manner which persisted even after electropolishing and made microscopic observation more difficult.

Electrolytic polishing was carried out by the method recommended by Rodda,[4] viz. by the use of a solution of 20 g. chromium trioxide in 1000 ml. of water. Satisfactory surfaces were obtained after 45 sec. with a current density of 16 amp./in.². This electro-polishing usually etched the grain boundaries adequately, but, if deeper etching was required, a reagent consisting of 15 g. sodium sulphate and 20 g. chromium trioxide in 1000 ml. of water was employed.

X-ray observations were made by the standard back-reflection technique, using cobalt radiation and reflections from the (105) and (114) planes.

Deformation was applied by stretching in a hand-operated tensile machine which could be encased in a furnace for work at elevated temperatures. Since it was necessary to cool the specimens before microscopic examination, it was not certain whether some of the features observed were a direct result of the deformation or whether they occurred during furnace-cooling. In order to overcome this objection, the changes in the microstructure of some specimens, such as those referred to in Section V, were followed while the specimen was being deformed at an elevated temperature. For this purpose, use was made of a small tensile machine which could be mounted on the stage of a Bausch and Lomb metallographic microscope and in which the specimen was surrounded by a small heating coil. In this way a given group of grains could be kept continuously under observation during the initial heating, whilst the specimen was being extended, and during cooling.

Specimens were generally extended at the rate used in normal tensile testing. Some specimens, however, were extended in a special machine at a low rate of approximately 0·15% elongation per hr. in order to examine the effect of the strain rate. For convenience the former will be referred to as rapid, and the latter, as slow straining.

III.—RAPID DEFORMATION AT VARIOUS TEMPERATURES

Microscopic examination of specimens stretched at room temperature revealed slip and twinning. Fig. 1 (Plate XXX) illustrates a typical area on a specimen extended 5%, in which slip was confined to one set of planes. However, in a large number of grains, a set of faint, more or less parallel bands was present, which were at an angle to the slip lines. This phenomenon is clearly shown in Fig. 2 (Plate XXX) in the

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direction indicated by the arrow, and appears very like a large number of small twins. Similar markings were noted recently by Jillson [5] and were attributed by him to "kinking". Another feature of the deformation was the existence of narrow bands running across the slip, the slip lines being deviated across them. These bands apparently provide a limit to the formation of twins, as shown in Fig. 2.

X-ray diffraction revealed that the arcs from individual grains consisted of a number of spots on a continuous background (Fig. 21, Plate XXXI), suggesting that the grains had broken down into a mixture of coarse fragments and very small fragments, the former being responsible for the spots and the latter for the continuous background. This type of fine structure has been correlated with kinking by Jillson, but, at least in the present case, the scale of the two phenomena appears to be quite different. From the microscopic appearance the only other evidence of discrete elements in the grains was the existence of the narrow deformation bands which could be considered to be the boundaries of areas in the grains of different degrees of tilting. However, though these were generally of about the right order of size, they were not observed in a very large number of grains. Whatever the cause of the fine structure, subsequent heat-treatment had the effect of sharpening the spots and reducing the continuous background, thus producing arcs consisting of a number of sharp spots. This suggests that the cause of the fine structure after recovery was present in the deformed state and was only made more easily observable by the heat-treatment. In comparing these structures with those resulting from deformation at elevated temperature, it should be noted that the grain boundaries in Fig. 1 (Plate XXX) are fairly sharp.

On raising the temperature of deformation, the most striking feature observed was a very definite sub-grain or "cell" structure in a large number of grains. This is illustrated in a very spectacular manner by Fig. 3 (Plate XXX), which shows the structure of a specimen extended 3·1% at 200° C. Fig. 4 (Plate XXX), which shows the same main grain shifted to the right and photographed with very slightly oblique illumination, clearly demonstrates that the surface of the grains was inclined differently from cell to cell. This was later confirmed by observations on the surface, using Tolansky [6] fringes. It will be noted that the cell boundaries are very regular and well marked, in contrast to those in aluminium, which are very irregular and difficult to observe.

Comparison between Figs. 5 and 6 (Plate XXX) which illustrate the condition after a 5% elongation at 100° and 200° C., respectively, shows that the cell structure is more clearly marked and is present in many more grains at the higher temperature. There was no significant difference in the cell sizes at the two temperatures. Recrystallization occurred during the process of elongation to 5% at 250° C.

At all temperatures, slip took place in the vast majority of grains, and there was no change in spacing or thickness of the slip lines. This type of behaviour contrasts markedly with that of aluminium, in which both the spacing and "width" of the slip lines increased with the temperature and, at the highest temperatures, no slip at all was observed. However, in some grains of zinc, no slip was observed; this was always the case in those grains manifesting the most clearly developed cell structure.

Generally, the slip lines in the grains containing both cells and slip were continuous across the cell boundaries, and were either straight or slightly deviated. The slight directional changes appeared consistent with the tilting of the grain surface from cell to cell after the occurrence of slip. In a few cases very marked directional changes in the slip were found. This was so in grains in which the cell boundaries were more or less at right angles to the slip direction and is illustrated by Fig. 7 (Plate XXX), in which it appears to be similar to the effect obtained at room temperature (Fig. 2).

In large grains especially, the cell size was observed to vary within the grains, being finer along one of the boundaries and increasing in size towards the centre of the grain. Long narrow cells were often found lying parallel to the grain boundaries, as illustrated in Fig. 3 (lower right grain), giving the appearance of traces of a migrating boundary. That migration is not the explanation will be demonstrated later.

The increasing facility of twinning with increasing temperature of deformation in zinc, reported by Davidenkov, Kolesnikov, and Fedorov,[7] was not borne out by the present work. It was found that twinning was absent in fine-grained specimens and much diminished in coarse-grained ones, after deformation at elevated temperatures.

On electropolishing and re-etching the specimens, both the old boundaries and the cell boundaries reappeared, proving that the network of boundaries within the grains was not purely a surface effect and that recrystallization to a fine grain-size was not responsible (cf. the complete grain shown in Figs. 3 and 8, Plate XXX). The effect referred to earlier, viz. the presence of long narrow cells parallel to the grain boundaries, is shown also in Fig. 8 (Plate XXX) after repolishing and re-etching (lower right grain). The fact that the original boundary and the cell boundaries reappeared refutes the idea that the phenomenon was due to boundary migration since, if the latter were the case, only one line would reappear on the site of the final position of the boundary.

In order to study the relative orientation of the cells, a specimen that exhibited a cell structure was extended at room temperature after repolishing and re-etching. It was found that grains containing a very well-developed cell structure, such as the grain in Fig. 3 (Plate XXX), showed less tendency to slip than others, a greater overall extension being necessary before they manifested any slip. This may be interpreted in two ways; either the existence of
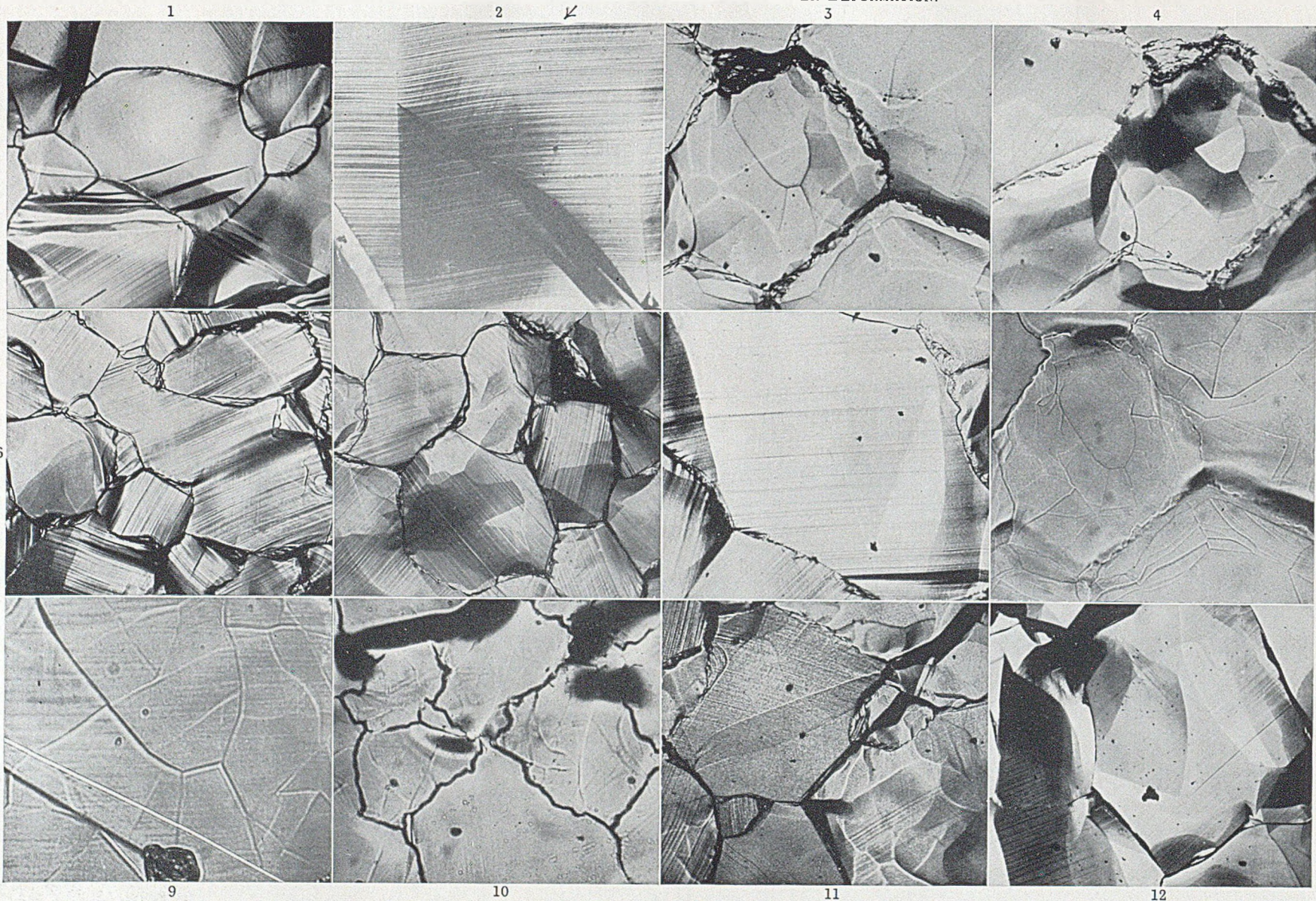
Fig. 1.—5% Extension at 21° C.   × 100.

Fig. 2.—5% Extension at 21° C.   Coarser grain-size than in Fig. 1.   × 100.

Fig. 3.—3·1% Extension at 200° C.   Vertical illumination.   × 100.

Fig. 4.—3·1% Extension at 200° C.   Oblique illumination.   × 100.

Fig. 5.—5% Extension at 100° C.   × 100.

Fig. 6.—5% Extension at 200° C.   × 100.

Fig. 7.—3·1% Extension at 200° C.   × 100.

Fig. 8.—Fig. 3 After Repolishing and Re-Etching.   × 100.

Fig. 9.—Centre of Grain on Left in Fig. 8, after 3% Extension at Room Temperature, Showing Slip Lines and Twin Crossing Cell Boundaries.   × 250.

Fig. 10.—5% Extension at 200° C., Repolished and Re-Etched.   × 100.

Fig. 11.—10% Extension by Creep at 21° C.   × 100.

Fig. 12.—12% Extension by Creep at 100° C.   × 100.

PLATE XXX.
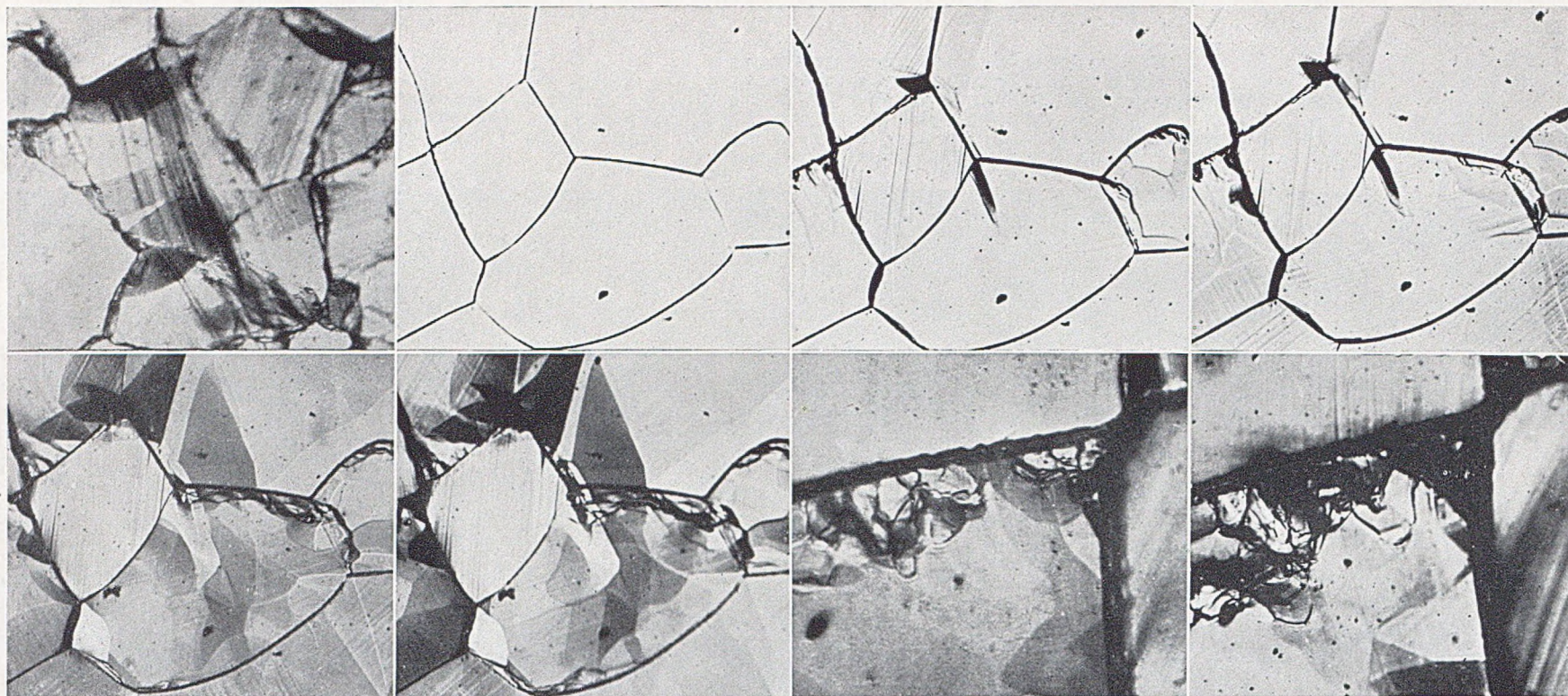
FIG. 13.—20% Extension by Creep at 21° C.   Very fine grain-size.   × 500.
FIG. 14.—Annealed Condition.   × 100.
FIG. 15.—Fig. 14 After 2·5% Extension at 220° C.   × 100.
FIG. 16.—Fig. 14 After 5% Extension at 220° C.   × 100.

FIG. 17.—Fig. 14 After 10% Extension at 220° C.   × 100.
FIG. 18.—Fig. 14 After 15% Extension at 220° C.   × 100.
FIG. 19.—Grain Boundary in Fig. 17 Marked by Arrow.   × 500.
FIG. 20.—Same Boundary as in Fig. 18.   × 500.

X-Ray Diffraction Patterns.
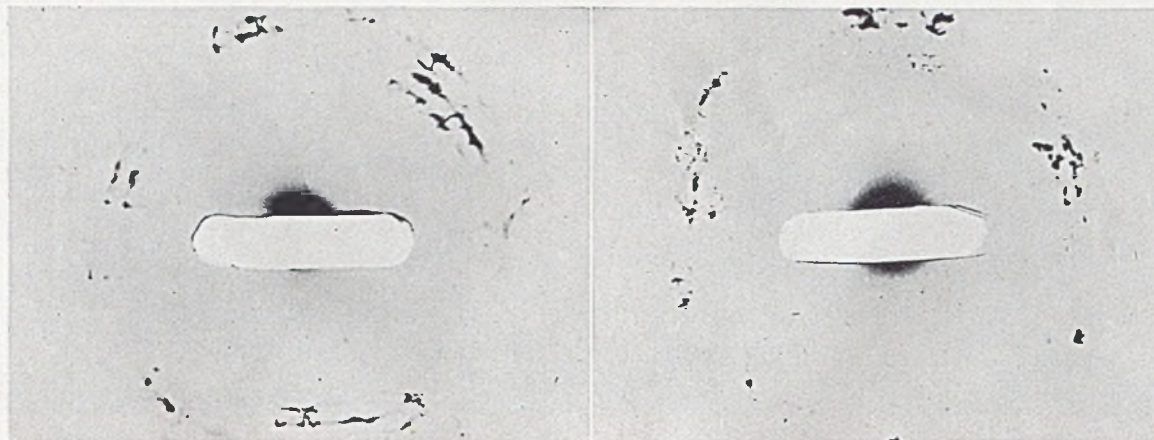
FIG. 21.—After 3·1% Extension at 21° C.   FIG. 22.—After 3·1% Extension at 200° C.

Plate XXXI.

cells inhibited slipping or these grains were of such an orientation that slipping was difficult and thus they had found it easier to deform by cell formation and movement originally at the elevated temperature.

Fig. 9 (Plate XXX) illustrates the condition of the grain shown in Figs. 3, 4, and 8 after a 3% extension at room temperature. A number of points should be noted in this photograph. The general direction of the slip across the grain is the same, a fact which opposes the view that the sub-grain structure is due to recrystallization to a fine grain-size. However, there are some slight changes of direction from cell to cell and discontinuities in the slip lines across the cell boundaries. The former observation is consistent with the idea that slight relative rotations of the cells occurred during the deformation at elevated temperature. The observation of discontinuities in the slip is rather more difficult to understand. It can only be explained by relative movement along the cell boundaries if it be assumed that only certain groups of planes across a cell were capable of slipping and, in two neighbouring cells, had moved so that these planes were out of line across the boundary. A more reasonable explanation is that probably there is a difference of level between neighbouring cells at the boundary.

Another striking feature of deformation at an elevated temperature was the " broadening " of the grain boundaries. Such broadening was probably caused partly by boundary migration and, partly, by the relative movement of contiguous grains which resulted in deep shadows. The boundary-migration type of broadening is most clearly shown in Figs. 19 and 20 (Plate XXXI), in which it will be noted that the broad " boundary " consisted of a number of very fine, wavy lines, suggesting that the tendency to migrate was not uniform along the length of the boundary. Repolishing and etching revealed that the broad boundaries showed up as a single line, but that the new boundary was very " ragged ". The cell structure appeared to be related to bumps in the grain boundary. This is illustrated by Fig. 10 (Plate XXX) which shows the same specimen as that in Fig. 6 after repolishing and re-etching, though the area photographed was different.

This observation strongly suggests that a boundary migrating into a grain with a well-developed cell structure does so by absorbing individual cells, the probability of a particular cell being absorbed depending on its orientation relative to the growing grain. Burgers [8] has drawn attention to a similar effect in a recrystallized grain growing into a polycrystalline matrix. The existence of cells parallel to the grain boundaries may be responsible for the jumps often noted during grain growth.[9] The whole question of boundary migration in grains with a cell structure is the subject of further investigation.

X-ray examination showed that the back-reflection arcs from individual grains consisted of a number of separate spots, and there can be no doubt that the cells are responsible for the spots. The overall difference in orientation of the cells in a particular grain was estimated at 2° or 3°. Fig. 22 (Plate XXXI) shows a typical diffraction pattern.

IV.—THE EFFECT OF SLOW STRAINING

Since it was considered that decreasing the rate of straining would have a similar effect to raising the temperature, a few experiments were carried out at a low rate of strain. The slow straining was carried out at two temperatures, 21° and 100° C., at the rate of 0·15% extension per hr. At both temperatures the specimens manifested a cell structure similar to that discussed in Section III. It was more marked at the higher temperature (cf. Figs. 11 and 12, Plate XXX). Cottrell and Aytekin [10] have also recently made the same observation with polycrystalline zinc at elevated temperature under creep conditions.

In both cases the X-ray diagrams showed a Debye ring upon which a number of spots were superimposed. The spots tended to be rather diffuse in both cases, quite unlike the reflections from new recrystallized grains, and microscopic examination did not reveal any trace of recrystallization. The grain boundaries were " broad " at both temperatures, though more so at the higher temperature.

Twinning was present in the specimens strained at room temperature, though to a much lesser degree than after rapid extension. It was absent at the higher temperature. Repolishing and re-etching confirmed again that the cell structure was not a surface effect and also that the boundaries had become very ragged.

A specimen having a grain-size about the same as the cell size observed in the preceding experiments was slowly strained at the same rate at 21° C. It was thought that the grains might act as cells and that the deformation would occur by relative movement of small grains, but this did not prove to be so. Slip was present but, in contrast to the coarser-grained specimens, it was confined to broad bands and was not uniformly spread over the grains. This is similar to the effect of raising the deformation temperature or lowering the rate of deformation in the case of aluminium. Many grains manifested a cell structure. Twinning was absent, thus confirming the observation made in Section III that it occurs more readily the larger the grain-size, other things being equal. Fig. 13 (Plate XXXI) illustrates the condition after 20% elongation. It will be noted that the direction of slip changes across the cell boundaries in the grain at the centre of the photograph. The same grain at 10% elongation showed straight slip and no cells, so that during the extension from 10 to 20% the grain must have " buckled " along the cell boundaries, thus giving rise to the deviations of the slip shown in Fig. 13 (Plate XXXI). A similar effect was probably responsible for the changes in slip direction shown in Fig. 7 (Plate XXX).

It may be concluded that in zinc, as in aluminium

cell formation is facilitated by decreasing the rate of strain, and that zinc resembles aluminium more closely the smaller the grain-size.

V.—THE PROCESS OF CELL-STRUCTURE FORMATION

In order to study the formation of a cell structure, a specimen was kept under observation continuously during the deformation at an elevated temperature. During the period of heating to the desired temperature before deformation, some grains showed evidence of slip which was possibly due to the anisotropy of thermal expansion.[11] In a few grains, too, a faint cell structure was observed which may be reasonably attributed to the same cause.

Fig. 14 (Plate XXXI) is a photograph of a typical area of a specimen taken at 220° C. before deformation. During the initial stages of deformation at 220° C. slip occurred in the majority of grains under observation. However, in the very early stages of deformation, some grains developed a cell structure before slip was observed. An example of such a grain is shown on the extreme right of Figs. 15–18 (Plate XXXI). The grains which exhibited slip in the early stages did not begin to show cells very clearly until the extension had reached about 5% (Fig. 16, Plate XXXI).

It will be noted that, accompanying the increasing deformation, the surface of the grains became tilted at different angles from cell to cell, and slight movements of the cells and changes in shape occurred. This latter is most strikingly shown near a grain boundary at high magnification in Figs. 19 and 20 (Plate XXXI), corresponding to 10 and 15% elongation respectively. In many instances cell structures apparently began as slight " puckering " near the grain boundary, possibly because of the greater degree of deformation along the grain boundaries.

The most important observation made during the deformation was that the initiation and subsequent developments in the cell structure occurred continuously with increasing extension, as though the formation and movement of the cell structure were a direct result of the deformation and not caused by a two-stage process of deformation followed by recovery.

An example of the broadening of a grain boundary by migration is shown in Figs. 19 and 20 (Plate XXXI), and illustrates quite clearly the raggedness of the migrating " front ".

VI.—DISCUSSION

The evidence adduced in the present paper shows clearly that, on deformation at elevated temperature or, if the strain rate is slow enough, at room temperature, the grains of polycrystalline zinc break down into a number of distinct elements or " cells ", indicating that zinc behaves in a comparable way to that of aluminium deformed under similar conditions. However, zinc differs from aluminium in that the slip-line spacing and " thickness " was found to be invariant over the ranges of strain rate and temperature considered in those experiments, and the cells show up in a much more striking way directly on deformation. The X-ray diagrams were similar to those obtained from aluminium.

The question has been raised in discussion of the work of Wood and his colleagues [12, 13] as to whether or not the cell structure obtained on deformation at elevated temperature is due to polygonization. Before embarking on a discussion of this matter, it is well to be clear as to the nature of the experimental evidence in both cases.

Considering in the first place the X-ray diffraction data, Cahn observed spotty asterisms in severely bent single crystals on annealing at a temperature very close to the melting point. Similar effects have been obtained by a number of other workers under conditions of deformation at room temperature and subsequent heating at an elevated temperature. Turning now to the question of cell formation, Wood and his co-workers, as well as a number of others, have found that discontinuous Debye arcs or asterisms can be obtained directly upon deformation at elevated temperatures, or as in the present work, at room temperature, if the deformation rate is low enough. Moreover, in the experience of the author, the spots on the X-ray diagrams indicate that the size and degree of perfection of the elements within the grains are greater when they are produced directly by deformation at elevated temperatures. Thus the X-ray evidence indicates some similarity between the final product in both cases.

On the other hand, the microscopic appearance is very different. Cahn [14] has shown that recovery after cold deformation produces long narrow elements, the boundaries of which run perpendicular to the slip planes, but the cells shown in the present paper have quite a different shape, and their boundaries bear no particular relation to the slip.

These, then, are the two major experimental facts upon which the concepts of " polygonization " and " cell formation " are based; one set of data shows a similarity between them and another a very striking difference. It is contended that the only experimental evidence for saying that cell formation is due to polygonization is the similarity of the X-ray diagrams.

Apart from this, a number of observations recorded in the previous section cast some doubt on the interpretation of cell formation as resulting from the same mechanism as polygonization, as interpreted by Cahn. In the first place, cells formed in a number of grains as soon as deformation began and developed strongly, though no slip was observed in the grains. If no slipping occurred, then the explanation of cell formation in terms of Cahn's theory of polygonization is improbable. Of course it may be argued that slip occurred, but not to an observable extent. If this were so, it is difficult to understand why polygonization did not occur in those grains showing much more marked slip. This can only mean that either there is a critical deformation at which " poly-

gonization " occurs most readily, or the orientation of the grains was such that their active slip plane lay parallel, or nearly parallel, to the specimen surface.

If the former were so in the present case, it would be expected that those grains which showed evidence of deformation by slip initially would not form a cell structure after a much greater degree of deformation, but in fact such a sequence was observed. If, on the other hand, it is suggested that slip was not visible in certain grains because the slip plane was parallel to the specimen surface, it is difficult to understand why slip appeared when the specimen was repolished and re-etched and then strained at room temperature.

These facts strongly suggest that slip is not a necessary prerequisite to the formation of cells, and therefore that the mechanism of their formation cannot be explained along the lines of Cahn's theory of polygonization.

An alternative view, which does not encounter these difficulties, is that at elevated temperatures some of the grains break down directly, on deformation, into a number of fragments identical with those shown in the photomicrographs, and these are responsible for the spottiness of the X-ray diagrams. It is suggested that such a direct breakdown occurs because at elevated temperature some grains find it easier to deform by the formation and relative movement of cells than by slip. In the grains which showed slip initially, it must be assumed that after a certain amount of deformation further changes in shape could not be provided by slipping and so cell formation and movement occurred. Of course, the problem remains as to why cell formation is easier than slipping at elevated temperature.

The basic difference between the two points of view can be most clearly seen from a consideration of the atomic movements involved in each theory. The theory put forward in the present paper suggests that deformation at elevated temperatures causes the movement of atoms from their equilibrium lattice positions, but that the thermal energy available is sufficient to enable them to find new equilibrium positions continuously and spontaneously during the deformation; whereas the explanation of cell formation based on Cahn's theory of polygonization assumes a wholesale disordering of a large number of atoms, followed by a wholesale reordering. The latter theory therefore in effect presupposes that deformation at elevated temperatures involves the simple arithmetical addition of the effect of temperature and the effect of deformation, which is very different from the idea that, on deforming at elevated temperatures, the two factors combine to produce a different mode of deformation.

ACKNOWLEDGEMENTS

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THE SIGMA PHASE IN BINARY ALLOYS OF THE TRANSITION ELEMENTS *

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SYNOPSIS

The appearance of the sigma phase in alloys of elements of the First Long Period can be satisfactorily accounted for if it is assumed that the incidence of this phase is determined by a critical excess of electrons in bond orbitals over vacancies in the $3d$ atomic orbitals. In the alloys of cobalt and iron with chromium; iron, nickel, and cobalt with vanadium; and chromium and vanadium with manganese, predictions of the boundary of the homogeneous sigma-phase region can be made on the basis of this hypothesis which accord well with experimental data where these are available. The fact that the sigma field does not include the equiatomic composition in certain alloy systems is also explained. It is probable that sigma phases occur in alloys between transition elements of the other long periods.

I.—INTRODUCTION

IT was first shown by Bain and Griffiths [1] that a hard, brittle, and non-magnetic phase was formed in iron–chromium alloys of composition close to an equiatomic ratio, when the alloys were slowly cooled or annealed for long periods of time at temperatures below 950° C. The existence of this " sigma " phase was denied by other workers, notably by Adcock,[2] but its presence has since been adequately confirmed by Wever and Jellinghaus,[3] by Eriksson,[4] and by Bradley and Goldschmidt.[5] The extent of the sigma-phase region in equilibrium conditions in high-purity iron–chromium alloys had been determined by Cook and Jones,[6] whose diagram is reproduced in Fig. 1. The extension of the sigma-phase field into various ternary alloys containing iron and chromium has been investigated by many workers, and much study has also been devoted to the important effects which the formation of this phase has on the properties of industrial alloys, particularly of the austenitic stainless steels.

Until recently it was assumed that this phase was confined to alloys that contained both iron and chromium. In 1948, however, Sully and Heal [7] showed that a phase isomorphous with the sigma phase existed in the cobalt–chromium system. The most recently determined equilibrium diagram of this system, due to Elsea, Westerman, and Manning,[8] is shown in Fig. 2. The phase isomorphous with sigma is the gamma phase, the range of homogeneity of which is indicated to be from 53·5 to 58 wt.-% (56·6–61·0 at.-%) chromium.

Andrews [9] has shown that another isomorphous phase exists in the iron–vanadium system, the original investigation of which was made by Wever and Jellinghaus [10] (Fig. 3).

Goldschmidt [11] found that the phase based on FeMo in the iron–molybdenum system also has the sigma structure. In this system, however, the phase is

stable only at temperatures above 1180° C. and decomposes to α-molybdenum and the ϵ-Fe$_3$Mo$_2$ phase at lower temperatures. This behaviour is markedly different from that in the iron–chromium, cobalt–chromium, and iron–vanadium systems, in which sigma is stable at relatively low temperatures and transforms below the melting point of the alloys in which it occurs.
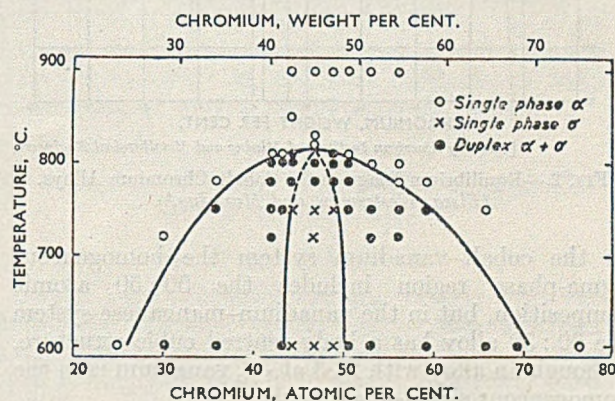
FIG. 1.—Sigma Phase in Iron–Chromium Alloys. (Cook and Jones.[6])

Recently, the sigma phase has been encountered in four additional alloy systems. It has been found in nickel–vanadium and in cobalt–vanadium alloys by Pietrokowsky and Duwez [12] and by Pearson, Christian, and Hume-Rothery,[13] and by the latter investigators in chromium–manganese and vanadium–manganese. None of these systems has yet been examined in detail, but Pearson, Christian, and Hume-Rothery give the range of homogeneity of the sigma phase as 55 to more than 65 at.-% vanadium in the nickel–vanadium system, and from approximately 72 to 83 at.-% manganese in the chromium–manganese system.

FIG. 2.—Equilibrium Diagram for Cobalt–Chromium Alloys.
(*Elsea, Westerman, and Manning.*[8])
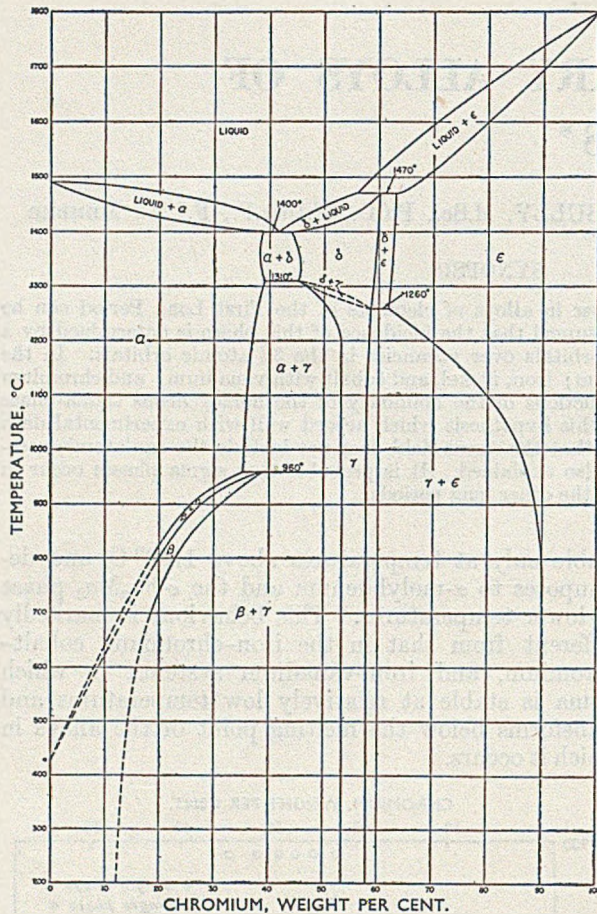
[*Courtesy American Institute of Mining and Metallurgical Engineers.*]

In the cobalt–vanadium system the homogeneous sigma-phase region includes the 50 : 50 atomic composition, but in the vanadium–manganese system the 50 : 50 alloy has a body-centred cubic structure, although an alloy with 24·3 at.-% vanadium is in the homogeneous sigma-phase field.

II.—ELECTRON COMPOUNDS

The existence of isomorphous phases of characteristic structure in different alloy systems at compositions which differ from one system to another has been satisfactorily explained by Hume-Rothery and his co-workers [14] on the basis that the phase appears at a critical ratio of valency electrons to atoms. Thus, the body-centred cubic β phase of the alloys of copper, silver, and gold with B sub-group and certain other elements occurs at a critical electron : atom ratio of 3 : 2. Similarly, it has been shown in the same alloys that γ-brass structures occur at electron : atom ratios of 21 : 13 and close-packed hexagonal structures at a ratio of 7 : 4. Such phases have come to be known as " electron compounds ".

When the sigma phase was known only in the iron–chromium system, in which its range of homogeneity

includes the equiatomic composition, there was no reason to suppose that similar considerations applied to the formation of this phase. Indeed, Andrews [9] has suggested that its formation depends upon the onset of an ordered structure. However, it is now clear that in the systems cobalt–chromium, nickel–vanadium, chromium–manganese, and chromium–vanadium the range of homogeneity of the sigma phase does not include the equiatomic composition, and it is worth re-examining the suggestion, first made by Sully and Heal,[7] that the sigma phase is an electron compound.

III.—ELECTRONIC STRUCTURE OF TRANSITION METALS

The chemical behaviour of transition metals is satisfactorily explained by the quantum theory, on the basis that the outermost quantum shell is filled to a constant level, while the next inner shell is incompletely full and is filled progressively as the atomic number increases within the transition group of elements. Thus, the electronic structure of the free atoms of the elements from potassium to copper is as set out in Table I.

Although the behaviour of free atoms is satisfactorily accounted for in this way, there are certain properties of the transition elements that are not explicable on this basis, notably the high bonding energy, high resistivity and specific heat, and the appearance of ferromagnetism in iron, cobalt, and nickel.
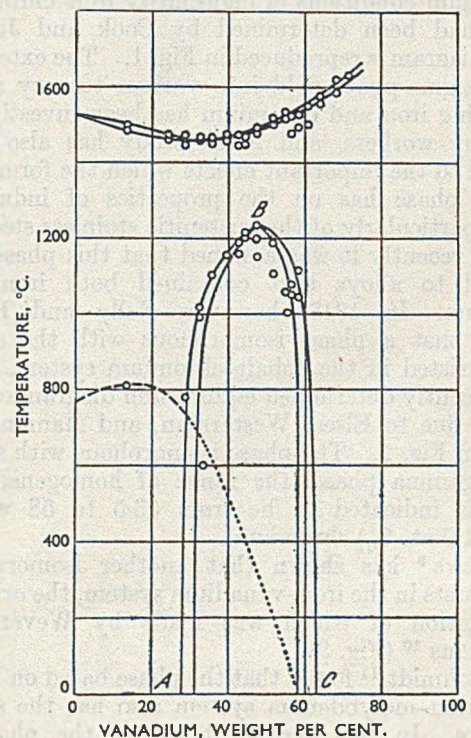
[*Courtesy K. W. Institut für Eisenforschung.*]

FIG. 3.—Sigma Phase in Iron–Vanadium Alloys. (*Wever and Jellinghaus.*[10])

It was first proposed by Mott and Jones [15] that the numbers of electrons in the $3d$ and $4s$ bands of the transitional First Long Period elements were statistically non-integral, the average electron density being such that there were vacancies in the $3d$ band; the

TABLE I.—*Electronic Structure of Free Atoms of Transition Metals*

Element	Atomic Number	Number of Electrons for Indicated Quantum Number and Sub-Group						
		1	2s	2p	3s	3p	3d	4s
K . .	19	2	2	6	2	6	...	1
Ca . .	20	2	2	6	2	6	...	2
Sc . .	21	2	2	6	2	6	1	2
Ti . .	22	2	2	6	2	6	2	2
V . .	23	2	2	6	2	6	3	2
Cr . .	24	2	2	6	2	6	5	1
Mn . .	25	2	2	6	2	6	5	2
Fe . .	26	2	2	6	2	6	6	2
Co . .	27	2	2	6	2	6	7	2
Ni . .	28	2	2	6	2	6	8	2

number of unoccupied states was equal to the saturation intensity of magnetization in Bohr magnetons per atom, these values being 2·2 for iron, 1·7 for cobalt, and 0·6 for nickel.

This theory has been extended by Pauling,[16] who considers that bonding in the transition metals from chromium to nickel inclusive is due to both $4s$ and $3d$ electrons, but that the $3d$ electron states fall into two distinct groups. Of these 2·44 electron states or orbitals do not participate in bonding. Pauling calls these atomic orbitals. Two electrons, with opposite spins, can be accommodated in each orbital, so that 4·88 electrons can be accommodated in Pauling's atomic orbitals. The remaining 5·12 electrons, which make up the permitted maximum number of 10 in the $3d$ band, are accommodated in 2·66 " bond " orbitals, and electrons in these orbitals together with the $4s$

TABLE II.—*Electronic Structure Based on Pauling's Theory*

Element	No. of Electrons in Bonding Orbitals		No. of Electrons in 3d Atomic Orbitals			Vacancies in Atomic Orbitals
	3d	4s	Paired	Unpaired	Total	
Cr . .	5·12	0·66	0	0·22	0·22	4·66
Mn . .	5·12	0·66	0	1·22	1·22	3·66
Fe . .	5·12	0·66	0	2·22	2·22	2·66
Co . .	5·12	0·66	1·56	1·66	3·22	1·66
Ni . .	5·12	0·66	3·56	0·66	4·22	0·66

electrons are available for bonding. The electrons in the $3d$ atomic orbitals will remain unpaired as long as possible, and the saturation magnetic moment may be equated with the number of unpaired electrons in the $3d$ atomic orbitals.

The electronic structures predicted by Pauling for the elements from chromium to nickel are given in Table II.

On Pauling's theory, vanadium is regarded as having 5 and titanium 4 electrons available for bond formation, i.e. they have no electrons in atomic orbitals.

IV.—ALLOYING BEHAVIOUR OF TRANSITION METALS

In most of their alloys with non-transition metals, the transition metals chromium to nickel behave as if they have either zero or negative valency. For example, Ekman [17] showed that in the iron–zinc and nickel–zinc systems γ-brass structures occur at Fe_5Zn_{21} and Ni_5Zn_{21}, respectively. This conforms to the 21 : 13 electron : atom ratio, if iron and nickel behave as zero-valent, while zinc has its normal metallic valency of 2.

On the other hand, in some other systems these elements behave as if they have a negative valency, i.e. they accept electrons from the structure as a whole and contribute none themselves. Raynor [18, 19] and his co-workers, in studying the alloys of aluminium with the transition metals iron, cobalt, and nickel, have shown that, if it is assumed that the transition-metal atoms absorb electrons to the full extent permitted by the vacancies in the $3d$ atomic orbitals, the appearance of compounds of the type X_2Al_9, where X represents one or more transition metals, is characterized by an excess electron : atom ratio of 2·06–2·28. It is apparently assumed that the transition elements themselves contribute no electrons to the structure, i.e. that the electrons in Pauling's bond orbitals play no part in determining the structure.

V.—ALLOYS OF THE TRANSITION METALS WITH EACH OTHER

1. Mode of Bonding

In the alloys of the transition metals with one another existing theory permits of no close definition of the detailed mode of cohesive bonding. If it is assumed that all the electrons in Pauling's bond orbitals are available as bond electrons, then the ratio of bond electrons to atoms remains constant at a value of 5·78 for all the alloys of chromium, manganese, iron, cobalt, and nickel with each other. If it is also assumed that vanadium contributes five electrons to the structure, then in the alloys of the transition metals with vanadium the electron : atom ratio can vary only between limits of 5·0 and 5·78. Obviously, on this basis it is not possible to explain the formation of the sigma phase by a critical electron : atom ratio. Another possibility which may be considered, however, is that under certain conditions an exchange of electrons can occur between bond and atomic orbitals, so that the latter are filled to the fullest permitted extent of 4·88 electrons per atom. Since the number of vacancies varies from one transition metal to the next, the ratio of atoms to electrons in excess of those required to fill the $3d$ vacancies can then vary considerably with composition.

In considering the alloys of vanadium with the other transition metals, the simplest assumption is that it contributes 5 bonding electrons to the structure as a whole, but an additional assumption has been made that, like the following elements
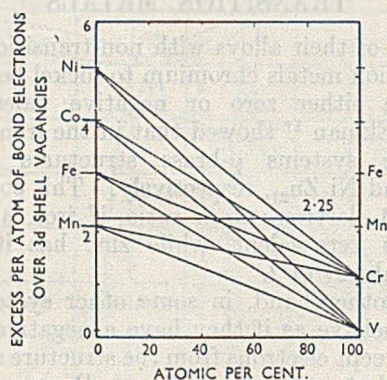
Fig. 4.—Excess Electron : Atom Ratios for Alloys of Elements from Vanadium to Nickel, Assuming Participation of $4s$ Electrons in Elements Chromium to Nickel.

chromium to nickel, it also has 2·44 atomic orbitals accommodating 4·88 electrons, which are normally empty.

With these assumptions and with the further assumption that in the sigma-phase structure electrons are absorbed to the full extent permitted by the vacancies in the $3d$ atomic orbitals, the ratio to atoms of the excess of bond electrons over $3d$ vacancies is found to vary considerably from system to system.

In considering the number of bonding electrons available, the simplest assumption is that both $3d$ and $4s$ electrons contribute to the structure, so that

TABLE III.—*Sigma-Phase Boundary in Transition-Metal Alloys*

System	Observed Boundary, At.-%	Range of Homogeneity, At.-%	Predicted Boundary, At.-%		
			$4s$ Electrons Contributing (Ratio 2·25)	$4s$ Electrons Not Contributing (Ratio 1·7)	$4s$ Electrons Contributing (Ratio 1·7)
Fe–Cr	43·5 Cr	43·5–50·0 Cr	43 Cr	38 Cr	...
Co–Cr	56·6 Cr	56·6–61 Cr	62 Cr	59 Cr	...
Ni–Cr	Not	observed	72 Cr	69 Cr	...
Fe–V	30 * V	30–60 * V	29 V	32 V	...
Co–V	40–45 V	40–45–>54 V	47 V	52 V	...
Ni–V	55–60 V	55–>65 V	57 V	63 V	...
Mn–Cr	17 Cr	17–28 Cr	...	...	24 Cr
Mn–V	?	Includes 24·3 V	...	...	20 V

* Values not precisely defined.   See discussion below.

there are 5·12 $3d$ electrons and 0·66 $4s$ electrons, or 5·78 electrons in all, available per atom of transition metal. Fig. 4 has been constructed on this model and shows the variation with composition of the ratio to atoms of electrons, in excess of those required to fill $3d$ atomic orbitals, for alloys of nickel, cobalt, and iron with manganese, chromium, and vanadium. It is

found that an excess of electron : atom ratio of 2·25 accounts for the occurrence of sigma phase in the alloys with chromium and vanadium and also predicts fairly successfully the compositions at which the sigma phase appears in these systems.[20] The compositions are tabulated in the fourth column of Table III. The agreement with the experimentally determined boundary is in most cases quite satisfactory, and moreover the theory successfully accounts for the observed fact that the range of homogeneity of the phase does not include the equiatomic ratio in the systems cobalt–chromium and nickel–vanadium.

Although at first sight an excess electron : atom ratio of 2·25 appears to account satisfactorily for the appearance of the sigma phase in the systems shown in Table III, recent information indicates that the model chosen is not correct in detail. The Debye-Scherrer X-ray pattern of the sigma phase is characterized by a group of strong lines close together at low Bragg angles, and lends itself fairly readily to an approximate determination of the volume of the first
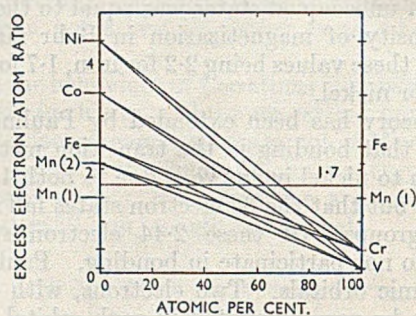
Fig. 5.—Excess Electron : Atom Ratios for Alloys of Elements from Vanadium to Nickel, Assuming no Participation of $4s$ Electrons in Elements Chromium to Nickel except for Manganese (2).

Brillouin zone. This enables an assessment to be made of the number of electrons that can be accommodated in the zone and consequently of the electron : atom ratio for the structure. In this way Douglas [21] has derived an approximate value of 1·72 for the sigma phase in cobalt–chromium alloys, while the author's colleagues Heal and Silcock [22] confirm this value and obtain an approximate value of 1·7 for the sigma phase in nickel–vanadium alloys at a composition of 62·5 at.-% vanadium.

These results show that the proposed electron : atom ratio of 2·25 cannot be correct and that, if the structure is an electron compound, one of the assumptions made must also be incorrect. A consideration of the results suggests that the incorrect assumption may be that the 0·66 $4s$ electrons per atom play a part in the bonding process. If, instead, it is assumed that only the $3d$ electrons are involved, it is found that an excess electron : atom ratio of about 1·7 is appropriate to the sigma-phase composition ranges. Fig. 5 has been compiled by ignoring the 0·66 $4s$ electrons per atom for the elements chromium to nickel and considering only the excess of $3d$ bond electrons

(5·12 per atom) over 3*d* vacancies. The same assumptions as before are made about vanadium (5 electrons per atom and 4·88 3*d* vacancies). The predicted composition of the sigma-phase boundary for an excess electron : atom ratio of 1·7 is given for the various systems in the fifth column of Table III. A detailed discussion of the degree of agreement in the various systems is given below.

2. ALLOYS WITH CHROMIUM

(a) *Iron–Chromium*

According to Cook and Jones [6] (Fig. 1) the lower-chromium boundary of the sigma-phase region at 600° C. is at 43–44 at.-% chromium. The predicted value for an electron : atom ratio of 1·7, ignoring the 4*s* electrons, is 38%. In view of the difficulty of achieving equilibrium conditions in alloys of this type, the agreement may be regarded as satisfactory. The homogeneous single-phase region extends to 50 at.-% chromium which corresponds to an excess electron : atom ratio of 1·46. Further investigation of these alloys, and those of the systems discussed below, is required in order to determine whether the structure can tolerate this degree of electron deficit or whether the electron : atom ratio is maintained at a higher level by the existence of a lattice deficient in chromium atoms as the chromium content increases across the range of homogeneity of the phase. An example of a lattice of this type, in which the electron : atom ratio is maintained approximately constant, is the Mn_3SiAl_9 structure referred to by Robinson.[23]

(b) *Cobalt–Chromium*

The lower-chromium boundary of the sigma region, according to Elsea, Westerman, and Manning [8] (Fig. 2) is at 56·6 at.-% chromium. The predicted boundary is 59 at.-% chromium, which agrees quite well and lies within the observed range of 56·6–61·0 at.-% chromium.

(c) *Nickel–Chromium*

The theory predicts the appearance of the sigma phase in the nickel–chromium system at about 70 at.-% chromium. In fact, the sigma phase has never been observed in this system, which has been the subject of several investigations. Beck and Manly [24] have, however, noted that in the cobalt–chromium–nickel alloys the sigma phase persists for a considerable way towards the nickel–chromium boundary, and they infer that a tendency may exist to form a sigma phase in the nickel–chromium alloys. A possible reason for the non-appearance of the phase may be that its upper temperature limit of stability, which varies from system to system, is below about 600° C., so that transformations in these alloys do not take place within experimentally reasonable periods of time. It may be noted, although it may be fortuitous, that there is some correlation in the alloys with chromium, between the Curie temperature of the ferromagnetic component of the system and the

upper temperature limit of stability of the sigma phase. In the iron–chromium alloys, the upper temperature limit, according to Cook and Jones,[6] is at 800°–820° C., the Curie point of iron being 770° C.; while in cobalt–chromium alloys the sigma phase persists to 1250° C., the Curie point of cobalt being 1150° C. If this apparent correlation has any real significance, the upper temperature limit of stability of sigma may be related to the Curie point of nickel, which is only 360° C. In the alloys with vanadium, referred to below, no such considerations apply, however, the sigma phase being formed in nickel–vanadium alloys quite readily at the appropriate compositions at temperatures of the order of 700°–800° C.

3. ALLOYS WITH VANADIUM

(a) *Iron–Vanadium*

The predicted boundary of the sigma region at the lower-vanadium end is 32 at.-% vanadium. Unfortunately, the range of homogeneity of the phase in this system is not reliably established. The diagram published by Wever and Jellinghaus [10] (Fig. 3) suggests a lower limit of just over 30% vanadium and a wide range of homogeneity up to about 58% by weight (60 at.-%). This at first sight indicates good agreement with the predicted value, but a closer examination of the data presented by Wever and Jellinghaus suggests that the range of homogeneity of the sigma phase may be narrower than that given by the diagram. The main evidence in favour of a wide range of homogeneity is provided by lattice-spacing measurements of the sigma phase which are reproduced in Table IV.

TABLE IV.—*Lattice Spacings for Two Lines of X-ray Spectrum of Sigma Phase in Fe–V Alloys (from Wever and Jellinghaus [10]).*

Vanadium Content		Lattice Spacing, Å.	
Wt.-%	At.-%	d_1	d_2
38·4	40·5	1·980	1·898
46·4	48·5	1·980	1·907
47·7	50·0	1·993	1·909
52·2	54·6	1·995	1·913
63·6	65·7	2·013	1·926

In a two-phase region the compositions of the separating phases, and consequently the lattice spacings, should be constant for a particular temperature of equilibrium. The increase of spacing over the above range of composition is interpreted by Wever and Jellinghaus as indicating that the sigma phase is homogeneous over most of the range. This conclusion may, however, be invalidated by the fact that the alloys on which these measurements were conducted were slowly cooled from 1400° C. and were not brought into equilibrium at a particular temperature by long isothermal heat-treatments. It would probably be unwise to draw any conclusion from Wever

and Jellinghaus's work, other than that it indicates a fairly wide range of homogeneity for the sigma phase, the precise limits of which must await further investigation. The predicted value for the low-vanadium boundary of 32 at.-% may, therefore, be too low in vanadium, but the extent of the disagreement cannot be assessed until the diagram is clarified. If it is assumed that the other boundary of the sigma-phase region is at 60 at.-% vanadium, then the sigma structure in these alloys can tolerate a much wider range of electron deficit than can the chromium alloys.

(b) *Cobalt–Vanadium*

The predicted boundary in this system is at 52 at.-% vanadium. There is no published work which enables the limits of the sigma-phase region to be precisely defined. Duwez and Baen [25] have shown that an alloy of the equiatomic composition lies in the homogeneous sigma field, and this is confirmed by Pearson, Christian, and Hume-Rothery.[13] The author's colleague, Miss J. M. Silcock, has carried out an X-ray investigation of some binary cobalt–vanadium alloys which were heated for long periods at 600° C. and at 700° C. The range of alloys examined was not sufficiently wide to enable the full extent of the homogeneous sigma-phase field to be determined, but the lower limit was found to be at an approximate value of 40–45 at.-% vanadium. The upper limit was greater than 54 at.-% vanadium. The predicted value thus lies to the vanadium-rich side of the correct boundary, but within the homogeneous sigma field.

(c) *Nickel–Vanadium*

The predicted boundary in this system is at 63 at.-% vanadium. Pearson, Christian, and Hume-Rothery [13] give the sigma range as 55 to more than 65 at.-% vanadium, and Duwez and Baen [25] note that the equiatomic composition lies in a two-phase region. The author's colleague, Miss J. M. Silcock, has confirmed that the lower limit is in the range 55–60 at.-% vanadium. The predicted value is not seriously in error and accords with the fact that the sigma region lies wholly to the vanadium-rich side of the stoichiometric composition NiV.

4. ALLOYS OF MANGANESE

If manganese shows the same behaviour as is suggested for the other metals in the same group, from chromium to nickel, the excess electron : atom ratio for the element would be $5 \cdot 12 - 3 \cdot 66 = 1 \cdot 46$. If this were so, it would be expected that alloys at the manganese-rich end of the binary systems with iron, cobalt, and nickel, would tend to a sigma structure. As far as is known, however, this is not the case, although Bradley and Goldschmidt [5] have stated that similarities exist between the X-ray pattern of the sigma phase and that of α-manganese. Instead, the sigma phase appears in the manganese–chromium and manganese–vanadium systems in which the excess

electron : atom ratio computed in the manner described above is appreciably less than $1 \cdot 7$. These facts suggest that manganese can behave in a way dissimilar from the other metals of the same group. Anomalous behaviour by manganese has been noted in other alloy systems. For example, in copper–aluminium ternary alloys the transition metals except manganese all behave as if they have a valency $\leqslant 1$, but Hume-Rothery [26] has shown that at concentrations less than 25% manganese behaves as a divalent element. A possible reason for this is that in manganese the 4s electrons participate in bonding, so that this element can have an excess electron : atom ratio of $5 \cdot 12 + 0 \cdot 66 - 3 \cdot 66 = 2 \cdot 12$. In this case, as will be seen from Fig. 5, sigma-phase formation at an electron : atom ratio of $1 \cdot 7$ is predictable in the systems manganese–chromium and manganese–vanadium.

In manganese–chromium alloys the range of homogeneity of the phase is given by Pearson, Christian, and Hume-Rothery [13] as 83–72 at.-% manganese (17–28 at.-% chromium). The predicted value at an electron : atom ratio of $1 \cdot 7$, assigning an excess electron : atom ratio of $2 \cdot 12$ to manganese, is 24 at.-% chromium, which lies within the observed range although on the chromium-rich side of the observed boundary.

In manganese–vanadium alloys the only information available at the present time is that an alloy with $24 \cdot 3$ at.-% vanadium is a sigma phase.[13] The boundary predicted on the above basis is at 20 at.-% vanadium, which is consistent with this observation.

VI.—DISCUSSION OF RESULTS

The above approach to the problem of sigma-phase formation in the transition-metal alloys is somewhat speculative. It has, however, been shown on the basis of Pauling's theory of the electronic structure of the transition metals, that a number of predictions which conform to observed experimental facts can be made, if the factor that determines the appearance of the sigma phase is a critical excess of bond electrons over valencies in 3d atomic orbitals. Predictions can be made almost equally successfully with an excess electron : atom ratio of $2 \cdot 25$ and on the assumption that 4s electrons in the elements chromium to nickel contribute to bonding, or with an excess electron : atom ratio of $1 \cdot 7$ assuming no contribution from the 4s electrons, except in the case of manganese. Determinations of the volume of the first Brillouin zone in cobalt–chromium and nickel–vanadium alloys appear to favour the second model.

On this basis it is possible to explain the non-appearance of sigma phase in alloys of iron, nickel, and cobalt with each other. It is also possible to account for its appearance at approximately the correct composition in the systems iron–chromium, cobalt–chromium, iron–vanadium, cobalt–vanadium, nickel–vanadium, manganese–chromium, and manganese–vanadium. In particular, this hypothesis successfully explains the observed facts that the

homogeneous sigma-phase field does not include the equiatomic composition in the systems cobalt–chromium, nickel–vanadium, manganese–chromium, and manganese–vanadium. Such consistent agreement with experiment is hardly likely to be fortuitous and strongly suggests that the sigma phase is a type of electron compound.

Little is known about the detailed electronic structure of the *d* and *s* states in the transition elements of the other long periods. For this reason no attempt is made to account for Goldschmidt's [11] observation that a sigma phase is formed in alloys of iron and molybdenum, and the observation of Kasper, Decker, and Belanger [27] that it occurs in manganese–molybdenum alloys. Obviously, however, similar considerations could apply to these alloys, and there is thus a strong possibility that similar phases may be formed in alloys between transitional metals of long periods other than the first. It is to be noted that, according to a recent publication,[28] the structure of β-uranium is also isomorphous with the sigma phase.

ACKNOWLEDGEMENT

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SOME METALLOGRAPHIC OBSERVATIONS ON THE FATIGUE OF METALS *

By P. J. E. FORSYTH,† A.I.M., MEMBER

SYNOPSIS

A metallographic investigation of the effects of cyclic stresses on the microstructure of an aluminium–$\frac{1}{2}$% silver alloy has shown that factors other than simple slip are involved in the mechanism of fatigue at room temperature.

There is evidence of a recovery process associated with the formation of deformation bands and crystallites during fatiguing of the metal, and it is suggested that the observed anomalies in the effect of stress concentration on crack progress are the result of crystallite formation at the roots of the cracks.

I.—INTRODUCTION

EWING and Humfrey [1] were among the first to make a study of the microstructural changes brought about by cyclic stressing. Their observations led them to suggest an "attrition" theory of fatigue in which the repeated applications of an unsafe stress produced repeated slip, resulting in attrition of the slip planes; this attrition eventually led to cracking and ultimate failure of the metal. Gough [2] examined metallographically several different metals, mainly in the form of single crystals, and concluded that the process of fatigue is one of strain-hardening by slip on the operative glide planes of the crystals until cracking occurs on these planes. From this he concluded that the strain-hardening process is identical with that of a crystal subjected to a static stress. He resolved the problem of fatigue cracking into one of cohesion between the metal atoms which will only be solved when a satisfactory theory of static rupture strength has been evolved. Gough also suggested that slip is accompanied by fragmentation of the original metal crystal into small crystallites in the region of the planes where slip has occurred, and that these crystallites suffer a slight rotation or re-orientation from the original crystal direction.

Theories that have so far been advanced to explain the phenomenon of fatigue in metals assume that during cyclic stressing a process of work-hardening takes place on the crystallographic slip planes. If the stress in the strain-hardened region eventually exceeds the rupture strength of the material, then cracking occurs.[3] No distinction has in the past been drawn between fatigue at high and at low stresses, since it has been assumed that the deformation is associated with slip alone; the degree of stressing would then control only the rate of work-hardening and the fatigue life of the metal. The characteristic shape of the fatigue curve for a metal might suggest that two mechanisms were in operation, but the complex nature of the strain-hardening process has provided an alternative explanation.

The subject of deformation under static stress has been investigated more thoroughly than that of deformation under cyclic stresses. The classical idea of plastic deformation by glide on the operative crystallographic planes has long been known. Tammann [4] considered that the only possible means of deformation was by slip or twinning, but more recent work has shown that other forms of deformation may occur. Elam [5] has shown that deformation in
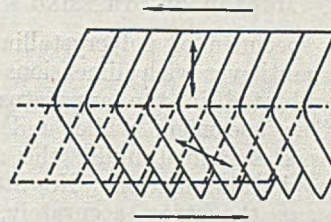
FIG. 1.—A Schematic Representation of a Lattice where Kinking has Occurred. The dotted lines represent the original lattice position and the continuous lines the new "kinked" or re-orientated material. (*After Orowan.*[6])

β-brass does not take place by slip on any definite crystal plane, but is brought about by movements of a complicated nature. These movements produce bands in the crystal, some of which have an orientation different from that of the original crystal and others which are thought to be layers of more highly strained material. A mechanism of deformation termed "kinking" has been described by Orowan,[6] who considers that there are many indications that deformation bands in metals, hitherto considered to be glide or twin bands, may be in reality "kink" bands. The mechanism of kinking is illustrated diagrammatically in Fig. 1. Honeycombe [7] has shown that kinking can produce deformation bands

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in single crystals of aluminium with as little as 1% extension. These bands are described as layers of varying orientation which are joined by narrow bent regions.

Flexural glide, either with or without accompanying relaxation in the form of polygonization, can also be considered as a possible form of deformation. The result of either form of deformation is a local variation in orientation, whereas pure glide causes no such change. These variations may be very small, and hence difficult to detect. Polygonization need not necessarily result in equiaxial crystallites. The term is used to describe the recovery of any form of lattice curvature. If the curvature is in one direction only, as in the case of kink bands, the resulting boundaries will be straight.

In undertaking the work described in the present paper, it was considered that developments in metallographic technique over recent years might bring to light features associated with the deformation of metals during cyclic stressing which have possibly escaped observation in the work previously reported. It is felt that the observations which have been recorded are of sufficient importance to justify a reconsideration of the present theories of fatigue by those working in the field of metal physics.

II.—EXPERIMENTAL TECHNIQUE

1. Method of Stressing

The fatigue specimen of polycrystalline metal in the form of a cantilever with dimensions $64 \times 8 \times 2$ mm., was vibrated in the apparatus described previously.[8] The specimen, while undergoing the fatigue test, was examined under the microscope, using stroboscopic illumination. No attempt was made to measure the stress accurately, since the investigation was concerned with the behaviour of the individual grains, for which the stress system would be complex. In the present paper the expression " high stress " is used where the maximum stress of the cycle is above the stress level of the " knee " of the S/N curve, and " low stress " is used where the maximum stress of the cycle is below the stress level of the " knee " of the curve.

2. Materials Examined

Most of the tests have been made on a high-purity aluminium–$\frac{1}{2}$% silver alloy, although a few were made later on a commercially pure copper and an Armco iron. The aluminium–silver alloy was chosen because the solute causes negligible hardening as a result of the similarity in size of the aluminium and silver atoms, and because the solid-solubility range of silver in aluminium is sufficiently extensive to permit the study of the fatigue behaviour of single-phase and duplex structures. Unless otherwise stated, the observations and illustrations refer to the aluminium–$\frac{1}{2}$% silver alloy.

3. Preparation of Specimens

(i) *Aluminium–$\frac{1}{2}$% Silver Alloy:* This was cold rolled to 2 mm. from $\frac{1}{2}$-in. dia. extruded bar, machined, and then heat-treated for $1\frac{1}{2}$ hr. at 465° C., followed by cold water quenching. This treatment produced a grain-size of about 0·2 mm. dia.

(ii) *Commercial copper:* This was cold rolled to 2 mm. from $\frac{1}{2}$-in.-dia. rolled bar, machined, and then heat-treated for 2 hr. at 800° C. in a nitrogen atmosphere and cooled in the furnace. This treatment produced a grain-size comparable with that of the aluminium–silver alloy specimens.

(iii) *Armco Iron:* This was cold rolled to 2 mm. thickness from $\frac{1}{4}$-in. plate, machined, and then annealed for 2 hr. at 900° C. in a nitrogen atmosphere and cooled in the furnace.

After heat-treatment the specimens were electro-polished without subsequent grinding, so as to ensure that no work-hardened surface layer existed.

The conditions of electrolytic polishing were as follows : Electrolyte, 50% orthophosphoric acid in water (specific gravity of the solution 1·35) ; temperature, 30°–40° C. ; current density, 15 amp./dm.² The specimen was rotated within a cylindrical aluminium cathode.

The etching reagents found most suitable were :

(a) Aluminium–$\frac{1}{2}$% silver alloy : 50% HF in water, followed by a dip in conc. HNO_3.
(b) Commercial copper : $FeCl_3$ + HCl in alcohol.
(c) Armco iron : 2% nital.

III.—EFFECT OF STRESS ON THE STRUCTURE

The stresses imposed on the specimen were not measured accurately, owing to the complexity of
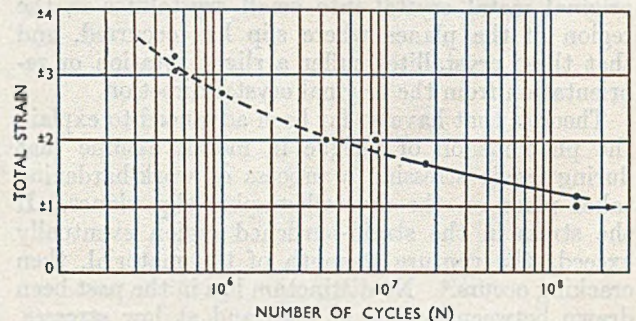
Fig. 2.—Strain/Log N Curve for Aluminium–$\frac{1}{2}$% Silver Alloy. The strain is measured in arbitrary units.

– – – – Widespread deformation with marked crystallite formation.
——— Deformation bands with faintly defined crystallites.

stress distribution introduced by the notch that was provided to localize failure to the field of view, and because the materials were being stressed above their elastic limit. The form of curve obtained on the aluminium–silver alloy by operating under conditions
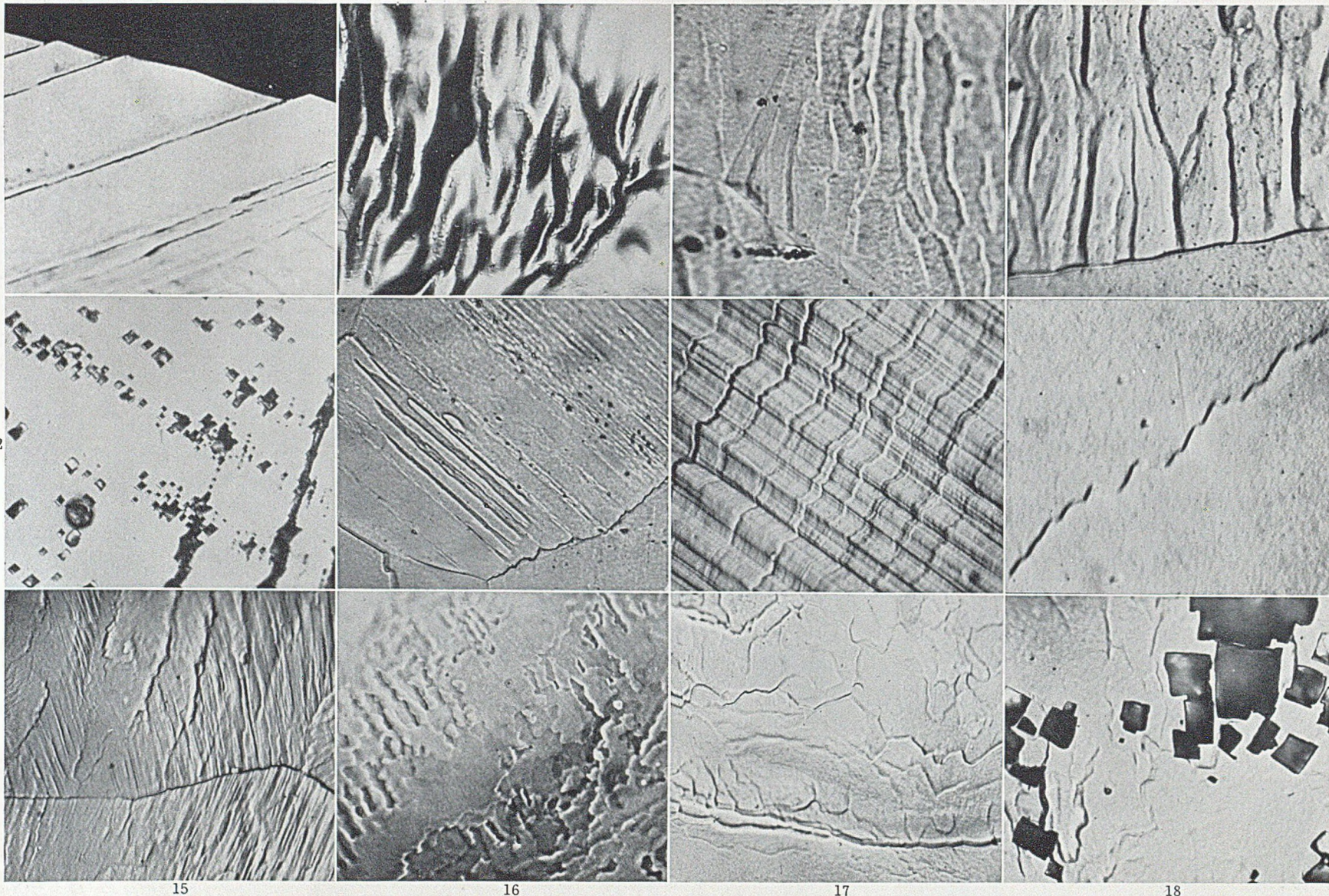
FIG. 7.—Slip Displacement Produced by Static Stress. × 1500.
FIG. 8.—Surface Corrugations Produced by High Fatigue Stress. × 200.
FIG. 9.—Deformation Bands Produced by Cyclic Stresses. × 1500.
FIG. 10.—Deformation Bands Produced by Static Stresses. × 1500.
FIG. 11.—Deformation Bands Revealed by Etch-Pits. × 100.
FIG. 12.—Deformation Bands Parallel to Slip Planes in Heavy Slip Regions. × 500.
FIG. 13.—Relatively Large Displacements Produced by Slip Indicated by the Change of the Operative Slip Planes During Test. × 500.

FIG. 14.—Grain-Boundary Irregularities Produced by Deformation. × 1500.
FIG. 15.—Polygonized Kink Bands with Sharply Defined Boundaries. × 1000.
FIG. 16.—Crystallites Produced by Interaction of Slip and " Kink "-Type Deformation Bands (Phase-contrast illumination). × 700.
FIG. 17.—Polygonization Near an Original Grain-Boundary. × 500.
FIG. 18.—Relatively Large Differences in Orientation Between Neighbouring Crystallites Indicated by Etch-Pits. × 500.

PLATE XXXII.
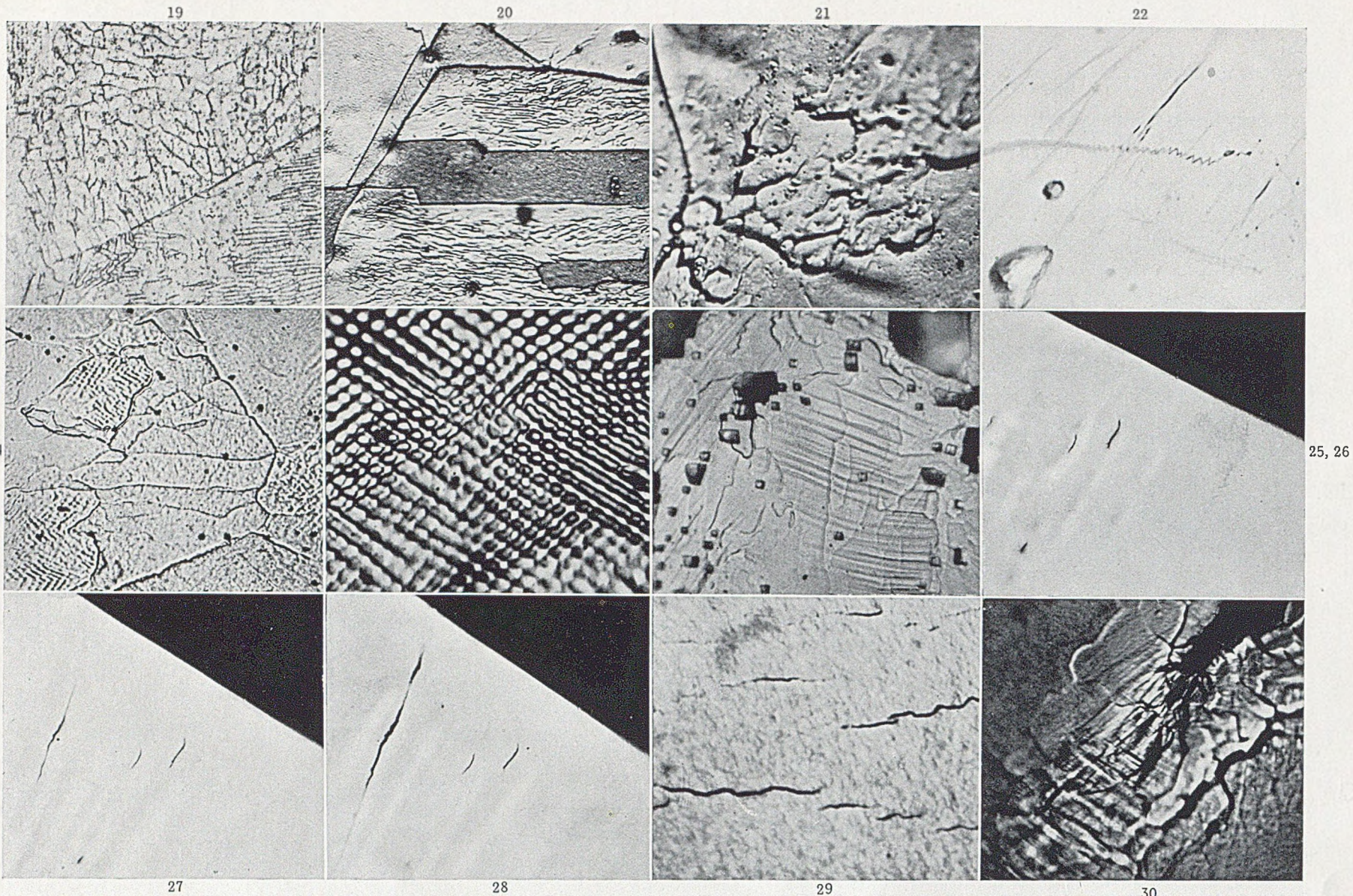
FIG. 19.—Precipitation at Crystallite Boundaries After Fatiguing and Ageing an Aluminium–4% Silver Alloy. × 500.

FIG. 20.—Crystallites in a Fatigued Copper. × 1000.

FIG. 21.—Crystallites in a Fatigued Armco Iron. × 1500.

FIG. 22.—Boundary Movement in Armco Iron. × 1000.

ALUMINIUM–½% SILVER ALLOY.

FIG. 23.—Cellular Network Regions Produced by Fatigue Stresses (Etched). × 500.

FIG. 24.—Cellular Network Pattern Produced by Fatigue Stresses (Unetched). × 1500.

FIG. 25.—Changes in Direction of Slip Bands in an Aggregate of Crystallites. × 500.

FIG. 26.—Appearance of Specimen Surface after 10⁷ Cycles. × 500.

FIG. 27.—As Fig. 26, after a Further 60,000 Cycles. × 500.

FIG. 28.—As Fig. 27, after a Further 60,000 Cycles. × 500.

FIG. 29.—Initial Fatigue Cracks. × 2000.

FIG. 30.—Region of Crystallites at Root of Fatigue Crack. × 1500.

PLATE XXXIII.

of constant strain for each specimen tested is shown in Fig. 2. The strain is plotted in arbitrary units derived from the deflection of the specimen.

During the course of the work it was found that the surface film formed on the aluminium–silver alloy during electrolytic polishing impeded the appearance of slip bands during deformation, although the formation of fine slip bands could readily be observed if the surface was etched to remove the film before stressing. This feature has been described in a recent paper by Wilms.[9] The practice was therefore adopted of etching the specimens before testing them. Fig. 7 (Plate XXXII) shows the displacement produced on slip bands by a static stress.

1. Observations at High Cyclic Stress

The high degree of deformation which may occur under high fatigue stresses can be seen from the rippled surface in Fig. 8 (Plate XXXII). Removal of the surface by electrolytic polishing of a specimen, which had been tested at a stress level above the knee of the fatigue curve but not at sufficiently high a level to produce the marked ripples to be seen in Fig. 8, followed by etching, revealed deformation bands of the type illustrated by Fig. 9 (Plate XXXII). It will be observed that these bands are not associated with any particular set of planes in the lattice, and are similar to kink bands which have been produced by static stress, as shown in Fig. 10 (Plate XXXII). In Fig. 11 (Plate XXXII), the deformation bands have been revealed by etch pits and by the slip bands.

Another type of band was formed parallel to the (111) slip planes in regions of heavy slip, although they were not usually as complete as those illustrated in Fig. 12 (Plate XXXII). The less-complete bands appeared to be elongated regions of high local curvature which were not removed by subsequent repolishing. The clarity with which the bands were defined depended on whether or not the regions of high local curvature had polygonized. The process of polygonization which occurs at elevated temperatures has been described by Cahn.[10] In the present tests, polygonization has taken place at room temperature, presumably because of the more severe lattice curvature.

The way in which these bands are formed by lattice curvature is depicted diagrammatically in Fig. 3, from which it will be seen that if the two halves of the stress cycle slip in opposite directions on groups of parallel planes (a) and (b), the general shape of the grain will remain unaltered. This is illustrated in Fig. 13 (Plate XXXII), where the extent of deformation is rendered clearly visible by slip on a second set of planes. The photographed area may not quite have reached the stage of polygonizing, but it is evident that when polygonization does occur in the regions of lattice curvature produced by the slip a new orientation will be derived, and it is suggested that this is the mechanism by which bands are formed. Another consequence of this form of slip is the movement that

occurs at the grain boundaries (see Fig. 14, Plate XXXII). This movement will be favoured by a condition in which there is little difference in orientation between two adjacent grains, i.e. where the restraint due to the neighbouring grains is small.
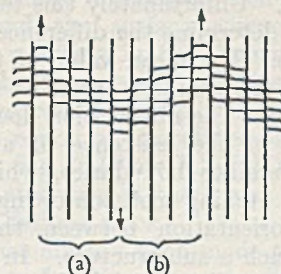
Fig. 3.—A Schematic Diagram Showing the Displacement Produced in a Crystal by Cyclic Stress. The bands marked (a) and (b) are deformed alternately as the stress direction changes.

It has been observed that the presence of deformation bands, produced either by kinking or polygonization of the curved lattice in regions of heavy slip, impedes the slip process. Since the appearance of such bands becomes more marked with increasing stress, it may well be that slip is not the predominant process of deformation at high stresses. This is illustrated by Fig. 15 (Plate XXXII), which shows a specimen that had been tested and then repolished and etched to reveal the deformation bands. These had sharply defined boundaries, indicating that polygonization had occurred. The specimen was then tested further to produce slip bands on the surface, and it will be seen from the photomicrograph that these bands are very incomplete and often confined to particular deformation bands.

As the result of further slip within the crystal, deformation bands formed in the initial stages of cyclic stressing are frequently broken down into crystallites. A photomicrograph taken with phase-contrast illumination reveals this effect in Fig. 16 (Plate XXXII). The formation of crystallites near grain boundaries, where the interaction of slip in neighbouring crystals may cause lattice bending with subsequent polygonization, is shown in Fig. 17 (Plate XXXII). In this particular photograph it will be seen that the boundaries of polygonized crystallites form contours. The etching of a specimen which contains these crystallites reveals that the boundary steps following the curvature are of the same sign; this is shown diagrammatically in Figs. 4 and 5.

Fig. 4 shows the original surface after fatigue (a), shows how repolishing tends to flatten the surface (b), and indicates the " stepped " surface (c), which will be observed after etching with a reagent that attacks the grains preferentially. The variation in orientation between adjacent crystallites may be appreciable, as is apparent from the etch-pits shown in Fig. 18 (Plate XXXII) where the crystallite boundaries are

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clearly visible, the whole field being within one original grain. Etch figures have indicated that this difference may be as great as 30°, although the more usual variation is of the order of 10°. The use of X-ray technique has confirmed that a mean variation of 10° existed in the orientation of crystallites within one original crystal. Unfortunately this technique could not be used to determine the difference between two adjacent grains. The mean value of the crystallite size was 10 μ, the smallest crystallites encountered being about 1 μ. The observation does not exclude the possibility of the existence of a substructure within the crystallite boundaries, which could not be revealed by etching methods owing to the small difference in orientation between the component "blocks" of such a substructure. In fact, an aluminium–4% silver alloy which had been tested in the solution-heat-treated condition (i.e. as a single-phase
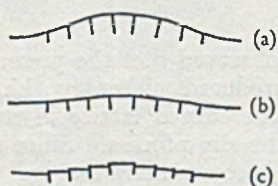
Fig. 4.—A Schematic Representation of Polygonization in a Region of Curvature (*a*) After Deformation, (*b*) After Repolishing, and (*c*) After Repolishing and Etching.

Fig. 5.—A Three-Dimensional Representation of the Boundary Contours Produced by Etching a Curved Region as Shown in Fig. 17 (Plate XXXII).

alloy) to form crystallites, and then aged at 300° C. to effect precipitation of the γ phase on the crystallite boundaries, suggested evidence of a substructure within the crystallites, as will be seen in Fig. 19 (Plate XXXIII).

It is of interest to know whether crystallites are formed at room temperature during the fatigue testing of metals which have higher recrystallization temperatures than the aluminium–silver alloy has. For this purpose specimens in commercially pure copper and Armco iron were subjected to cyclic stress in a similar manner to the aluminium–silver alloy. The results are shown in Figs. 20 and 21 (Plate XXXIII), from which it is evident that the formation of crystallites does occur in copper and iron. It was also observed that, as in the aluminium–silver alloy, local boundary movement took place during the testing of these metals (see Fig. 22, Plate XXXIII).

A cellular network pattern was often observed on the polished surface of the aluminium–silver alloy specimens after fatiguing and is illustrated by Figs.

23 and 24 (Plate XXXIII). The specimen shown in Fig. 23 had been fatigued, re-polished, and etched to reveal crystallites, and then further tested to produce the cellular structure. It will be seen that the cellular structure appears in areas where crystallites have been formed, and that some of the crystallites, presumably with a favourable orientation for slip, show this structure very clearly. The repolishing and the etching treatment to which this specimen was subjected removed all evidence of the cellular pattern, but after further testing it re-appeared in certain crystallites, indicating that more deformation had taken place. Fig. 24 shows that these differences in orientation are observable in the unetched condition.

2. Observations at Low Cyclic Stresses

The essential difference observed to occur in specimens subjected to high and to low cyclic stresses was the absence of the kink-type deformation band in the latter. Slip was initially the predominant process at low stresses, but this eventually led to the formation of the type of deformation band associated with slip bands as the result of polygonization. Such an example is shown in Fig. 12 (Plate XXXII), in which the specimen had been tested for 150×10^6 cycles when the photomicrograph was taken.

3. Effects of Crystallites on Deformation

An important effect of the presence of crystallites within a crystal would seem to be the increased resistance to slip offered by such an aggregate, as compared with the original crystal. Thus the slip bands in Fig. 25 (Plate XXXIII) change direction at the boundaries of the crystallites with a consequent decrease in the mean free path of slip. It will be observed that the slip bands are straight within any particular crystallite. Bragg [11] has advanced the theory that the specific shear energy required to initiate slip in a single crystal is an inverse function of the square of the crystal size, and Wood and Rachinger [12] have developed a theory of the strength of metals which depends on a limiting crystallite size for a particular metal. It is clearly important that the properties of such an aggregate be investigated.

4. The Formation and Propagation of Cracks

The experiments described above have been concerned with a study of the structural changes which take place in a metal subjected to cyclic stressing before rupture of the crystal lattice. In this stage of the fatigue test it has been shown that the deformation which occurs is the result of a combination of crystallographic slip and the formation of deformation bands and crystallites. It might be expected that ultimate rupture of the lattice would be associated in some way with the structures so formed, and an attempt has been made to obtain evidence of this. However, the development of a

suitable technique to reveal the nature of a crack when it first appears has presented a difficulty. Thus, in attempting to determine whether or not the crack forms at the boundaries of the crystallites or deformation bands, an etched specimen in which such a structure had previously been developed by cyclic stressing was kept under observation during a further period of fatigue until a crack was formed. However, slip within the crystallites and the formation of the cellular structure prevented an accurate observation of the way in which a crack formed in the structure. If, after the formation of a crack, the specimen was electrolytically polished and etched to remove evidence of slip and of the cellular structure for better observation of the crystallite boundaries, detailed observation of the original form of the crack was prevented by rounding of the edges of the crack during the polishing.

Figs. 26–28 (Plate XXXIII) show the formation and growth of a crack as observed in a polished specimen. It will be noted that the crack first forms below that edge of the specimen which lies in the plane of maximum stress. Another feature of interest which has
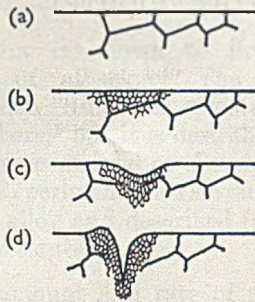
Fig. 6.—A Schematic Representation of the Formation of a Fatigue Crack in Aluminium–$\frac{1}{2}$% Silver Alloy at High Stress.

been observed is that it is not necessarily the first crack formed which propagates through the specimen to cause ultimate failure; it should be appreciated that observations have been limited to the surfaces of the specimen and that no attempt has yet been made to explore the path of the crack into the interior of the specimen.

Fig. 29 (Plate XXXIII) reveals, at a higher magnification, the corrugated nature of the cracks. It was found that the corrugations were of the same order of magnitude as the crystallites, which suggests a possible correlation between the two.

As the result of the examination of cracks in a large number of specimens, the impression has been formed that the cracks sometimes follow the crystallite boundaries and at other times are transcrystallite in character. However, it has not been possible to produce clear evidence of this in any of the individual photomicrographs that have been taken at room temperature, although the cracks which formed in a specimen fatigued at 150° C. definitely followed the crystallite boundaries.

A schematic description of the way in which a

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crack is believed to form in an aluminium–$\frac{1}{2}$% silver alloy at high fatigue stresses is given by Fig. 6. Thus, the original grains of the annealed structure (*a*) are broken locally into small, faintly defined crystallites (*b*), and as fatiguing continues the crystallites become more sharply defined with evidence of surface flow (*c*); the crystallite formation then becomes more widespread and the surface corrugations become more severe, resulting in the formation of a deep groove (*d*), which eventually initiates a crack. The crack then progresses through the specimen, during which stage a region of crystallites precedes the root of the crack. Fig. 30 (Plate XXXIII) shows the condition of the structure in the region of the root of an advancing fatigue crack.

Reference has already been made to the fact that the first crack to form is not necessarily the one which propagates through the metal to produce failure. This would be difficult to appreciate if the region at the root of a fine crack were considered to be one of strain-hardened material subjected to a high stress concentration; however, the formation of crystallites ahead of the advancing crack may offer an explanation. Thus the relaxation which is associated with the formation of crystallites will result in a lower rate of strain-hardening than would occur in the absence of crystallites; in addition, the effectiveness of the crack as a stress-raiser will be reduced by the presence of branch cracks associated with the boundaries or cleavage planes of the crystallites.

IV.—CONCLUSIONS

Evidence obtained from a metallographic study of the structural changes that occur when an aluminium–$\frac{1}{2}$% silver alloy is subjected to cyclic stressing at room temperature has indicated that the process of fatigue is associated with the following phenomena :

(*a*) Slip occurs at all stresses which lie on the endurance curve.

(*b*) At a high stress level, i.e. above the "knee" of the S/N curve, deformation bands of the "kink" type occur in the early stage of fatigue, and deformation bands form along slip planes as the result of polygonization in regions of high lattice curvature.

(*c*) At a low stress level, i.e. below the "knee" of the S/N curve, deformation bands only of the second type described in (*b*) are formed.

(*d*) Crystallites are formed as the result of deformation bands being broken down by subsequent slip, or of the polygonization of regions of high lattice curvature, particularly near grain boundaries.

(*e*) The mean value of the crystallite size is about 10 μ, the smallest value being about 1 μ.

(*f*) The difference in orientation between neighbouring crystallites is up to about 10°.

(*g*) Final rupture of the lattice takes place in a region of crystallites, although it has not been definitely established whether the resulting crack forms at the crystallite boundary or within the crystallite; such

evidence as has been obtained, however, suggests that cracks originate at the crystallite boundaries.

(*h*) A region of crystallites is formed ahead of an advancing crack, which probably reduces the rate of strain-hardening in the lattice, whilst the development of branch cracks along crystallite boundaries reduces the effectiveness of the crack as a stress-raiser.

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A MECHANISM OF STRESS-CORROSION IN ALUMINIUM–MAGNESIUM ALLOYS *

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SYNOPSIS

The attack on aluminium and aluminium-rich alloys, immersed in neutral chloride solutions, tends to become restricted to a limited area owing to the autocatalytic nature of the anodic reaction. On homogeneous alloys, corrosion causes scattered pits which can increase either in depth or in area; but in the case of alloys possessing a path of easy corrosion, the attack takes the form of trenching and is confined to the tips of the cracks so formed. This latter type of corrosion leads to great mechanical weakening, even though the actual rate of attack per unit area may not be very different in the two instances.

The nature of the path of easy corrosion is unknown, but there is evidence that it is sub-microscopic in width. No indication has been found that, in the case of aluminium–magnesium alloys, it consists of the equilibrium second phase, β.

From experiments carried out on aluminium–7% magnesium alloy, it is concluded that stress exerts no influence during the first stages of the process (the greater part of the life), but that corrosion plays a vital role during the final rapid cracking.

I.—INTRODUCTION

STRESS-CORROSION is known to involve a highly localized type of attack.[1-4] The mechanism of localized corrosion has been treated in a paper published elsewhere,[5] but it is desirable to summarize here the conclusions arrived at therein. This is done in Section II. Experiments carried out on aluminium–7% magnesium alloy and described in later Sections, were designed to establish :

(1) The amount and rate of corrosion taking place before failure.

(2) The effect of stress on the corrosion reaction.

(3) The effect of corrosion on the progress of cracks.

(4) The nature of the easily corrodible material found at the grain boundary.

Certain aspects of the mechanism of stress-corrosion are discussed in the light of these experiments and of the mechanism of localized attack outlined below.

II.—LOCALIZED CORROSION OF ALUMINIUM

The slow rate of corrosion of aluminium and its alloys is due to the protective film of oxide which normally covers the metal, so that corrosion can take place only when either metal ions or some anion penetrate the film. (The two processes can be regarded as similar for the present purpose.) Aluminium ions arriving at the outer surface of the film can either leave the metal by becoming solvated or remain on the surface and form a solid product. If all the ions arriving at the outer surface are solvated, the film cannot thicken, and the continuous removal of ions from beneath it will undermine the film. Eventually, therefore, the surface should become practically film-free. If, on the other hand, most of the ions arriving at the surface form, for example, an oxide, the film will thicken and the reaction must diminish rapidly with time.

In the case of a pure aluminium anode immersed in a neutral chloride solution, the proportion of ions that can become solvated is found to increase with time and with rise in current density,[5] and the film therefore tends to become thinnest, or weakest, at the place where the reaction is most rapid. In other words, the anodic reaction on aluminium is auto-catalytic. This explains why corrosion, having started at one point, continues there in preference to any other points.

There is, however, an upper limit beyond which the rate of the anodic reaction cannot readily increase. This is believed to be due to the fact that aluminium anodes generally corrode faster than would be expected from Faraday's law, owing to a local reaction between aluminium and water by which a solid product and hydrogen [5,6] are formed. The anodic reaction is thus autocatalytic only up to a given point, beyond which thickening of the film due to the local reaction balances undermining due to the current flowing to distant cathodes. It is probable, therefore, that current densities lower or higher than that corresponding to minimum filming will be unstable. That is to say, for a given total corrosion rate the active anode area tends to become constant, and the rate of linear penetration down into the metal at the anodic points also tends to be constant.

The anodic behaviour of alloys is more complicated, but, qualitatively, aluminium–magnesium alloys

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should behave in much the same way as pure aluminium. It is likely that, at a given pH, the proportion of soluble corrosion product resulting from attack on an alloy containing magnesium will increase with the magnesium content. Any relatively magnesium–rich paths should therefore corrode preferentially. It is still true, however, that for any alloy composition there is a rate of reaction for which filming is a minimum, and that faster or slower rates of corrosion are unstable. This minimum rate is not likely to alter by a large factor with small changes in composition.

III.—AMOUNT AND RATE OF CORROSION

Unless otherwise stated, the alloy used in the experiments now described had the following composition : magnesium 7·3, copper 0·01, manganese 0·07, iron 0·045, silicon 0·009%, remainder aluminium. It was rolled to 18 S.W.G. sheet and given a solution-treatment before ageing.

Specimens were aged for 2 days at 200° C., and some were then bent beyond the yield point into U

cracking can occur after only a few micro-equivalents of metal have corroded and that there is no obvious difference between stressed and unstressed specimens.

IV.—EFFECT OF STRESS ON CORROSION

The stress-corrosion life of specimens stressed throughout the entire duration of a test was compared with that of specimens which had first undergone a period of corrosion without stress, in order to establish whether the stress affected the corrosion reaction throughout the course of the experiment. Some of the results are given in Table II, from which it can be seen that the total life is not much affected if stress is absent during the greater part of the life of the specimen. It therefore appears probable that during this period corrosion is taking place at numerous grain boundaries and not only at those that will finally cause failure. Only a small proportion of the aluminium found in the solution during the experiments described in Section III therefore comes from the actual crack. The area of the fracture surface (which was intercrystalline) was probably of the

TABLE I.—*Effect of Stress on Corrosion Rate.*

Based on amount of aluminium found in solution after the corrosion of stressed and unstressed specimens.

Specimen No.	Nature of Solution	pH of Solution		Amount of Solution Used, c.c.	Duration of Experiment, min.	Exposed Area, cm.²	Stressed or Unstressed	Aluminium Found in Solution, $\times 10^{-6}$ g.
		Original	Final					
1	} 3% NaCl {	~4	...	25	10	0·5	S	4·8
2		~4	...	25	10	0·5	U	4·8
3	} 3% KCl {	2·45	2·46	35	15	1·0	S	12
4		2·45	2·49	35	15	1·0	U	12
5		2·45	2·49	35	15	1·0	S	26
6		2·45	2·52	35	15	1·0	U	36
7	} 3% KCl {	3·5	...	5	7	0·5	S	4
8		3·5	...	5	7	0·5	U	4

loops. These were forced into rectangular glass vessels, thus subjecting them to an elastic stress arising from the pressure against the walls of the vessel. Slightly acidified chloride solutions were added.

When the stressed specimens failed they were removed, together with the appropriate unstressed controls. The aluminium content of the solution was then estimated using Aluminon.[7] Results are given in Table I, from which it will be seen that specimens can fail when only a few microgrammes of aluminium have corroded and that no significant difference exists between the rate of corrosion of stressed and unstressed specimens.

The corrosion rate was also studied by following the changes in pH occurring in an almost neutral solution. This method has the advantage of being practically independent of the amount of diffusion of corrosion product out of the small cracks and of being highly sensitive. The experiments confirmed that

order of 1 cm.², and from the quantity of aluminium found, the width of the corroded path may be estimated at less than 100 Å.

The effect of stress on corrosion during the last stages of the process was studied with the aid of a microscope fitted with a water-immersion lens. Stress was applied by means of the apparatus shown in Fig. 1 (Plate XXXIV), which is operated by metal bellows inside the cylindrical tube. It was noticed that a few hydrogen bubbles sometimes formed soon after immersion, but that large quantities appeared only a few minutes before failure, emerging from cracks which were already developing. These experiments prove that there is an increase in corrosion rate in the cracks during the final stages, but that the increase is not sufficiently great to be detected by the method described in Section III. It is impossible to tell how much of the increased corrosion is taking place at the tip of the crack and is connected with the cracking, and how much is taking place on the newly formed

Plate XXXIV.
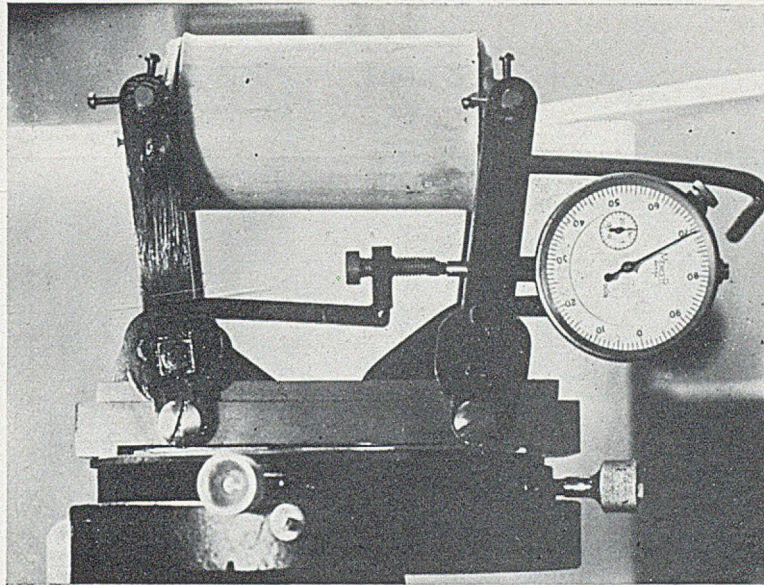
Fig. 1.—Apparatus Used for Stressing Specimens while under Microscopic Examination.

(a)   (b)   (c)
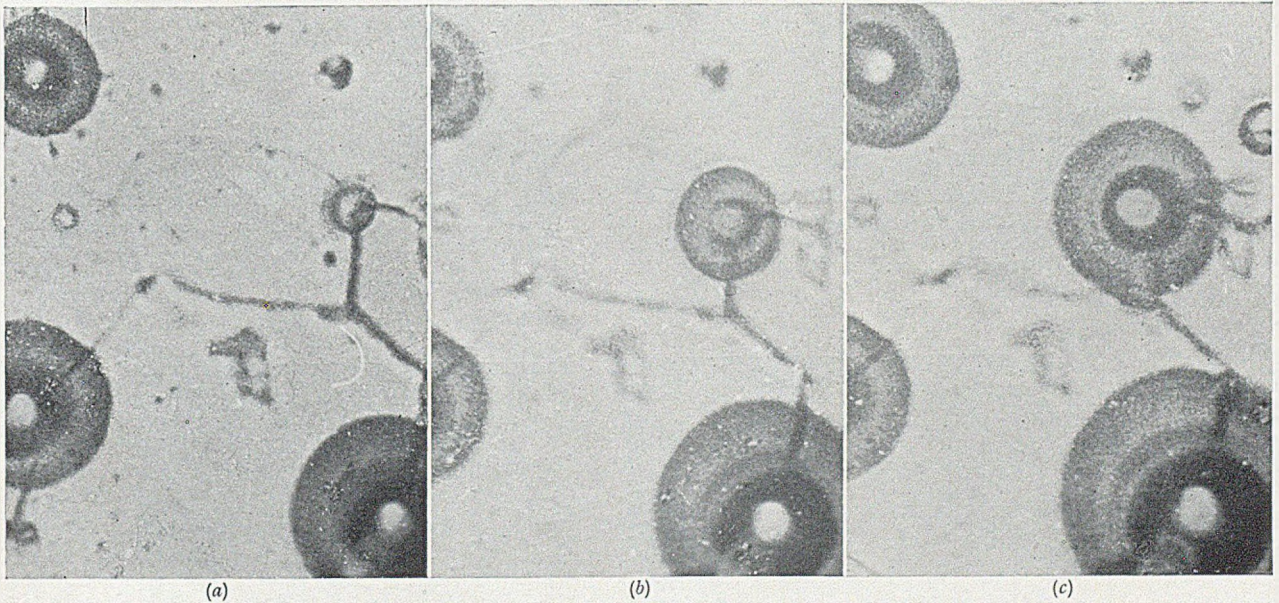
Fig. 2.—Evolution of Hydrogen Bubbles from a Specimen of Aluminium–7% Magnesium Alloy Cracking by Stress-Corrosion. (a), (b), and (c) were taken at intervals of approx. 1 sec. × 1000 approx.

Fig. 6.—Grain-Boundary in Al–7% Mg Alloy Aged for 6 Hr. at 200° C. Etched for 10 sec. in Keller's Reagent. × 1500.

Fig. 7.—Same Field as Fig. 6, after Etching for 2 Min. × 1500.

Fig. 8.—Grain-Boundary in Al–7% Mg Alloy Aged for 2 Hr. at 200° C. Etched in Lacombe's Reagent. × 1500.

PLATE XXXV.

ALUMINIUM–7% MAGNESIUM ALLOY.
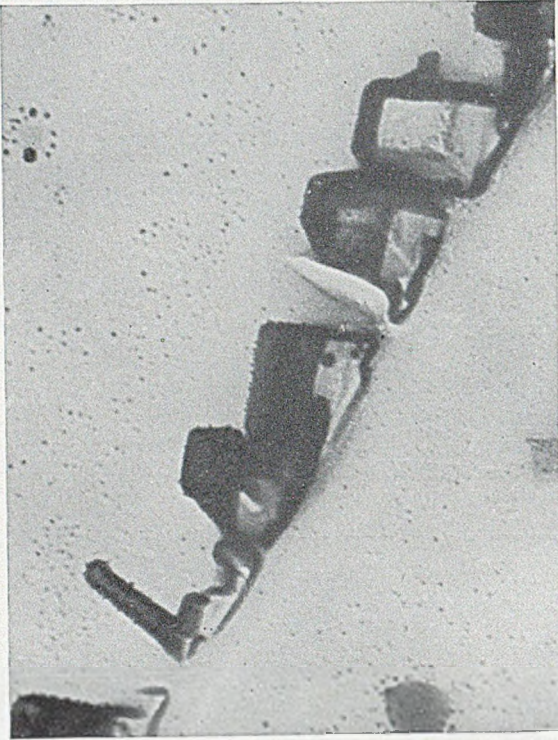
FIG. 9.—Electron Micrograph of Grain-Boundary in Cor-
rosion-Susceptible Alloy. Oxide replica (*Nutting*).
× 15,000 reduced by $\frac{1}{3}$ linear in reproduction.
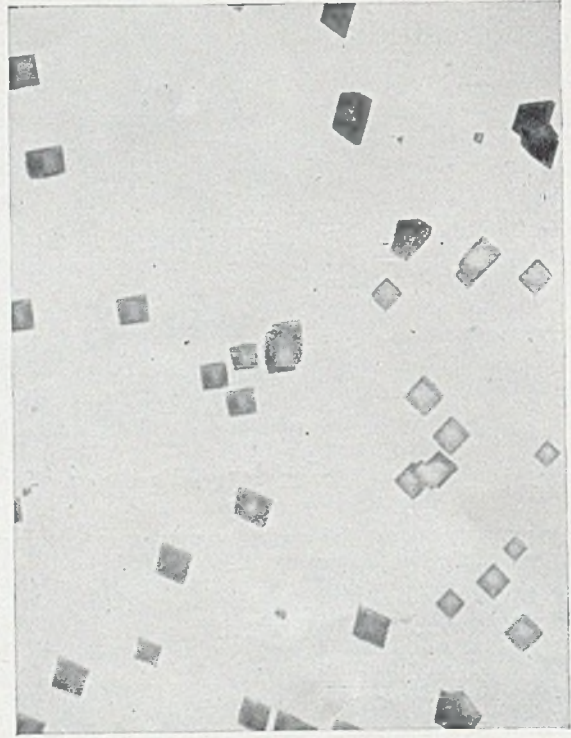
FIG. 10.—Grain-Boundary in Alloy Quenched from 430° C.
Etched in Lacombe's Reagent. × 1000.
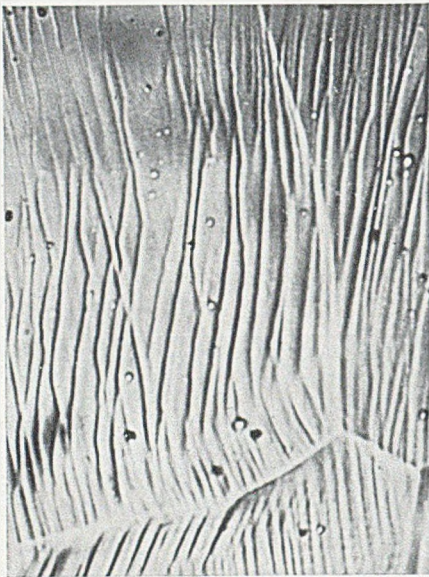
(a)

(b)

FIG. 11.—Illustrating Difference in Appearance of Slip Bands in (a) Quenched and (b) Aged Alloy
(2 Days at 200° C.), After Stressing to 80% of U.T.S. × 1000.

walls (which are presumably becoming filmed over with oxide), and is the result of mechanical failure. If the stress was relaxed during the stage of rapid

TABLE II.—*Effect of Stress during the First Stages of Stress-Corrosion.*

Corrosive solution used : 3% NaCl solution in Cambridge tap-water.

Time of Corrosion without Stress	Time of Corrosion under Stress	Total Time to Failure
Batch 1		
	min.	min.
0	40	40
0	50	50
0	55	55
min.	min.	min.
50	5	55
50	10	60
50	10	60
50	10	60
50	10	60
50	10	60
50	10	60
Batch 2		
	hr.	hr.
0	1–2	<2
0	1–2	<2
0	1–2	<2
hr.	min.	hr.
4	<1	~4
4	<1	~4
4	<1	~4

hydrogen evolution, the evolution continued for an appreciable time, thus showing that at least part of the increase in corrosion rate is not directly due to the stress.

An attempt was made to follow cinematographically the progress of a crack and the evolution of hydrogen from it. Figs. 2 (a)–(c) (Plate XXXIV) show the rate of growth of hydrogen bubbles at a late stage (10–20 sec. before failure) in the life of a specimen. The film was taken at about 18 frames/sec. and there is about 1 sec. difference between each picture.

V.—EFFECT OF CORROSION ON CRACKING

During the microscopic studies just described, with specimens having a life of about 1 hr., cracks became visible only a few minutes before actual failure. Those cracks which started at the edge of the specimen generally required 3–5 min. to advance a distance of one to two grains, but after that they progressed very rapidly and produced failure within a few seconds. It was also found that the presence of a crack could generally be detected by means of an extensometer, such as a Mercer dial instrument. Doubt was felt as to whether corrosion was still playing an essential part during this rapid final stage, and it was decided to investigate whether cracking could be halted by electrochemical means.

Specimens (solution-treated and aged for $2\frac{1}{2}$ days at 175° C.) were loaded through levers to 37% of their ultimate tensile strength. The movement of a second lever, arranged to give a total mechanical advantage of 140, was recorded on a rotating drum. A glass cylinder containing 3% sodium chloride solution in tap-water was held in position round the specimen by means of a rubber cork. A steel spring was used to make electrical contact with one of the levers when the specimen had extended a given amount, thus completing a circuit which made the specimen cathodic with respect to a second electrode placed in the solution. Time/extension curves obtained with and without cathodic protection are shown in Fig. 3. Curve B was obtained when a current of about 0·3 amp. was applied to the specimen after it had extended 2×10^{-3} in., halting the crack despite the fact that it had already penetrated almost half-way into the specimen.

The mechanism of this cathodic protection is interesting, since it is impossible to obtain a potential low enough to produce equilibrium between aluminium and its ions in aqueous solution. The application of the external current presumably causes an increase in the pH of the solution inside the crack, which favours precipitation and filming of the surface. The formation of such films seems to be sufficient to halt the progress of the cracks. As a rule, cracking did not readily start again, even when the protective current was subsequently interrupted, and in one case at
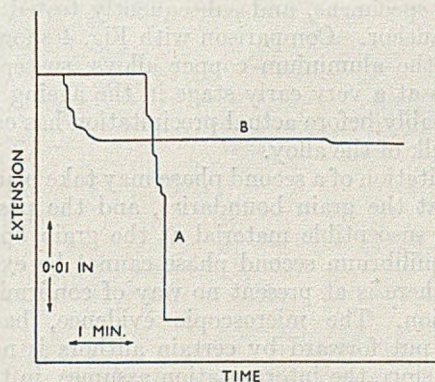
FIG. 3.—Time/Extension Curves for Specimen of Aluminium–7% Magnesium Alloy Failing by Stress-Corrosion.

Curve A: Normal.
Curve B: Specimen protected cathodically after cracking had started.

least an entirely new crack is believed to have developed and caused failure, while the old crack remained inactive.

VI.—EFFECT OF AGEING ON SUSCEPTIBILITY

The susceptibility to stress-corrosion of many aluminium alloys is known to increase to a maximum with time of ageing, and then to decrease. This maximum has been shown to correspond with the

beginning of an increase in hardness, in the case of aluminium–copper alloys, by Robertson,[4] who has also shown that the susceptibility practically vanishes by the time the maximum hardness has been attained.
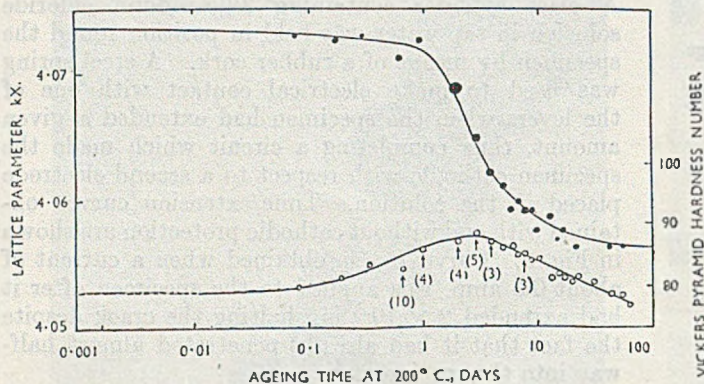
FIG. 4.—Variation of Lattice Parameter (upper curve) and Vickers Diamond Pyramid Hardness Number (lower curve) with Ageing Time at 200° C. for Aluminium–7% Magnesium Alloy. (*Perryman and Hadden.*[8])

The arrows indicate ageing treatments used for stress-corrosion tests at an initial stress of 6·2 tons/in.² and the figures beside the arrows give the corrosion-stress lives in days.

Perryman and Hadden [8] studied the ageing of a commercial aluminium–7% magnesium alloy, and some of their results are reproduced in Fig. 4. Fig. 5 illustrates the stress-corrosion life of specimens cut from the sheet used by Perryman and aged together with his specimens, and subsequently tested by the present author. Comparison with Fig. 4 shows that, as with the aluminium–copper alloys, susceptibility is serious at a very early stage in the ageing process and probably before actual precipitation has occurred in the bulk of the alloy.

Precipitation of a second phase may take place most rapidly at the grain boundaries, and the possibility that the susceptible material at the grain boundary is the equilibrium second phase cannot be excluded, though there is at present no way of confirming this assumption. The microscopic evidence, based on etching, put forward by certain authors is not convincing, since the interpretation assumes, in the first place, a knowledge of the causes of preferential etching.

Microscopic examination of the aged aluminium–magnesium alloy has shown that the size of the so-called " particle " tends to increase with time of etching (see Figs. 6 and 7, Plate XXXIV). An electron-micrograph obtained by Dr. J. Nutting (Fig. 9, Plate XXXV) indicates that the grain-boundary etching marks are sometimes pits and that the shape of these is mainly determined by the orientation of the grains. This can also be demonstrated by means of Lacombe's reagent [9] (HNO₃–HCl–HF), and Fig. 8 (Plate XXXIV) illustrates the appearance of an aged alloy etched with this reagent. It will be seen that the appearance of the boundaries is governed by the relative orientation of the boundary and the grains, and that one boundary which is favourably orientated

has become continuously etched. Under a low magnification, such a specimen gives the impression of having an uninterrupted layer of second phase at one boundary and particles at the remainder. The as-quenched alloy is not preferentially etched at the grain boundary, as shown by Fig. 10 (Plate XXXV), in which the presence of the various grains can only be inferred from the shape of the pits. It is therefore clear that ageing does render the boundary more susceptible to attack, but neither the shape nor the size of the pits or trenches seems to be connected with the original structural defect of chemical or physical origin, which caused the susceptibility. Preferential grain-boundary etching may not, therefore, be due to the presence of a second phase, as claimed by certain workers, but may be attributable for instance, to segregation of one of the components at the boundary. According to Whitwham and Evans,[10] physical strain may be responsible for such an attack, though it seems more likely that in the present case the presence of a higher proportion of relatively soluble magnesium compounds makes the boundary more reactive.

The increase, with ageing, in susceptibility to stress-corrosion (as opposed to intercrystalline attack) may not be entirely due to the increase in chemical reactivity of the boundary material, since the mechanical properties of the alloy also alter with ageing. The changes are not very noticeable when using conventional testing methods (U.T.S., hardness, &c.), but Figs. 11 (*a*) and (*b*) (Plate XXXV) show the appearance of slip bands on a quenched and an aged alloy strained to the same extent. The difference in the
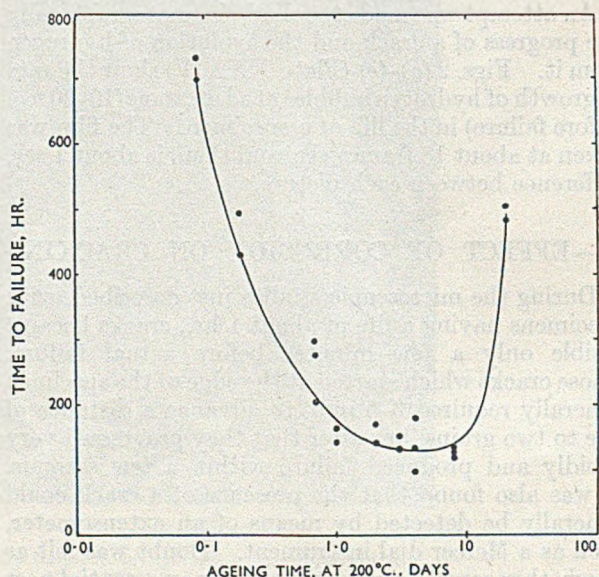
FIG. 5.—Showing Time to Failure of Specimens of Aluminium–7% Magnesium Alloy Aged at 200° C., Made Anodic in 10N-NaCl Solution Under an Applied E.M.F. of 2 V. and Stressed by Levers to 6·2 tons/in.².

slip mechanism is clear. This observation has been confirmed by an electron-microscope study carried out by Dr. A. F. Brown.

VII.—DISCUSSION

On ageing an aluminium–magnesium alloy a layer of easily corrodible material is formed at the grain boundaries. Analytical work tends to indicate that the width of this layer is submicroscopic, and metallurgical evidence shows that susceptibility to stress-corrosion is at a maximum at an early stage in the ageing process. It is therefore possible that the easily corrodible material is not the equilibrium second phase, β, but that the susceptibility is attributable to the segregation (adsorption) of solute atoms at the grain boundary as a first stage in the ageing process. It seems practically inevitable that such segregation must take place before the separation of the second phase, which, in the case of aluminium–magnesium alloys, has a large and complex unit cell. The retention of a layer at a grain boundary when real precipitation starts is unlikely, if the interfacial energy between the new phase and the parent metal is more than half that between two grains. The decline in susceptibility after longer periods of ageing may therefore be due to the separation of a second phase and the destruction of the original segregate grain-boundary layer.

It is believed that preferential attack will take place where the magnesium content is highest, but that the rate of attack per unit area at such points is not likely to be much greater than on pure aluminium. However, the presence of the paths of easy corrosion results in the active anodes remaining at the tips of the trenches, so that the very rapid penetration during intercrystalline corrosion is due more to the continuous attack at the boundary than to an actual increase in rate of attack.

With highly susceptible material, stress does not appear to play any essential role during the greater part of the life of a specimen. When visible cracking starts, there seems to be a slight increase in the rate of attack, although it is not known how much of this extra corrosion takes place at the tip of the crack.

Active anodes carry only very thin oxide films, and it may be that the absence of the normally thicker films facilitates in some manner the propagation of cracks. This view is supported by work on single crystals of zinc and cadmium,[11-13] which has shown that the presence of oxide films can reduce greatly the rate of creep and presumably of other types of deformation. Menter and Hall[13] have suggested that the film hinders the movement of dislocations. It is interesting to note that, in the present case, cathodic treatment, which is believed to cause filming, halts the cracking.

A second way of explaining the rapid advance of cracks, in spite of the relatively slow rate of corrosion, is to assume that the removal of atoms from the lattice provides a continuous source of mechanical weakness.

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METALLURGICAL ABSTRACTS AUTHOR INDEX

SYMBOLS AND ABBREVIATIONS FOR USE IN "METALLURGICAL ABSTRACTS"

Å.	Ångström units $= 1 \times 10^{-8}$ cm.
abs.	absolute
A.C.	alternating current
addn.	addition(s)
amp.	ampere(s)
amp.hr.	ampere-hour(s)
appn.	application(s)
approx.	approximate(ly)
aq.	aqueous
at.-%	atomic per cent.
at. wt.	atomic weight(s)
atm.	atmosphere(s) [pressure]
A.W.G.	American wire-gauge
B. & S.	Brown & Sharpe (gauge)
b.c.c.	body-centred cubic
°Bé.	degree Baumé [scale]
B.H.N.	Brinell hardness number(s)
b.h.p.	brake horse-power
b.p.	boiling point(s)
B.Th.U.	British thermal unit(s)
B.W.G.	Birmingham wire-gauge
°C.	degree Celsius (formerly centigrade) [scale]
C	coulomb(s)
cal.	calorie(s)
c.c.	cubic centimetre (s)
c.d.	current density(ies)
C.G.S.	centimetre-gramme-second (units)
chem.	chemical
cm.	centimetre(s)
coeff.	coefficient(s)
compn.	composition(s)
conc.	concentrated (adj.)
const.	constant (adj. or noun)
contg.	containing
cryst.	crystalline
crystn.	crystallization
c./s.	cycles per second
cwt.	hundredweight(s)
D.C.	direct current
d	density(ies)
detn.	determination(s)
dia.	diameter(s)
dil.	dilute (adj.)
dm.	decimetre(s)
D.P.N.	diamond pyramid (hardness) number(s)
elect.	electric, electrical(ly)
electrochem.	electrochemical(ly)
e.m.f.	electromotive force(s)
estn.	estimation(s)
eV.	electron volt(s)
°F.	degree Fahrenheit [scale]
F	Faraday's constant
f.c.c.	face-centred cubic
f.p.	freezing point(s)
ft.	foot; feet
ft.-lb.	foot-pound(s)
g.	gramme(s)
g.-atom	gramme-atom(s)
g.-mol.	gramme-molecule(s)
gal.	gallon(s)
grav.	gravimetric(ally)
H.F.	high-frequency
h.c.p.	hexagonal close-packed
H-ion	hydrogen-ion
h.p.	horse-power
h.p.-hr.	horse-power-hour(s)
hr.	hour(s)
in.	inch(es)
in.-lb.	inch-pound(s)
indust.	industrial
°K.	degree Kelvin absolute [temperature scale]
kc.	kilocycle(s)
kc./s.	kilocycle(s) per second
kg.	kilogramme(s)
kg.cal.	kilogramme-calorie(s)
kg.m.	kilogramme-metre(s)
km.	kilometre(s)
kV.	kilovolt(s)
kVA.	kilovolt-ampere(s)
kW.	kilowatt(s)
kWh.	kilowatt-hour(s)
kX	Crystal Ångström(s) $= 1000$ Siegbahn X-units
l.	litre(s)
lb.	pound(s)
L.F.	low-frequency
liq.	liquid (adj.)
M	molar [solution]
m.	metre(s)
m.amp.	milliampere(s)
math.	mathematical(ly)
max.	maximum (adj. or noun)
Mc.	megacycle(s)
Mc./s.	megacycle(s) per second
mech.	mechanical
met.	metallurgical(ly)
mg.	milligramme(s)
ml.	millilitre(s)
min.	minute(s); minimum (adj. or noun).
mm.	millimetre(s)
m.m.f.	magnetomotive force(s)
mol.	molecule(s)
m.p.	melting point(s)
mμ	millimicron(s) $= 1 \times 10^{-7}$ cm. $= 10$ Å.
mV.	millivolt(s)
N	normal [solution]
N.P.L.	National Physical Laboratory (Teddington, Middlesex)
No.	number(s)
N.T.P.	normal temperature and pressure
Oe.	oersted(s)
opt.	optical(ly)
oz.	ounce(s)
P.C.E.	pyrometric cone equivalent
p.d.	potential difference [electric]
pH	hydrogen-ion concentration
phys.	physical(ly)
p.p.m.	part(s) per million
prepn.	preparation
prodn.	production
pptn.	precipitation
P.S.	proof stress
qual.	qualitative(ly)
quant.	quantitative(ly)
°R.	degree Réaumur [scale]
recrystn.	recrystallization
ref.	reference(s)
resp.	respective(ly)
r.p.m.	revolution(s) per minute
sci.	scientific
sec.	second(s)
sepn.	separation(s)
soln.	solution(s)
sp.	specific
sp. gr.	specific gravity(ies)
spectrochem.	spectrochemical(ly)
sq.	square
S.W.G.	standard wire-gauge (Imperial)
tech.	technical(ly)
temp.	temperature(s)
t.p.i.	threads per inch
U.T.S.	ultimate tensile stress(es)
V.	volt(s)
VA.	volt-ampere(s)
vol.	volume(s)
W.	watt(s)
Wh.	watt hour(s)
wt.-%	weight per cent.
yd.	yard(s)
Y.P.	yield point(s)
%	per cent.
γ	microgramme $= 1 \times 10^{-6}$ g.
λ	wave-length
μ	micron(s) $= 1 \times 10^{-4}$ cm.
$\mu\mu$	1 millionth micron $= 1 \times 10^{-10}$ cm. $= 0.01$ Å.
Ω	ohm(s)
°	degree(s) arc
'	minute of the arc
"	second of the arc
/	per
<	less than
>	greater than
≮	not less than
≯	not greater than
≦	equal to or less than
≧	equal to or greater than
≠	not equal to
≡	identically equal to
≅	approximately (or essentially) equal to
∼	about
∝	varies as
‖	parallel
⊥	perpendicular

1—PROPERTIES OF METALS

*A Calculation of the Elastic Constants of Aluminium. R. S. Leigh (*Phil. Mag.*, 1951, [vii], 42, (325), 139–155).—The elastic const. of Al in different directions for homogeneous shear without change of vol. are calculated, from consideration of how each of the terms comprising the total energy of the metal is affected. The electrostatic contribution is very anisotropic, but this is balanced by the effect of the Fermi energy, so that Al is elastically almost isotropic. The Fermi energy factor is made up of contributions from the full first zone, and from electrons overlapped into the second zone. To obtain agreement with the experimental values of the elastic const., it is necessary to assume that there are some overlapping electrons across the sq. faces of the Brillouin zone, as well as across the hexagonal faces. The possible presence of un-occupied states in the first zone cannot readily account for the experimental results. The theory is extended to cover variation of the elastic const. with change of electron : atom ratio. On decreasing the electron : atom ratio, the curves show a marked effect of 2·67 electrons/atom when the overlap across the sq. Brillouin zone faces just disappears.
—P. C. L. P.

*Anisotropy of Electrical Resistivity of Cold-Rolled Cubic Metals and Alloys [Al, Cu, Ag, Brass, &c.]. T. Broom (*Phil. Mag.*, 1951 [vii], 42, (324), 56–62).—Annealed strips of the following materials were cold-rolled to 95% reduction : commercially pure Al contg. 0·5% Fe and 0·2% Si ; tough-pitch Cu contg. 0·12% O ; fine Ag ; mild steel contg. 0·07% C and 0·21% Mn ; (α + β) brass contg. 36·5% Zn ; and sterling Ag contg. 7·2% Cu. The elect. resistivity at 1° C. was measured in directions of 0, 45, and 90° to that of rolling. The results for fine Ag are not reported because it is believed to have recrystallized at room temp. The transverse re-sistivity of the steel, brass, and sterling Ag specimens less than that in the rolling direction. The reverse was the case for the Cu. No significant effect was found for Al, where the experimental errors were relatively larger. The mean temp. coeff. for the range −183° to 0° C., was found to be isotropic within ~0·5% in all the specimens. The effects are attributed to oriented dislocations rather than to second phases or general lattice strains.—P. C. L. P.

*Efficiency and Mechanism of Barium Getters at Low Pressures. S. Wagener (*Brit. J. Appl. Physics*, 1951, 2, (5), 132–138).—Quant. experiments are reported with a new type of vacuum gauge incorporating a Ba getter, used at very low pressures. Limiting pressures were found for a number of gases. The results show that atoms or metastable molecules, produced by impact of electrons on normal gas molecules, are absorbed by the getter, but few gas ions are absorbed. The position of the getter with respect to the electron dis-charge has a marked effect on limiting pressure. 14 ref.
—R. W. C.

*On the Electrical and Thermal Conductivity of Bismuth Single Crystals in a Transverse Magnetic Field. E. Grüneisen, K. Rausch, and K. Weiss (*Ann. Physik*, 1950, [vi], 7, (1/2), 1–17).—Two differently orientated single crystals were used for conductivity measurements in variable magnetic fields. The previously reported change of resistance when the magnetic field is reversed (G. and Gielessen, *ibid.*, 1936, [v], 26, 449 ; *Met. Abs.*, 1936, 3, 489) was found to be due to experimental conditions and not a fundamental property of the crystals. Kohler's equation for change of elect. resistance, ρ, in a magnetic field $\Delta\rho/\rho_0 = F(H/\rho)$ was not obeyed (*ibid.*, 1938, [v], 32, 211 ; *Met. Abs.*, 1938, 5, 530). The thermal resistance varied much less than the elect. resistance with change of H. The electronic and lattice contributions to thermal conductivity were separated on the assumption that the latter is independent of H, and for one of the crystals the observed Wiedemann–Franz–Lorentz const. thus derived agreed closely with that given by Sommerfeld's theory.—J. W. C.

*High-Temperature Thermal Analysis Using the Tungsten/Molybdenum Thermocouple [Determination of Freezing Point of Chromium]. (Greenaway, Johnstone, and McQuillan). See col. 317.

*Self-Diffusion in Cobalt. Foster C. Nix and Frank E. Jaumot, Jr. (*Phys. Rev.*, 1951, [ii], 82, (1), 72–74).—Radio-active Co^{60} was evaporated from a W filament on to metallo-graphically polished surfaces of 99·06 and 99·7% pure Co, to give a film <0·05 μ thick. The radioactivity of each specimen was determined before and after heat-treatment for 18 hr. at 1050°, 1150°, and 1250° C. Precautions were taken to reduce errors due to evaporation from the surface, and no difference was found between the two purities of Co. The diffusion coeff. D can be deduced from the measurements if the absorption coeff. μ is known. μ was determined directly by electroplating successive layers of inactive Co and measuring the decrease in radioactivity. The values of D obtained can be represented by the equation $D = 0.37 \exp(-Q/RT)$ cm.²/sec., where $Q = 67,000$ kg.cal./mole. The values of D and Q are consistent with the Langmuir–Dushman equation. Values of Q/T_m and Q/E are calculated, where T_m is the m.p. in °K. and E the binding energy ; these agree with the values obtained for other metals.—P. C. L. P.

*Initial Magnetization and Coercive Force of a Single Crystal of Cobalt Along the Axes of Easy and Difficult Magnetization. Charles Guillaud and Roger Bertrand (*Compt. rend.*, 1948, 227, (1), 47–48).—The results obtained by means of an astatic magnetometer are best represented by Rayleigh's law $J = aH + bH^2$ for both directions of magnetization. Values of a, b, and the coercive force at 20° and −196° C. are given.—N. B. V.

*The Creep and Softening Properties of Copper for Alternator Rotor Windings. N. D. Benson, J. McKeown, and D. N. Mends (*J. Inst. Metals*, 1951–52, 80, (3), 131–142).—The rotor windings of some large turbo-alternators are subject to

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

longitudinal compressive stresses because of the higher coeff. of expansion of the Cu relative to the steel rotor. These stresses may give rise to creep deformation, and the resistance to creep of a number of high-conductivity Cu has been investigated in the temp. range 130°–225° C. In cold-worked Cu resistance to creep may be reduced by softening occurring at the operating temp. In consequence, the softening characteristics of high-conductivity Cu have been determined over the same temp. range. The creep-resistance and the resistance to softening have been shown to be very much greater for Ag-bearing (0·1% Ag) than for Ag-free, tough-pitch Cu. The effect of the Ag addn. is such that the creep-resistance of the Ag-bearing material at 225° C. is equal to that of the Ag-free material at 130° C. The creep-resistance and the resistance to softening of O.F.H.C. Cu are greater than those of tough-pitch Cu when both are Ag-free. Ag-bearing (0·1%) O.F.H.C. and tough-pitch Cu have very similar resistances to creep and softening. In Ag-bearing tough-pitch Cu 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 max. resistance to creep coupled with max. resistance to softening is obtained with cold work of this degree.—AUTHORS.

The Effects of Impurities on the Working of Copper. Otto Nielsen (*Z. Erzberg. u. Metallhüttenwesen*, 1951, **4**, (5), 169–176).—A review of the literature dealing with the effects of small quantities of O, Ag, Sb, S, Se, Te, As, Pb, and Bi—as alloying elements or impurities—on the working properties of Cu. 23 ref.—E. N.

*Progress Report for October, November, and December, 1949 [Diffusion of Silver in Copper]. R. Smoluchowski (*U.S. Atomic Energy Commission Publ.*, 1950, (CAR-6), 4 pp.).—The rate of diffusion of Ag in Cu, using a sandwich of Cu and Cu with 4% Ag was measured by means of Ag pptn. in the grain boundaries on cooling from the annealing temp. A relationship was found between the diffusion rate and the orientation of the grains and grain boundaries. Measurements were also made by contact radiography on a columnar-grained Cu sample electroplated with radioactive Ag and annealed to allow diffusion.—B. W. M.

*Rupture of Oxide Films [on Copper] During Repeated Sliding. A. J. W. Moore and W. J. McG. Tegart (*Australian J. Sci. Research*, 1951, [A], **4**, (2), 181–184).—Particles from the oxide film are seen as non-metallic inclusions in the surface layers after repeated sliding has taken place on un-lubricated Cu. Grain growth is inhibited where they are present; annealing coalesces them. The presence of HCl during sliding reduces the number of particles formed, and lubrication eliminates formation of inclusions entirely. 10 ref.—S. M.

p–n Junction Rectifier and Photo-Cell. [Germanium.] W. J. Pietenpol (*Phys. Rev.*, 1951, [ii], **82**, (1), 120–121).—A letter. Two devices are described which utilize a p–n junction in Ge made by the process outlined by Teal, Sparks, and Buehler (*ibid.*, **81**, 637; *Met. Abs.*, this vol., col. 163). They are: (i) a rectifier capable of working at a voltage >1000 V.; and (ii) a highly sensitive photo-cell which responds to wave-lengths up to 1·9 μ with full quantum efficiency.—P. C. L. P.

Germanium Trigger Photocells. F. A. Stahl (*Elect. Eng.*, 1951, **70**, (6), 518–520).—The relatively large power-handling capacities of Ge trigger photocells, which are basically low-current high-voltage devices actuated by infra-red light, renders the cells particularly adaptable for certain uses. The cell behaviour results from the occurrence of n-type and p-type Ge in the same crystal, arranged so that both are in the contact circuit.—J. R.

*The Electrical Resistance of Gold, Silver, and Copper at Low Temperatures. E. Mendoza and J. G. Thomas (*Phil. Mag.*, 1951, [vii], **42**, (326), 291–303).—A review is given of work on the elect. resistivity of metals at low temp. There is evidence that the resistivities of Au, Ag, Ga, Na, Mg, Al, Mo, Co, and Ce all go through a min. value as the temp. decreases. In some of these metals, the effect has not been

confirmed by other workers. Apparatus is described for measuring the resistivity of wire specimens down to temp. below 1° K. High-purity Au showed a min. above 4° K. and a sharp rise of resistance below 0·25° K. Magneto-resistance experiments suggest that the content of ferro-magnetic impurities is <1 p.p.m. Highly pure Ag also showed a resistance min., though the form of the curves at lower temp. was sensitive to annealing treatments. Experiments on Cu contg. <0·01% impurities gave a curve similar to that for Au, but no Cu specimens were annealed. Some results are also given for Cu contg. 0·94% Sn as principal impurity. 33 ref.—P. C. L. P.

*Heat Conductivity of Steels and a Few Other Metals at Low Temperatures. J. de Nobel (*Physica*, 1951, **17**, (5), 551–561).—[In English]. Thermal conductivities, λ, of Al, Fe, Ni, Monel metal, Duralumin, mild steel, and of Ni-, Mn-, and Cr-steels were measured in 2 ways at various temp. between that of liq. H and that of liq. air. For pure metals, λ at very low temp. is proportional to T (abs. temp.), and a max. lying between 3° and 20° K. in the λ/T graph has been reported in some researches (e.g. Bremmer and de Haas, *ibid.*, 1936, **3**, 692; *Met. Abs.*, 1936, **3**, 455). Such a max. was found only for Al and Fe (purity >99·93%). Monel metal behaved anomalously: its λ at room temp. is lower than that of most of the alloys, but at lower temp. the alloy belongs to the better heat conductors. The steel samples gave values of n, in the Wiedemann–Franz law, ranging from 1·07 to 1·47. No decrease of the value of n accompanied a decrease of temp. Increase of B.H.N. is accompanied by a decrease of thermal conductivity. In the Ni steels, λ attains a min. value at 26% Ni; a few % of Ni cause a large decrease in the λ of Fe.—J. S. G. T.

Correlation Energy and the Heat of Sublimation of Lithium. Conyers Herring (*Phys. Rev.*, 1951, [ii]. **82**, (2), 282–283).—A letter. H. discusses a letter by Silverman and Kohn (*ibid.*, 1950, [ii], **80**, 912; *Met. Abs.*, this vol., col. 101), in which they reported a discrepancy of 4 kg.cal./mole between the theoretical and experimental binding energies. In addn. to the numerical error which S. and K. have corrected (cf. following abstract), H. suggests that the correlation energy was underestimated because the effective mass of the electrons should have been used. Smaller errors in the Coulomb and exchange energies arise from the same cause. The theoretical cohesive energy becomes 36·0 kg.cal./mole, in comparison with the most recent experimental value of 36·5 kg.cal./mole.—P. C. L. P.

Erratum: On the Cohesive Energy of Metallic Lithium. R. A. Silverman and W. Kohn (*Phys. Rev.*, 1951, [ii], **82**, (2), 283).—S. and K. amend Table I of their letter (*ibid.*, 1950, [ii], **80**, 912; *Met. Abs.*, this vol., col. 101) to allow for a numerical error in calculation.—P. C. L. P.

*Experiments on the Influence of [Calcium and] Hydrogen on the Grain-Size of [Chill-Cast] Magnesium and Magnesium Alloys. (Mannchen). See col. 278.

Arc-Cast Molybdenum Probed for High-Temperature Utility. J. L. Ham (*Steel*, 1951, **128**, (3), 106–108).—The prodn. and properties of Mo and its alloys are described.—E. J.

*Determination of Activation Energies for Secondary Creep [of Nickel]. T. H. Hazlett, E. R. Parker, and M. W. Nathans (*U.S. Atomic Energy Commission Rep.*, 1950, (COO-4), 39 pp.).—Tests on Ni show that the creep rates and temp. coeff. for secondary creep under const. stress are essentially independent of the previous thermal and strain history when the structure of the material is stable during the test and the testing temp. range is above a presumed limiting value. A method of accelerating creep testing within the higher-temp. range is suggested. Since the state of met. stable material becomes independent of previous history during secondary creep, the initial part of the test may be carried out at high temp. and correspondingly higher creep rates. Subsequently the temp. may be reduced and the secondary creep rates determined. The temp. coeff., Q, thus obtained permits interpolation of creep rates for intermediate temp., so that one relatively rapid test permits evaluation of creep rates over a temp. range for a given stress level.—B. W. M.

Change of State of Nickel During Secondary Creep at High Temperatures. K. B. Lloyd, T. H. Hazlett, and E. R. Parker (*U.S. Atomic Energy Commission Rep.*, 1950, (COO-5), 12 pp.).—The work showed a continuous change of internal structure in metals during primary creep, whereas in secondary creep the internal state approached a steady condition of grain-size. The room-temp. mech. properties of Ni are relatively unaffected by creep at 700° C., but this may not be true if the creep tests are made at lower temp.—B. W. M.

Improvements and New Results Relative to the Variation of Electrical Resistance of Ferromagnetic Substances [Nickel] with Temperature. Georges Mannevy-Tassy (*Compt. rend.*, 1950, 230, (12), 1150–1152).—New measurements of the elect. resistance of Ni have been made up to 1297° K. In the ferromagnetic state, the resistance can be represented by the equation : $R_T/R_0 = 8.158 \times 10^{-6} T^2 + 1.280 \times 10^{-3} T$, for temp. $T < 475°$ K. Reduced co-ordinates are used to compare the results with those for Fe and MnSb ; very good agreement is obtained with the former, but not with the latter.
—N. B. V.

The Initial Susceptibility of Nickel Under Tension. H. J. Peppiatt and B. N. Brockhouse (*J. Appl. Physics*, 1951, 22, (7), 985–986).—A preliminary letter. The initial susceptibility of a pure Ni wire was measured under various tensions and at various temp. The effect of cold work was also examined. The results are compared with Becker's theoretical treatment, and on the whole there is agreement.—R. W. C.

Ferromagnetic Resonance in Thin Ni Films. J. H. E. Griffiths (*Physica*, 1951, 17, (3/4), 253–258 ; discussion, 258). —[In English]. Ferromagnetic resonance in thin Ni films 0·1–3 μ thick deposited on mica is described and interpreted in accordance with Kittel's formula (*Phys. Rev.*, 1947, [ii], 71, 270 ; 1948, [ii], 73, 155). The experiments show that great care must be taken in ferromagnetic-resonance work to see that the material is completely free from strain ; very thin films adhere to the mica and retain considerable strain ; thicker films (>0·6 μ) have their strain relieved by the bond between the Ni and mica breaking, and give a normal resonance field, or the metal distorts, leaving a different strain on its two sides. The g value for Ni is independent of thickness, having a mean value 2·20 for thin films down to 0·1 μ thick. In discussion, *J. H. van Vleck* stated that it is largely a matter of taste and terminology whether G.'s ferromagnetic-resonance results were interpreted in terms of changes of demagnetization factor or of energy anisotropy.
—J. S. G. T.

On the Hall Effect in Ferromagnetics [Nickel]. N. Rostoker and Emerson M. Pugh (*Phys. Rev.*, 1951, [ii], 82, (1), 125–126).—A letter. The Hall coeff. of Ni deduced from Smith's experiments (*ibid.*, 1910, [i], 30, 1) appears to increase sharply near the Curie temp., if it is assumed that the magnetization remains const. after it has reached the ordinary saturation value. If the change in intrinsic magnetization after saturation is allowed for, this anomalous behaviour of the Hall coeff. disappears. There appears to be a significant paramagnetic contribution to the Hall effect near the Curie temp.—P. C. L. P.

The Adsorption of Hydrogen on Nickel Catalysts. II.—Sorption Isobars from 20° K. to Room Temperature. O. Beeck, J. W. Givens, and A. W. Ritchie (*J. Colloid Sci.*, 1950, 5, 141–147 ; *C. Abs.*, 1950, 44, 6228).—Cf. *ibid.*, 1948, 3, 505 ; *Met. Abs.*, 1949–50, 17, 3. Sorption isobars were determined for H at 0·1 mm. Hg pressure on evaporated Ni films, over the temp. range − 253° to 25° C. Slow activated sorption, interpreted as absorption of H into the interior of the metal structure, was observed, in agreement with previous results. The process is exothermic and, after initiation by raising the temp. to approx. −150° C., will continue at much lower temp.

On the Variation of the Moment and of the Curie Point of Hydrogenated Palladium. Jules Wucher (*Compt. rend.*, 1949, 229, (3), 175–177).—Measurements on thin sheets and wires of pure Pd contg. various amounts of H revealed that there is a straight-line relationship between the Curie const. and the number of H atoms/Pd atom, as there is also for

the susceptibility (Svensson, *Ann. Physik*, 1933, [v], 18, 299). Both the Curie const. and the susceptibility are zero for H/Pd = 0·64, suggesting that the 4d shell of Pd has 0·64 of a hole.—N. B. V.

Heat Capacity of Potassium and Three Potassium–Sodium Alloys Between 0° and 800° C. T. B. Douglas, A. F. Ball, and D. C. Ginnings (*U.S. Atomic Energy Commission Publ.*, 1951, (AECU-1017), 19 pp.).—Enthalpy changes were measured for K and 3 K–Na alloys with 45, 54, and 78 wt.-% K, resp., from which the heat capacities of solid and liq. K, the heat of fusion of K, and the heat capacities of the 3 liq. alloys were derived. The heat capacities of the alloys were found to correspond closely at each temp. in the range investigated with those calculated additively from the values found for pure liq. K and Na. As reported previously for liq. Na (*ibid.*, 1949, (AECD-2639) ; *Met. Abs.*, 1950–51, 18, 692) all the liquids show a min. in heat capacity at 500°–600° C.—B. W. M.

The Heat of Reaction of Graphite and Potassium. L. Quarterman and W. Primak (*U.S. Atomic Energy Commission Publ.*, 1950, (AECU-1018), 15 pp.).—An apparatus is described for measuring the heat of reaction of graphite with an excess of K over the range 66°–95° C. For high-d Acheson graphite, the heat of reaction was 81 ± 2 cal./g., compared with a value of 125 ± 5 cal./g. given by Fredenhagen and Cadenbach (*Z. anorg. Chem.*, 1926, 158, 24 ; *J. Inst. Metals* (Abstracts), 1927, 37, 404). A variation of ~5% was obtained for various samples from the same bar, and other forms of graphite gave results up to 15% lower than that for Acheson material.
—B. W. M.

Electrical Properties of Selenium. I.—Single Crystals. H. W. Henkels (*J. Appl. Physics*, 1951, 22, (7), 916–925).—Small Se crystals were made from the melt or from the vapour. Only the latter were twinned, and all had a good prismatic cleavage and were very soft. The " dark " resistivity was measured on a number of crystals of different purity, ∥ and ⊥ to the c axis, at different temp. and fields. Transient changes of field were also studied. The steady dark resistivity (established after thermal cycling) was ~1–2 × 10⁵ Ω cm. ∥ c axis and ~5 × 10⁵ Ω cm. ⊥ c axis. The uniform values obtained with different samples indicate that Se is an intrinsic semi-conductor. Other data (including thermo-elect. and Hall-effect measurements) are discussed in the light of the semi-conductive mechanism.—R. W. C.

Recovery of Selenium Rectifiers After a Voltage Pulse in the Blocking Direction. K. Lehovec (*J. Appl. Physics*, 1951, 22, (7), 934–939).—The capacity of a Se rectifier is reduced by a voltage pulse in the blocking direction. The recovery rate was measured as a function of several variables, and its acceleration by illuminating the Se was studied. The capacity reduction is ascribed to a removal of electrons from elevated states in the electrode/Se boundary. During recovery this is gradually reversed.—R. W. C.

Change of Activation Energy with Impurity Concentration in Semi-Conductors [Silicon]. L. Pincherle (*Proc. Phys. Soc.*, 1951, [A], 64, (7), 663–664).—A letter. A calculation is made, on defined assumptions, of the variation of activation energy with impurity content for Si. The agreement is not good, but the results show that a contributing cause of the effect is the screening of trapping centres by free carriers.
—R. W. C.

An Achromatic Doublet of Silicon and Germanium. R. G. Treuting (*J. Opt. Soc. Amer.*, 1951, 41, (7), 454–456).— By the use of solid crystals of Si and Ge, each of which has a high refractive index and a different dispersive power, it is possible to calculate the configuration of achromatic doublet lenses of high aperture ($f/1$ to $f/2$) in the infra-red beyond 2 μ. The possibility of growing single crystals with sufficient purity and freedom from internal strain is discussed.
—E. van S.

Changes in Very Thin Films of Silver. Marcel Perrot (*Compt. rend.*, 1947, 224, (23), 1629–1631).—Thin films of Ag (0–10·7 mμ thick) were deposited *in vacuo* on wedge-shaped glass plates, and their reflectivity to light of 5461 Å. wave-length measured after various periods, beginning

immediately after deposition. It was found that for a given film thickness the reflection factor varies with time, tending to a limiting value R_t', which is the more rapidly attained the thinner is the film. This change, which is termed "spontaneous ageing", is ascribed to re-arrangements in the cryst. structure of the film. In the presence of air, R_t' increases, but returns to its original value when the vacuum is restored, however long an interval may have elapsed.—N. B. V.

*The Effect of Gas Adsorption on the Changes in Thin [Silver] Films. Marcel Perrot and (Mlle) Suzanne Arcaix (*Compt. rend.*, 1949, 229, (22), 1139–1142).—Cf. P., *ibid.*, 1947, 224, 1629; preceding abstract. Ag films of two different thicknesses were deposited on glass and measurements made of the variation with time of their reflectivity to light of 5461 Å. wave-length and their elect. resistance, while maintained *in vacuo* (10^{-5} mm. Hg). The reflection factor was found to vary little, but the elect. resistance passed through a min. value soon after deposition of the film and then gradually tended to a limiting value; the sharpness of the min. depended on the film thickness; the change in elect. resistance was also measured at various pressures up to atmospheric of: (a) dry air, (b) laboratory air, and (c) air saturated with water vapour. In (a) the resistance changed little with pressure, but in (b) and (c) it increased approx. linearly, the effect being much more marked in (c). The changes are reversible.—N. B. V.

Corrigenda : Self-Diffusion in Tin [Single] Crystals. P. J. Fensham (*Australian J. Sci. Research*, 1951, [A], 4, (2), 229–230).—Cf. *ibid.*, 1950, [A], 3, 91; *Met. Abs.*, 1950–51, 18, 246. Corrigenda are given of some arithmetical errors made in calculating diffusion coeff.—T. G.

*Adiabatic Magnetization of a Supraconductor [Tin]. R. L. Dolecek (*Phys. Rev.*, 1951, [ii], 82, (1), 102).—A letter. No temp. change $> 10^{-4}°$ K. was found to accompany the adiabatic magnetization of a supraconducting polycryst. Sn sphere. Calculations are made of the upper limits to a possible entropy change on magnetization in the magnetic fields used, which were well below the critical fields necessary to destroy supraconductivity.—P. C. L. P.

*Arc Melting of Titanium Metal [and the Mechanical Properties of the Resulting Ingots]. (Radtke, Scriver, and Snyder). See col. 319.

Titanium Rapidly Growing as Useful Engineering Material. W. T. Tiffin and P. C. Hoffman (*Materials and Methods*, 1951, 33, (2), 57–59).—Methods of prodn., mech. properties, corrosion-resistance, and present and future appn. of Ti are briefly reviewed.—J. W. C.

Titanium : How It Fits In Your Tomorrow. A. H. Allen (*Steel*, 1951, 128, (2), 54–58, 79, 80, 82).—The prodn. and properties of Ti metal and alloys are described. The properties are compared with those of other alloys, and details are given of the compn. and heat-treatments of a number of Ti alloys.—E. J.

U.S.A.F. Tests Titanium in Airplane Structurals. —— (*Steel*, 1950, 127, (22), 60–61).—A brief review is given on the properties and prodn. of Ti and its alloys and of their possible uses in aircraft construction.—E. J.

†Tungsten. E. L. Reed (*U.S. Atomic Energy Commission Publ.*, 1947, (AECD-2700), 54 pp.).—A complete summary is given of the known data on W, with 111 ref. The report includes data on the phys. properties, stability in various media, fabrication and casting, methods of prepn. of the solid metal and its powder, chem. analysis, microscopical examination, nuclear properties, behaviour on irradiation, and alloying characteristics.—B. W. M.

Erratum : On the Pressure–Volume and Pressure–Compressibility Relation of Metals. P. Gombás (*Phys. Rev.*, 1951, [ii], 82, (2), 287).—An error in the expression for the compressibility was made in a previous communication (*ibid.*, 1947, [ii], 72, 1123; *Met. Abs.*, 1947–48, 15, 500). The ordinates of Figs. 2 and 3 require adjustment.—P. C. L. P.

Why Machine Parts Fail. VII.—Failures Originating on the Drawing Board. VIII.—When Not to Blame the Designer. Charles Lipson (*Machine Design*, 1950, 22, (11), 158–162;

(12), 151–156).—Cf. *ibid.*, (10), 97; *Met. Abs.*, this vol., col. 169. [VII.—] L. stresses that 30% of all fatigue failures can be traced to the design's being faulty. Stress concentrations can be avoided by correct design of keyways, splines, fillets, oil holes, threads, and gear-teeth. Figures are given for the stress-concentration factors due to various types of fillets, threads, and keyways. Fatigue failures resulting from combined effects and from over-stressing are briefly discussed. [VIII.—] Deals with failures caused by defective metallurgy, improper fabricating methods and assembly techniques, and other factors out of the control of the designer. The failures can be due to surface decarburization, quenching cracks, inclusions, casting defects, hot shortness, the presence of residual stresses, toolmarks, grinding cracks, surface roughness, forging defects, faulty welds, or bad salvaging treatment. Excellent photographs of fractured components illustrate the text.—D. M. L.

*On the Relation Between the Hardness Number and Coefficients of the Tensile Test. A. Krupkowski and W. Truszkowski (*Arch. Mech. stos.*, 1950, 2, (3), 235–259; *Appl. Mechanics Rev.*, 1951, 4, 291).—Values of the exponent n in Meyer's law $L = Ad^n$ were measured for ball indentations in several metals cold worked in tension by various amounts (L = load, d = impression dia.). The law is found to hold right up to a hemispherical impression. In the fully annealed state, $n = 2.65$ for Au, 2.51 for Cu, 2.50 for 70 : 30 brass, 2.46 for Ni, 2.43 for Ag, 2.36 for Al, and 2.31 for Monel metal. After heavy prestraining in tension, n is very nearly 2.0 for all metals; the ratio of load to projected area is then const. during indentation and is found to be approx. 2.9 times the corresponding tensile yield stress. For annealed or moderately worked metals, empirical relations are proposed between certain characteristics of the hardness and tension tests.

Some Problems of the Theory of Creep. Yu. N. Rabotnov (*Vestn. Moskov. Univ.*, 1948, (10), 81–91; *Appl. Mechanics Rev.*, 1951, 4, 288).—[In Russian]. Various one-dimensional engineering formulations of creep theory are enumerated, and it is pointed out that none of these provides a simultaneous description of the intimately related phenomena of creep, relaxation, plastic heredity, work-hardening, and the so-called "reverse" creep. A one-dimensional theory is next formulated, as an extension of Volterra's theory of elastic heredity, which takes all these phenomena into account. An experimental confirmation of the resulting creep equation is presented. The theory is next applied to problems of pure bending in a thin beam and of deformation in a tube subjected to internal hydrostatic pressure. Stress distributions are computed in each case in terms of the const. associated with a creep function. An extension of the theory to three-dimensional stress distributions is made under condition that all stresses, as well as strains, vary proportionally to the same time-dependent parameter.

Physical Problems of Frictional Resistance. II.—Energy Considerations in Frictional Resistance. III.—Wear and Frictional Resistance. Rembert Ramsauer (*Kolloid-Z.*, 1948, 111, (3), 145–155; 1949, 112, (1), 26–35; *C. Abs.*, 1949, 43, 6879; 1950, 44, 6231).—Cf. *ibid.*, (2), 83; *Met. Abs.*, 1949–50, 17, 565. [II.—] The work involved in moving one surface with respect to another is the sum of the work required to deform elastically and plastically the surface irregularities plus a work term which is similar to that involved in the slipping of shear planes in crystals. In the case of metallic faces *in vacuo*, the surface atoms approach within crystal distances, so that for a tangential strain appreciable forces must be overcome. The decrease in friction caused by a lubricant results from the reduction in shear strength of the surfaces. With increasing ordering of the lubricant mol., the shear strength decreases. Ordering is promoted by adsorption on the surfaces. Beyond this adsorbed film and depending on the strength of the inter-mol. forces, other ordered layers are produced having a quasi-cryst. condition with favoured glide planes ∥ the surface. Any tendency for the liq. mol. to associate or to orientate under the influence of an applied tangential force disturbs this ordering action of the surface forces. [III.—] The amount of wear at the

interface of two rubbing surfaces is not simply related to the frictional loss. When the frictional losses are high, the degree of wear may be small and vice versa. A liq. lubricant may actually increase the amount of surface wear—the more so, if surface-active materials are added to the liquid above certain concentrations. This phenomenon is discussed in terms of surface-tension forces.

The Internal Friction of Solids and [Its Relation to the] Viscosity [of Liquids]. Albert Kammerer (*Compt. rend.*, 1950, **231**, (23), 1285–1286).—Math. Since the damping of elastic vibrations in solids increases with rise of temp., it would appear that internal friction cannot be compared with the viscosity of liquids, which diminishes with rise of temp. K. resolves the apparent contradiction by showing that the modulus of elasticity of the solid decreases more rapidly with rise of temp. than does the viscosity of the liquid.—N. B. V.

Internal Friction and Material Losses of Solid Bodies. I.—General Theory. E. Skudrzyk (*Österr. Ing.-Arch.*, 1949, **3**, (4), 356–373; *Appl. Mechanics Rev.*, 1951, **4**, 212).—S. attempts to treat elastic after-effects by the common expedient of superposing ordinary elastic and viscous effects (cf. Weissenberg, *Abhandl. Preuss. Akad. Wiss., Phys.-Math. Klasse*, 1931, (2)). His theory is tacitly limited to infinitesimal displacement gradients, since he does not distinguish between Eulerian and Lagrangian co-ordinates and since he employs local rather than material time derivatives. He proposes the familiar expression $t^i_j = f(\partial/\partial t)\xi^k{}_{,k}\delta^i{}_j + g(\partial/\partial t)(\xi^i{}_{,j} + \delta_j{}^i)$, where t^i_j is the stress tensor, ξ^i is the displacement vector, and f and g are power series. S. notes that (questions of convergence being neglected) in a harmonic oscillation f and g reduce to complex power series in the frequency, so that the response of the material may then be characterized by four real functions of frequency. He discusses the definition of elastic moduli and damping coeff. appropriate to various special cases in terms of these functions.

†**Physical Problems in Electrical Contacts : Critical Résumé.** K. R. Schade (*Rep. Brit. Elect. Research Assoc.*, 1948, (U/T107), 47 pp.; *Elect. Eng. Abs.*, 1950, **53**, 344).—A review of present knowledge on elect. contacts, with particular attention to the material wear. The influence of chem. and met. properties of the contact material in determining the circuit conditions to be observed to reduce arc formation, material migration, and welding of the contacts, is discussed.

Temperature of Two Metals in Contact. W. Karush (*U.S. Atomic Energy Commission Publ.*, 1950, (AECD-2967), 6 pp.).—Schematic models are considered to describe the nature of the contact between two metals, and certain derivations are made.—B. W. M.

Correlation of Activation Energies for Self-Diffusion in Cubic Metals. J. C. Fisher and R. E. Hoffman (*U.S. Atomic Energy Commission Unclassified Rep.*, 1950, (S.O.2001), 6 pp.).—Activation energies for self-diffusion in cubic metals appear to be determined largely by the heat of sublimation, h_s, the atomic vol. V, and the elastic shear const., c_{44}, according to the relationship $Q = 0.325\, h_s + 0.216\, c_{44} V$. This supports the idea that the diffusion process involves the creation of lattice vacancies and the motion of atoms into vacancies.—B. W. M.

Note on Surface Tension and the Theory of Holes. Robert L. Choate and Louis H. Lund (*J. Chem. Physics*, 1951, **19**, (8), 1062).—A letter. Recent observations by Mukherjee attributing vol. increase at fusion to additional hole formation (*ibid.*, (4), 502; *Met. Abs.*, this vol., col. 170) are extended to surface-tension considerations. The equation $\frac{1}{2}kT = 4\pi\sigma^2\beta\gamma$ ($\beta < 1$) is derived, where k is the Boltzmann const., T is the abs. temp., and β is a const. modifying the surface tension γ operative at a macroscopic surface.—J. R.

*The Simultaneous Determination of Surface Tension and Contact Angle. T. P. Hoar and D. V. Atterton (*Research*, 1951, **4**, (1), 42).—A letter. The well-known formula for the pressure p to force a liquid into a cylindrical capillary, $p = -2\sigma \cos\theta/r$, where σ is the surface tension of the liquid, θ the contact angle, and r the capillary radius, does not permit the separate evaluation of σ and θ, but only of $\sigma \cos\theta$. H.

and A. have constructed conical capillaries of apex angle 2α into which liquid is forced by pressure p.

Then $\quad p = -\dfrac{2\sigma\cos(\theta - \alpha)}{r_1}\quad$ or $\quad p = -\dfrac{2\sigma}{r_2}\left(\dfrac{\cos\theta}{\cos\alpha} + \tan\alpha\right)$,

where r_1 is the capillary radius at the level where the liquid leaves contact with the capillary, and r_2 the radius at the level of the liquid tip, the liquid surface being taken as spherical. Measurement of p, α, and r_1 or r_2 for two conical capillaries of different α gives simultaneous equations which permit separate evaluation of σ and θ. H. and A. have so far applied the method to molten Sn and its contact angle with silica.—F. A. F.

*The Shape of Reflected Interference Fringes from Interferometers Coated with Thin Metal Films. J. Holden (*J. Opt. Soc. Amer.*, 1951, **41**, (8), 504–512).—The shape or intensity distribution of reflected interference fringes from an interferometer which involves the thin film of metal as its front surface undergoes a cyclic change as the film thickens; these changes are related to the structure of the film as well as to its thickness. Consequently they can be used as an aid to recognizing the structure of deposited metal films <500 Å. thick. A knowledge of the changes is also useful in selecting the most suitable metal, and the most suitable thickness, in the use of interferometry to examine surface structures. The metals Ag, Au, Al, Cu, and Cr are considered.—E. van S.

Heat Conduction in Simple Metals. M. L. Storm (*J. Appl. Physics*, 1951, **22**, (7), 940–951).—Math. The normal basic differential equation of heat conduction, $\nabla \cdot (K\nabla T) = S(\partial T/\partial t)$, is non-linear because K and S are functions of the temp. T. ($K =$ thermal conductivity, $S =$ product of density and sp. heat). It is shown that if $[1/(KS)^{1/2}](d/dT)\log(S/K)^{1/2}$ is const., it is possible to transform the above equation into a linear form, which is given. A treatment based on the modern theory of metals shows that KS should be approx. const. with respect to T, and that $\log(S/K)$ should be a linear function of T; the above function should therefore be const., as required. Examination of published data on Al, Ag, Cu, Fe, 0.8% C steel, Pb, Na, and Cd shows that KS is very nearly const. in all cases, and that $\log(S/K) \propto T$ for all the metals except Na and Cd, which have anomalies. The results of applying the transformed equation to an actual heat-conduction problem is compared with that of applying the basic non-linear equation.—R. W. C.

*Thermal Conductivity of Metals in a Strong Magnetic Field. Max Kohler (*Ann. Physik*, 1949, [vi], **5**, (3/5), 181–189).—The thermal conductivity of a metal in a strong magnetic field behaves very similarly to the elect. conductivity. The Grüneisen–de Haas linear relation between these two quantities is shown to be valid for divalent metals, or for any metal with overlapping energy bands such that the number of electrons in one band = the number of holes in the other. The theory of a new method for determining the lattice contribution to thermal conductivity from measurements of thermal conductivity only in different magnetic fields (as distinct from the Grüneisen–de Haas method, which requires elect.-conductivity measurements in addn.) is given, and results calculated from experimental measurements for Be crystals are in agreement with those obtained by the Grüneisen–de Haas method. In a strong magnetic field there is a simple relation between the Hall coeff. A_H and the Righi–Leduc coeff. A_{RL}: $\dfrac{A_H}{T \cdot A_{RL}} = \dfrac{\pi^2}{3}\left(\dfrac{k}{e}\right)^2$; where $T =$ temp. in °K., $k =$ Boltzmann const., and $e =$ elementary charge. The results are obtained without detailed assumptions about electron–lattice interaction or electron-energy distribution.—J. W. C.

*Magneto-Resistance Change and Conductivity Types.— I.–II. Max Kohler (*Ann. Physik*, 1949, [vi], **5**, (1/2), 89–98; 99–107).—[I.—] Theoretical. The behaviour of the bivalent metals in a strong transverse magnetic field is shown to be homogeneous. In the limiting case of a very strong field, the elect. resistance increases as the sq. of the magnetic field, so that these metals become insulators on extrapolation to $H = \infty$. This result is obtained without special assumptions

about the electron–lattice interaction, the electronic-energy distribution, or the crystal structure. It is assumed that: (1) the number of holes in the first energy band is the same as the number of electrons in the second band; (2) the overlapping of the two bands is not too great; and (3) the electron gas is completely degenerated. The anomalous behaviour of these metals in a strong magnetic field arises from the opposing contribution of the two energy bands to the Hall field strength which becomes small (of the order of $1/H$ in relation to the ohmic field strength), so that the increase in conductivity is much smaller than the increase in magneto-resistance. [II.—] The elect. behaviour of metals can be divided into two types, of which typical examples are the univalent and bivalent metals, resp. The elect. resistance and Hall coeff. of univalent metals of cubic crystal symmetry in a very strong magnetic field are investigated without detailed assumptions on electron–lattice interaction or electron-energy distribution, and both are found to show saturation. The Hall coeff. is generally isotropic. In the case when the surface of occupied states does not touch the surfaces of the Brillouin zone, the transverse resistance change is isotropic and the longitudinal resistance change is ⩽ the transverse change. The Hall coeff. = that for free electrons. The resistance change in most univalent metals (K is an exception) is found experimentally to be anisotropic, from which it is concluded that the surface of occupied states at least approaches the Brillouin-zone surface. Values of the Hall coeff. calculated on the assumption that the energy distribution corresponds to free electrons are compared with experimental values for Au, Ag, Cu, Na, and K. Recent values for the Hall coeff. of the alkali metals indicate a stronger electronic bonding than was previously suspected. The question of an adequate definition of the term "strong magnetic field" is considered.—J. W. C.

Theory of the Magneto-Resistance Effects in Metals. Max Kohler (*Ann. Physik*, 1949, [vi], **6**, 18–38).—The increases in elect. and thermal resistance in a magnetic field are considered, and the fundamental statistical equation is solved generally for a simple two-band model. Approx. expressions are derived for the resistance changes in a transverse magnetic field and these agree with the interpolation formula of Sondheimer and Wilson (*Proc. Roy. Soc.*, 1947, [A], **190**, 435; *Met. Abs.*, 1947–48, **15**, 146). The validity of the expressions is verified by comparison with experimental results for pure W and Be. The Grüneisen method for determining lattice conductivity in strong transverse fields is established theoretically for metals of conductivity "type" I. (cf. preceding abstracts). The derived small abs. values for metals at low temp. are in general agreement with the theory, which requires that the lattice conductivity of metals compared with that of insulators should be depressed as a result of the strong interaction between lattice waves and electrons. It is concluded that the crystal structure as well as the valence-electron concentration influences the conductivity "type" of a metal. The theory of magneto-resistance changes can produce few general predictions, because of the involved mathematics, but the results are in good agreement with experiment.—J. W. C.

*Magneto-Resistance of Ferromagnetic Metals and Alloys at Low Temperatures. J. Smit (*Physica*, 1951, **17**, (6), 612–627).—[In English]. The elect. resistance of pure Ni and Fe, and Ni–Fe, Ni–Co, Ni–Cu, and Heusler's alloys was studied at room temp. and at temp. of liq. N and H in magnetic fields up to 22,000 Oe. Pure metals and alloys behave quite differently at low temp.: a max. in the magneto-resistance is found for alloys having ~1 Bohr magneton/atom. Pure metals show an increase in resistance with increasing field just as in the case of non-ferromagnetic metals. The value of the internal field, acting on the conduction electrons, is approx. equal to the induction. The positive difference between the resistance in a longitudinal and transverse field is explained by means of spin–orbit interaction. 12 ref.
—J. S. G. T.

*Testing the Interpretation of the Mechanism of Electrical Conductivity of Very Thin Metallic Films. Nicolas Mostovetch

(*Compt. rend.*, 1951, **233**, (5), 360–362).—Very thin metallic deposits have negative coeff. of resistance. The theory of contacts based on the tunnel effect explains these negative coeff. only by assuming that the distance between the grains varies inversely with the temp., leading to an expression of the form: $\sigma = A \exp(-bT)$, which is not confirmed by experiment. M. puts forward a possible interpretation, which assumes that the film consists of very small grains, that the work done by an electron moving between two grains is <0·1 eV., that the tunnel effect is negligible, and that the conductivity of the film is practically equal to that of the lacunæ.—J. H. W.

The Origin and the Relaxation Time Problem in the Electron Theory of Solid Conductors. Walter Schottky (*Ann. Physik*, 1949, [vi], **6**, 193–216).—A discussion of the classical theory of metallic conduction, the effect of Fermi–Dirac statistics being omitted. The question of the conditions under which it is appropriate to introduce a relaxation time τ (i.e. a free path) for the collision processes in an electron gas is considered, and it is shown that both the assumptions of the Riecke–Drude approach ("memory-extinguishing" collisions) or those of the Lorentz treatment (elastic collisions) justify this concept. In the more general case this is not so. The Lorentz–Boltzmann equation is derived and methods of determining the distribution function f in 6-dimensional phase space are considered. In the general case the relaxation time τ used in defining the collision operator is replaced by six partial relaxation times τ_J and a collision operator is thus defined for each perturbing function in terms of the appropriate τ_J.—J. W. C.

Photoconductivity in the Elements. T. S. Moss (*Proc. Phys. Soc.*, 1951, [A], **64**, (6), 590–591).—A letter. Thermal and opt. activation energies, E, of 10 photoconductive elements are compiled; they are closely the same in all cases. A qual. correlation between refractive index n and the threshold wave-length also exists. For Ge, Si, and diamond the relation $En^4 =$ const. holds approx. If it holds for grey Sn, recently reported to be photoconductive with E only 0·1 eV., then this has the high threshold wave-length of ~12 μ.—R. W. C.

*Empirical Correlation for a Thermoelectric Property of Supraconductors. M. C. Steele and Jules de Launay (*Phys. Rev.*, 1951, [ii], **82**, (2), 276–277).—A letter. It is pointed out that for Pb, In, Sn, and Tl, the thermo-e.m.f., E, of a supraconducting/normal junction of the same metal, gives an approx. const. value for $E_0\Theta^2/T_c^2$, where E_0 is the e.m.f. at $0°$ K., T_c the supraconducting transition temp., and Θ the Debye characteristic temp. Preliminary measurements of E_0 are given for Hg.—P. C. L. P.

On the Carriers of Electricity and Heat in Supraconductors. Z. Mikura (*J. Phys. Soc. Japan*, 1948, **3**, 338–342; *Physics Abs.*, 1950, **53**, 670).—[In English]. On the basis of the generally accepted concepts on metallic conduction and the structure of the intermediate state, it is argued that a carrier in a supraconductor has a very large elect. charge compared with the ordinary electron and a large mass proportional to the charge. The inference is that the electrons in a metal begin to condense (not necessarily in co-ordinate space, but rather in momentum space) into droplets as soon as supraconductivity takes place in the metal, and that each droplet behaves like a "giant electron".

*The Condition Determining the Transition Temperature of a Supraconductor. K. F. Niessen (*Physica*, 1951, **17**, (1), 33–42).—[In English]. In Heisenberg's theory of supraconductivity (*Ann. Physik*, 1948, [vi], **3**, 289; *Met. Abs.*, 1949–50, **17**, 424), the Fermi sphere, each element, h^3, of which is occupied by two normal electrons, is partly covered by vol. elements, h^3, each contg. only one supraconducting electron. Analogy between the phenomena and supraconductivity suggests a similar dependence upon temp. for the part, ω, of the Fermi sphere covered by Heisenberg's supraconducting layer and the relative magnetization, I/I_0 in ferromagnetic materials below the Curie temp. (I_0 being the max. magnetization at $T = 0$). Using this analogy, N. shows that the process of transition from the supraconducting to the norma

state comprises two parts, neither of which requires the expenditure of any energy, and thence deduces the thickness of the Heisenberg layer near the transition temp., and shows how the spherical extension of a layer hidden half-way in the Fermi sphere can be reduced suddenly to 0 at a definite temp. corresponding with Heisenberg's transition temp.

—J. S. G. T.

*Landau Diamagnetism and Meissner Effect [Supraconductivity of Metals]. A. Papapetrou (*Phil. Mag.*, 1951, [vii], 42, (324), 95–105).—The effect of one- and two-dimensional periodic magnetic fields on the free electrons of a metal is discussed in detail, but without taking account of the periodic electrostatic field of the lattice. If, in a supraconductor, the distribution of electrons in k-space is not spherical but shows prominences, these prominences will orient themselves to give max. paramagnetism. This may lead to a periodic internal magnetic field on a fine scale, which P. assumes to be a characteristic of the supraconducting state. Appn. of an external magnetic field will increase the energies of the electron states in the prominences and lead to their disappearance. A sufficiently large field will thus destroy supraconductivity.—P. C. L. P.

A Wave-Mechanical Model of the Supraconductor. H. Welker (*Z. Naturforsch.*, 1948, [A], 3, (8/11), 461–469; *Physics Abs.*, 1949, 52, 239).—W. constructs hypothetical wave-functions for the purpose of illustrating the type of electron interaction which may lead to supraconductivity. These wave-functions are degenerate, and magnetic interaction produces an energy gap. A qual. explanation of supraconductivity analogous to Bethe's theory of superstructures is given. It is essential for the model that the supraconductor should be a mixed conductor, as defect electrons are used to build up the hypothetical wave-function. Another consequence of the types of electron interaction assumed is a range of electron d within which supraconduction is possible. This is verified by comparison with the range of atomic vol. occupied by the known supraconductors. The max. electron d is, however, not in agreement with the theoretical prediction. Some conclusions concerning the lattice structure of supraconductors are indicated.

*On the Connection Between Supraconductivity and Mixed Conductivity. H. Welker (*Ann. Physik*, 1949, [vi], 5, (1/2), 1–13).—W.'s theory (*Z. Naturforsch.*, 1948, [A], 3, 461; preceding abstract) that magnetic interaction is responsible for supraconductivity leads to the conclusion that only " mixed " conductors, i.e. conductors with overlapping electron bands, can be supraconducting. The transition temp. $\propto$ the product of the translational zero-point velocities of the electrons and the positive holes. The total number of conduction carriers (electrons + positive holes) is estimated for the various elements and this number divided by the atomic vol. is plotted against the atomic number. The known supraconducting elements lie in a region bounded by straight lines which are not horizontal. Metals outside this region are either univalent with no positive holes, or have too small an electron concentration to produce a measurable transition point. A correction is calculated for the increase of ionic vol. with atomic number and when the reduced electron vol. are plotted the supraconductors then lie below a horizontal straight line. The connection between supraconductivity and electronic sp. heat and magnetic properties is discussed.

—J. W. C.

*A Non-Linear, Phenomenological Theory of Supraconductivity. M. von Laue (*Ann. Physik*, 1949, [vi], 5, (3/5), 197–207).—Math. A new form of the phenomenological theory which assumes, as did the earlier theories, the sepn. of conduction into ohmic and supraconducting parts. For the latter the London equations are retained, but the linear connection between momentum and c.d., formerly assumed, is replaced by a more general non-linear relation, which is not precisely defined. All essential conclusions of the older theories, so long as they refer to stationary conditions, are retained either completely or in essential features so that agreement with experiment is preserved. Differences must occur in alternating elect. fields when the amplitude of

oscillation becomes too large for the linear approximation to be valid. The new theory can include the notion of a max. c.d. and so approach the quantum theory of supraconductivity.

—J. W. C.

The Magnetic Threshold Value in the Theory of Supraconductivity. F. Möglich and R. Rompe (*Ann. Physik*, 1948, [vi], 3, 322–326; *Physics Abs.*, 1950, 53, 744).—The London relation between c.d. and vector potential is a soln. of the equation of motion of the plasma only as long as the kinetic energy of the drift velocity of the electrons $\frac{1}{2} mv^2 \ll \partial E_0/\partial n_0$, where E_0 is the zero point energy of the electron gas and n_0 the electron d. It is assumed that supraconductivity breaks down when $\frac{1}{2} mv^2 \simeq 0.01 \partial E_0/\partial n_0$. If one assumes for E_0 the value resulting from the Sommerfeld theory, magnetic threshold values of 10^4–10^5 are obtained. The fact that these values are 10–100 times larger than the observed values is ascribed to the fact that the Sommerfeld theory does not take account of the lattice field.

On the Theory of Supraconductivity. F. Möglich and R. Rompe (*Ann. Physik*, 1949, [vi], 6, 177–192).—A development of M. and R.'s theory of plasma oscillation (*ibid.*, 1948, [vi], 3, 322; preceding abstract). It is shown that this tends to a value for the energy d in agreement with the phenomenological theory of the Londons and von Laue. The interaction between the transverse vibrations of the plasma, which are the carriers of the supraconductivity, and the lattice disappears.—J. W. C.

*Cooling and Switching-on Processes According to von Laue's Theory. Gerhard U. Schubert (*Ann. Physik*, 1949, [vi], 5, (3/5), 213–236).—The fundamental equations of the von Laue form of the Londons' theory (*ibid.*, 1942, [v], 42, 65; *Met. Abs.*, 1946, 13, 315) are solved for the time variation of penetration depth. The first part of the paper develops the theory of the exclusion of the magnetic field from a supraconductor when supraconductivity is produced by lowering the temp. (Meissner effect), taking into account the time variation. The second part of the paper discusses the problem of the exclusion of a magnetic field which is suddenly applied to a supraconductor at const. temp.—J. W. C.

*The Energy–Momentum Tensor in the von Laue–London Electrodynamics of Supraconductors. Gerhard U. Schubert (*Ann. Physik*, 1949, [vi], 6, 163–168).—Math. The laws of conservation of energy and momentum for a supraconductor at rest can be written as a relation between quantities which are Lorentz-invariant. An expression for the force d is derived from the energy momentum tensor.—J. W. C.

*On the Theory of Incomplete Supraconductivity. Heinz Koppe (*Ann. Physik*, 1949, [vi], 6, 375–380).—The behaviour of a material, contg. small islands of a supraconducting phase in a normally conducting ground mass, is investigated theoretically. The equivalent circuit is a series connection of a supraconductor and a normal conductor, and the material will thus always show ohmic conductivity. The appn. of a magnet field has a marked influence on the conductivity, which falls from a high value, σ_1, to a lower value, σ_2, when the critical field strength for destruction of supraconductivity is reached. The apparent resistance of a cylinder of incompletely supraconducting material is calculated as a function of temp. and c.d.—J. W. C.

A Statistical-Mechanical Treatment of Conduction Electrons : An Attempt to [Deduce] the Theory of Supraconductivity. H. Ichimura (*J. Phys. Soc. Japan*, 1949, 4, 265–270; *Physics Abs.*, 1950, 53, 670).—[In English]. It is assumed that for metals for which the Brillouin zones overlap and the Fermi surface falls into this domain the inter-electronic interactions are most important for electrons near the Fermi surface. For these, a co-operative interaction of the Bragg–Williams type is assumed, i.e. the interaction energy/electron is assumed to be proportional to the number of electrons occupying special levels above the normal Fermi distribution. The nature of this energy is not discussed, and the proportionality const. is arbitrarily chosen. On this basis the thermodynamical potential and the sp. heat are calculated and found to represent correctly the behaviour of supraconductors near the critical temp.

*Electrostatic Plasma Oscillations in Metals. J. A. Kok (*Physica*, 1951, 17, (5), 543–547).—[In English]. The average elect. charge in a so-called " plasma " in an elect. discharge in gases is 0, since a " plasma " contains equal amounts of + and − charges. Electron oscillations in such plasma were discovered by Penning (*Nature*, 1926, 112, 301) and their frequency calculated by Langmuir (*Proc. Nat. Acad. Sci.* 1928, 14, 627). The equations of electrostatic plasma oscillations of electrons and ions in elect. discharges in gases are applied to metals, using the Fermi–Dirac law of distribution of electrons in metals and not that of Maxwellian statistics. Metallic-ion oscillations are shown to have some of the characteristics of plasma-ion oscillations in elect. discharges in gases ; they have an upper limit of frequencies of the same order of magnitude as the Debye max. frequency. 10 ref.—J. S. G. T.

*The Rediffused Component of Secondary Electron Emission from Metals. Pierre Palluel (*Compt. rend.*, 1947, 224, (21), 1492–1494).—The ratio of the number of electrons rediffused to the number of incident primary electrons gives the coeff. of rediffusion r, which is a certain fraction of the coeff. of secondary emission δ. P. has measured r for C, Al, Ca, Fe, Ni, Mo, Ag, W, and Pt over the range 1–18 kV. The curves obtained show that in each case r tends to a value which is practically independent of the primary velocity, irrespective of the nature of the metal surface, and that r increases with the atomic number.—N. B. V.

On the Mechanism of Electronic Rediffusion by Metals. Pierre Palluel (*Compt. rend.*, 1947, 224, (22), 1551–1553).—Cf. ibid., (21), 1492 ; preceding abstract. The coeff. of rediffusion r for a number of elements has been plotted against Z, the atomic number, for a primary velocity β of 0·25. It is found that for light elements ($Z < 30$) r increases linearly with Z, but beyond this r tends towards a const. value of 0·5. No periodicity was observed.—N. B. V.

Measurement of Electrical Fluctuations as a Method of Studying Processes in Metals. E. Ya. Pumper (*Izvest. Akad. Nauk S.S.S.R.*, 1949, [Fiz.], 13, (5), 596–614 ; *Physics Abs.*, 1950, 53, 703).—[In Russian]. By using a linear detector with large time-const., fluctuation processes could be detected in two types of system (ohmic resistance, thermionic tubes) which departed from the normal distribution law ; these are related to occasional phys. dis-equilibrium processes occurring in the system, and it is suggested that they occur at the expense of lattice energy.

Resistivity and Hall Constant of Semi-Conductors. Carl N. Klahr (*Phys. Rev.*, 1951, [ii], 82, (1), 109–110).—Cf. Jones, ibid., 81, (1), 149 ; *Met. Abs.*, this vol., col. 174. A letter. Generalized calculations of the resistivity and Hall coeff. are given, taking into account the mean free paths due to : (i) lattice scattering, (ii) ionized impurities, and (iii) non-ionized impurities. It is not assumed that the effects of these three factors are additive.—P. C. L. P.

*Fermi Levels in Semi-Conductors. R. A. Hunter, E. S. Rittner, and F. K. Du Pré (*Philips Research Rep.*, 1950, 5, (3), 188–204).—General formulæ for determining the Fermi level and the d of free-charge carriers in semi-conductors are derived. Special semi-conductor models are considered in detail, and a few appn. are discussed.—D. K. W.

The Term a/H in the Law of Approach [to Magnetic Saturation] and a New Theory of Magnetic Hardness. Louis Néel (*J. Phys. Radium*, 1948, [viii], 9, (5), 184–192).—Weiss found (ibid., 1910, 9, 373) that in strong fields (>1000 Oe.) the magnetization I of ferromagnetics approached the saturation value I_s in accordance with the equation $I = I_s(1 − a/H)$, where a is a coeff. termed the magnetic hardness. After showing that no satisfactory interpretation of the term a/H has so far been given, N. attributes it to the presence of cavities or non-magnetic inclusions, supporting this view by ref. to experimental results on porous sintered-Fe specimens. A method of calculation is developed which enables the law of approach to be determined in relation to the vol. of cavities ; in very strong fields (>100,000 Oe.) the approach varies with $1/H^2$, but in the range 2000–10,000 Oe. it varies with $1/H$, in accordance with experimental results. On the same basis a qual. interpretation is given of the great magnetic hardness of polyphase alloys, which is ascribed to segregation of non-magnetic phases and of inhomogeneous solid soln. The positive thermal variations observed by Sadron (*Ann. Physique*, 1932, [x], 17, 371 ; *Met. Abs.* (*J. Inst. Metals*), 1933, 53, 8) are also accounted for.—N. B. V.

*The Relation Between the Anisotropy Constant and the Law of Approach to Saturation of Ferromagnetic Materials. Louis Néel (*J. Phys. Radium*, 1948, [viii], 9, (6), 193–199).—As the field H increases, the magnetization I of a cubic polycryst. ferromagnetic approaches the saturation value I_s in accordance with the formula : $I = I_s(1 − b/H^2)$. A current theory states that $b = 8K^2/105I_s^2$, where K is the anisotropy const. N. shows that this theory is incorrect, as it neglects the magnetic interactions of crystallites having different orientations, and develops a new theory in which these interactions are taken into account. The formula so obtained is different from the previous one, though the latter is sufficiently accurate in the region of greatest practical interest (200–1000 Oe.). The value of b, however, is only about half that given above. The theory has been confirmed by new measurements carried out on well-annealed, high-purity Ni. The new formulæ enable K to be determined from measurements on polycryst. specimens.—N. B. V.

Dislocations, Cavities, and the Approach to Magnetic Saturation. William Fuller Brown, Jr. (*Phys. Rev.*, 1951, [ii], 82, (1), 94).—A letter. B. defends his paper on the effect of dislocations on magnetization near saturation (ibid., 1941, [ii], 60, (2), 139 ; *Met. Abs.*, 1942, 9, 38) against criticisms made by Néel (*J. Phys. Radium*, 1948, [viii], 9, 184 ; preceding abstract) which were accepted by Stoner (*Rep. Progress Physics*, 1950, 13, 142 ; *Met. Abs.*, 1950–51, 18, 253). B.'s formulæ are based on approximations which break down under high fields and cannot be integrated to an infinitely high value of the magnetic field, although valid over the range in which they were intended to be used. The integration over a stress singularity is also defended.

—P. C. L. P.

Properties of a Finely Powdered Ferromagnetic Having a Cubic Structure. Louis Néel (*Compt. rend.*, 1947, 224, (21), 1488–1490).—Math. N. shows that below a certain critical size magnetization takes place not by displacement of the domain boundaries but by rotation of the direction of the spin axes. This critical size is calculated to be 320 Å. for Fe.—N. B. V.

The Coercive Force of a Ferromagnetic Powder Having a Cubic Structure and Anisotropic Particles. Louis Néel (*Compt. rend.*, 1947, 224, (22), 1550–1551).—The theoretical considerations previously applied to round particles (ibid., (21), 1488 ; preceding abstract) are extended to ellipsoidal ones. For a compressed aggregate of Fe powder ($d = 4·5$), a coercive force of 500 gauss is calculated, in agreement with experimental values.—N. B. V.

The Effect of Thermal Fluctuations on the Magnetization of Very Fine Ferromagnetic Particles. Louis Néel (*Compt. rend.*, 1949, 228, (8), 664–666).—Cf. ibid., 1947, 224, 1488, 1550 ; preceding abstracts. Math. An expression is derived for the probability of spontaneous reversal of magnetization in a particle small enough to constitute a single domain, as a result of elastic deformation arising from thermal agitation.—N. B. V.

The Effect of Thermal Fluctuations on the Magnetization of Ferromagnetic Substances in Massive Form. Louis Néel (*Compt. rend.*, 1949, 228, (14), 1210–1212).—Math. The treatment of fine particles (ibid., (8), 664 ; preceding abstract) is applied to the displacement of Bloch boundary walls in solid materials, with similar results.—N. B. V.

*Magnetic After-Effect in the Rayleigh Domain. Jean-Claude Barbier (*Compt. rend.*, 1950, 230, (11), 1040–1041).—Experiments on ferromagnetic materials in massive form have shown that the decrease of $\sqrt{I_r}$ (I_r = residual magnetism) with time is proportional to the time elapsing after removal of the field and is independent of the field and of the period of appn. These results agree well with Néel's theory (*J. Phys. Radium*, 1950, 11, 49 ; *Met. Abs.*, 1950–51, 18, 15).—N. B. V

*Some Laws Relating to Magnetic After-Effect. Louis Lliboutry (*Compt. rend.*, 1950, **230**, (11), 1042–1044).—In weak fields the magnetic after-effect due to variations in the magnetic field can be determined by superimposing on it a small additional variation of the same sign. The results are related to Néel's theory (*J. Phys. Radium*, 1950, **11**, 49 ; *Met. Abs.*, 1950–51, **18**, 15).—N. B. V.

*Ferromagnetic Resonance. J. H. van Vleck (*Physica*, 1951, **17**, (3/4), 234–252).—[In English]. Kittel's fundamental formula, relating to ferromagnetic resonance, taking into account demagnetizing effects (*Phys. Rev.*, 1947, [ii], **71**, 270 ; 1948, [ii], **73**, 155) is discussed, and a macroscopic derivation of the formula is given. The " spectroscopic " splitting factor, g, in the formula for the resonance frequency, is not the same as the anomaly, g', in the gyromagnetic ratio. The relation $g - 2 = 2 - g'$, predicted by theory, is not confirmed by experiments, and although the cause of this discrepancy is not clear, it is stressed that anisotropic exchange forces of pseudo-dipolar structure may influence the breadth and proper frequency of resonance lines, particularly in non-cubic crystals. Experimental data on resonance in anti-ferromagnetics are very briefly discussed. 23 ref.—J. S. G. T.

The Atomic Magnetic Moments of Transition Metals and Their Alloys and the Interatomic Distances in Their Crystal Lattices. F. M. Galperin (*Izvest. Akad. Nauk S.S.S.R.*, 1949, [*Fiz.*], **13**, (5), 574–595 ; *Physics Abs.*, 1950, **53**, 690).—[In Russian]. The basic relationship for the atomic magnetic moment of a metal, $m = M + K_1(d_1 - D) + K_2(d_2 - D)$, is discussed, a theoretical explanation in terms of s–d electron transition or exchange being given. $M = 2(10 - \text{no. of}$ group in Periodic System), K_1 and K_2 are const., d_1 and d_2 are the distances from the atom considered to its nearest and next nearest neighbours, and D is the sum of the radii of the s and d orbits. Suitable modification permits the calculation of m from published structural data for pure metals and their solid soln. and for martensite and Ni_3Mn, Ni_3Fe, $MnCu_2Al$, and $CrTe$ as examples of partially and completely ordered alloys. In most cases calculated and observed moments agree to within 1–5%.

Introduction of the Idea of Intermittent Activation in the Electronic State of Ferromagnetics. Robert Forrer (*Compt. rend.*, 1950, **230**, (13), 1254–1255).—F. outlines his conception of interaction taking place intermittently between electrons in different states within the atom (e.g. $3d$ and $4s$ or $4p$), and on the basis of his theory explains the existence of ferromagnetism in Fe, Co, Ni, and certain alloys of Cr and Mn (CrTe, MnSb, &c.) and its absence in Cr and Mn themselves.
—N. B. V.

Collective Electron Ferromagnetism.—I.–II. H. Watanabe (*J. Phys. Soc. Japan*, 1948, **3**, 12–18, 317–322 ; *Physics Abs.*, 1949, **52**, 486 ; 1950, **53**, 682).—[In English]. [I.—] The magnetic properties of an electron model in which the d of levels in the energy band is const. with respect to the energy of electrons, are discussed with respect to the band form and the value of the exchange integral. The case where the d of levels is const. is worked out in detail. [II.—] The theory is used to calculate susceptibility above the Curie point and the sp. heat associated with ferromagnetism. The high-temp. susceptibilities are compared with experimental results for Ni and its alloys. Calculated electronic sp. heats for Ni and Co do not agree with the observed values.

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2 — PROPERTIES OF ALLOYS

Wrought Aluminium Alloys. Philip O'Keefe (*Materials and Methods*, 1951, **33**, (6), 89–104).—Materials and Methods Manual No. 71. A comprehensive review giving details of alloy compn., phys. and mech. properties, corrosion-resistance, methods of working and joining, and finishes. Detailed information is summarized in 8 tables.—J. W. C.

*What Is the Strength of Aluminium [Alloys] After Forming? L. V. Omelka and D. A. Paul (*Iron Age*, 1951, **168**, (7), 112–114).—O. and P. investigated the effects of stretching on the Y.P. and U.T.S. of various Al alloys during stressing parallel with the grain. The Y.P. and U.T.S. increased most in 52S, followed by K-150, 3S, and 2S in that order. The rate of work-hardening is highest in 52S, but K-150 shows the greatest percentage increase in Y.P. for a given amount of stretch.—J. H. W.

*Stress/Strain and Elongation Graphs for Alclad Aluminium-Alloy 24S-T86 Sheet. James A. Miller ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2094), 31 pp. ; *Appl. Mechanics Rev.*, 1951, **4**, 227).—Results of tests on duplicate longitudinal and transverse specimens of Alclad 24S-T86 sheets, with nominal thicknesses of 0·032, 0·064, and 0·125 in., are presented in the following forms : tensile and compressive stress/strain graphs and stress/deviation graphs to a strain of ~1% ; stress/strain graphs for tensile specimens tested to failure ; graphs of local elongation and of elongation against gauge-length for tensile specimens tested to fracture. The stress/strain and stress/deviation graphs are plotted on a dimensionless basis to make them applicable to sheets of this alloy with yield strengths which differ from those of the test specimens.

*Static and Impact Strengths of Riveted and Spot-Welded Beams of Alclad 14S-T6, Alclad 75S-T6, and Various Tempers of Alclad 24S Aluminium Alloy. H. E. Grieshaber ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2157), 44 pp. ; *Appl. Mechanics Rev.*, 1951, **4**, 284).—Moduli of failure in bending were found to be about the same as the material tensile strength on riveted beams, but less on spot-welded beams. The soundness of some welds, however, would not pass present specifications for aircraft welding. Ageing of beams after assembly was found of no advantage for static or impact strength. Impact strengths for riveted beams exceeded those for spot-welded beams in all alloys ; static strengths averaged 12% better. No correlation was found between toughness determined from tensile properties and ability to resist impact.

*Constitutional Diagram of the Aluminium–Copper–Magnesium–Silicon System. D. A. Petrov and N. D. Nagorskaya (*Zhur. Obshch. Khim.*, 1949, **19**, (11), 1994–2037 ; *C. Abs.*, 1950, **44**, 6371).—[In Russian]. The Al-rich vertex of the quaternary diagram was studied down to 50% Al by means of cooling curves taken with a Kurnakov pyrometer and by micro-examination. Previous data on the 3 Al-contg. ternary systems and on the quaternary system are reviewed. A 0·5-mm. Pt/Pt-Rh thermocouple was used to record the course of cooling, which took 40–70 min. The entire constitutional diagram is shown in 2 ways : as a horizontal and 3 vertical projections, and in perspective within a tetrahedron. Horizontal sections are shown at 90, 80, 70, 60, and 50% Al, and vertical (temp.) sections at 60 Al-8 Si, 80 Al-2 Si, 80 Al-0·2 Si, 70 Al-0·2 Si, 90 Al-4 Mg, 50 Al-46·5 Cu, 50 Al from 35 Cu-15 Si to 16 Mg-34 Si, 90 Al from 5 Cu-5 Si to 0 Cu-8 Si, 80 Al from 17·3 Cu-2·5 Mg to 18·5 Cu-0 Mg. This work suggested corrections of the best data available on the ternary diagrams. One invariant point in the Al–Mg–Si system, L (liquid) $\rightleftharpoons$ Al + Mg_2Si + Al_3Mg_2, is near 0·3 Si, 34·6 Mg ; the second, $L \rightleftharpoons$ Al + Si + Mg_2Si, is near 10 Si, 3·9 Mg. The invariant point in the Al–Cu–Si system is near 5·2 Si, 29·0 Cu. The W phase in the quaternary diagram contains Cu : Mg : Si in the atomic ratio 4 : 5 : 4 and between 30 and 40 wt.-% Al. There are 3 quaternary eutectic reactions involving Al : (1) $L \rightleftharpoons$ Al + $CuAl_2$ + Si + Mg_2Si at 505° C. and 28·3 Cu, 3·1 Mg, 4·0 Si ; (2) $L \rightleftharpoons$ Al + $CuAl_2$ + S + Mg_2Si at 500° C. and 29·6 Cu, 7·15 Mg, 0·45 Si ; (3) $L \rightleftharpoons$ Al + T + Al_3Mg_2 + Mg_2Si at 444° C. and 1·5 Cu, 32·9 Mg, 0·3 Si. There are 3 quaternary peritectics involving Al : (1) $L + Mg_2Si \rightleftharpoons$ Al + Si + W at 521° C. and 16·5 Cu, 6·0 Mg, 7·5 Si ; (2) $L + Mg_2Si \rightleftharpoons$ Al + $CuAl_2$ + W at 510° C. and 28·3 Cu, 3·6 Mg, 3·7 Si ; (3) $L + S \rightleftharpoons$ Al +

$Mg_2Si + T$ at 505° C. and 10·0 Cu, 25·5 Mg, 0·3 Si. S and T are phases in the Al–Mg–Cu diagram contg. 45 Cu, 17 Mg and 20 Cu, 30 Mg, resp. The vol. of primary crystn. of Al, Si, $CuAl_2$, Mg_2Si, W, S, T, and Al_3Mg_2 are limited by 19 surfaces of secondary crystn., which are shown with 18 lines of tertiary crystn. The phases U and Al_3Mg_4 that appear in the Al–Cu–Mg diagram were not considered in this work. Complete descriptions are given of primary, secondary, tertiary, and quaternary crystn. for the 90% Al and 60% Al horizontal sections; brief descriptions are given for the other three. In photomicrographs of the 60% Al sections, it is shown that various reactions can occur in a given specimen because of segregation. Brief descriptions of the vertical sections are given.

*Experiments on the Reaction of Aluminium–Magnesium Alloys with Steam. (Swain). See col. 295.

The Aluminium–Nickel Equilibrium Diagram. H. Groeber and V. Hauk (*Z. Metallkunde*, 1950, 41, (8), 283–284).—Annotated equilibrium diagrams—in wt.-% and at.-%—are given, embodying the results of recent investigations. Subsidiary diagrams show the variation of: (1) Curie temp. and lattice const. of Ni with addn. of up to 30% Al, and (2) the heats of formation over the whole range of compn.

—E. N.

*Physico-Chemical Investigation of Aluminium–Silicon–Lithium Alloys. E. A. Boom (*Doklady Akad. Nauk S.S.S.R.*, 1949, 67, (5), 871–874).—[In Russian]. Experiments, designed to demonstrate that the dispersed structure of modified Silumin-type alloys represents a ternary eutectic consisting of Al, Si, and a ternary silicide, are reported with Li, which acts in a similar way to Na, though less effectively. Li was sealed in a hole drilled in an Al–9% Si casting, which was then kept for 20–30 min. at 577° C., cooled, and sectioned. Micro-sections showed two diffusion zones followed by the usual modified structure. Unstable Li_3Si crystals were identified in the first diffusion zone by evolution of SiH_4, and comparison with synthetic Li_3Si. Thermal analysis of Al–Si alloys contg. 1–20% Si showed, on addn. of 1% Li, three thermal effects, and three phases were metallographically identified: primary crystals, binary eutectic, and ternary eutectic. Chem. analysis of the last, after removal of undissolved Si and solid soln., yielded the formula $Li_3Al_2Si_2$. Microhardness tests gave 38·3 kg./mm.² for the solid soln., 421 kg./mm.² for Si, and 946 kg./mm.² for the ternary silicide, which was then identified in the second diffusion zone.—A. G.

*Anelastic Effects Arising from Precipitation in Aluminium–Zinc Alloys. A. S. Nowick (*J. Appl. Physics*, 1951, 22, (7), 925–933).—The internal friction of supersaturated polycryst. Al–Zn wires (21 at.-% Zn) was measured by means of a Kê pendulum. The damping is high except after prolonged ageing. The damping/temp. curve is extrapolated to high temp. at which pptn. excludes direct measurement, by computation from static torsional creep measurements at lower temp., and in this way the absence of a damping peak is demonstrated. This implies the existence of a relaxation time spectrum without a peak even for very high relaxation times (a conclusion confirmed by measurements of torsional elastic after-effect). This behaviour can be explained satisfactorily only by the occurrence of " coupled relaxations ". The mechanism proposed for these involves the fine plates precipitated close to each other at sufficiently high temp.; relaxation of shear stress across one plate triggers relaxation at its neighbours. The occurrence of heavy damping is in fact correlated with the onset of such discontinuous pptn. Computation from the elastic after-effect measured at 52°–72° C. gives a value of 25·4 kg.cal./mole for the activation energy of diffusion of Zn in Al, which agrees well with direct high-temp. measurements. Atomic pair reorientation by diffusion is shown to play only a minor role in causing damping.—R. W. C.

*The Photoconductivity of Bismuth Sulphide and Bismuth Telluride. A. F. Gibson and T. S. Moss (*Proc. Phys. Soc.*, 1950, [A], 63, (2), 176–177).—A letter. Experiments by G. and M. do not support American claims that Bi_2S_3 and Bi_2Te_3 are photosensitive to wave-lengths as long as 7 μ.

Measurements indicate that these Bi compounds behave similarly to analogous compounds of Pb, but are inferior to them as photodetectors.—N. B. V.

*Non-Metallic Inclusions in Ferrochromes. R. Zoja (*Rev. Mét.*, 1951, 48, (8), 631–638).—Z. carried out an extensive investigation of the characteristics of non-metallic inclusions in 48 specimens of ferrochrome, contg. mostly: Cr 52·60–97·30, Si trace–3·66, Mn trace–0·27, S 0·005–0·027, P 0·010–0·059, and C <1%. He concluded that these inclusions are formed by the two constituents, α and β, being two soln. at the limit of solubility or derived therefrom by reduction. By reheating at increasing temp., all the structures of the untreated ferrochrome can be reproduced. The first phase in the process of transformation is very probably a simple sepn. of droplets or very small metallic dendrites of a supersaturated soln. Si is the most important reducing element. A very characteristic structure is obtained at sufficiently high temp. with Si ∼1%, namely, a ring of pure SiO_2 contg. a single metallic globule composed of all the products of reduction.—J. H. W.

*The Constitutional Diagram of the Chromium–Tungsten System. H. T. Greenaway (*Commonwealth Australia, Dept. Supply Development, Aeronaut. Research Lab. Rep.*, 1951, (SM.-134), 17 pp.).—Metallographic, X-ray, and thermal analyses were carried out. Alloys contg. up to 60% W were melted in a W-arc furnace; above 60% W the alloys were made by powder metallurgy. The liquidus rises, slowly at first, and then more rapidly, from the m.p. of Cr towards that of W. Above 1500° C., the solid alloys form a continuous solid soln.; below 1500° C., the single phase breaks down into two limited solid solns., thus forming a " loop " in the diagram. After homogenizing for 6 hr. at 1600° C., two-phase alloys at the Cr-rich end increase in hardness when annealed for up to 460 hr. at 1000°–1250° C.—S. M.

High-Saturation Magnetic Alloy [Cobalt–Iron and Nickel–Iron] with a Rectangular Hysteresis Loop. J. F. Libsch and Eberhard Both (*Elect. Eng.*, 1951, 70, (5), 420–421).—Hysteresis characteristics of 50 : 50 Co–Fe and Ni–Fe alloys produced by powder metallurgy are compared, and the effect of magnetic annealing on Co–Fe alloys of various compn. is illustrated.—J. R.

*Effects of an Ageing Treatment on Life of Small Cast Vitallium Gas-Turbine Blades. Charles A. Hoffman and Charles Yaker ([U.S.] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2052), 33 pp.).—An investigation was carried out to determine the effects of an ageing treatment on the life of small cast Vitallium gas-turbine blades operated at a blade temp. of approx. 1500° F. (815° C.) and a stress of 20,000 lb./in.² at the blade-failure plane. Twenty blades aged for 48 hr. at 1500° F. were compared with 33 unaged blades. Ageing, which has been reported to harden cast Vitallium and to improve its stress–rupture life, apparently improved the time for initial blade failure, the average life, and the uniformity of life of blades used in the investigation. The lives of the last blades to fail were not appreciably affected by ageing treatment. Statistical analysis of blade-life data did not indicate a significant improvement in mean life or uniformity of life of blades. This fact, however, does not necessarily prove that ageing is without beneficial effects, but rather indicates that further investigation is desirable to obtain more conclusive results. A comparison of the lives of unaged Vitallium blades with stress–rupture data for Vitallium bars evaluated at substantially the same conditions indicated a relation between stress–rupture life and blade life. Both samples were progressively hardened by pptn. during operation. After ∼35 hr. of operation, they were at the same hardness, which increased slightly thereafter. Met. examination revealed that blade failure was initiated by intercryst. cracking.—AUTHORS.

Some Notes on Aluminium Bronzes. David D. Stead (*Australasian Eng.*, 1951, (March), 44–51).—The commercial alloys are of two types: (a) the α or single-phase alloys, contg. 4–7·5% Al, and (b) the α/β or two-phase alloys, known commercially as " duplex " or " complex " bronzes, generally contg. 9–11% Al, plus other elements. The commonest

alloying elements added to improve the properties are Fe, Mn, Ni, and Pb, but Si, Be, Cr, and Zn are also used. The effects of these elements are described. Brief notes on the following subjects are included : phys. properties, heat-treatment, corrosion-resistance, casting, forging, cold working, joining, machinability, and appn. 22 ref.—T. A. H.

*Manufacture and Properties of Resistance Wires of Composition Cu 82·5, Mn 12·0, Al 4·0, Fe 1·5%. A. Krupkowski and Z. Misiołek (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1950, 2, (4), 273–284).—[In Polish]. The charge consisted of Cu (99·98% purity), Al (99·5%), 90 : 10 Cu–Mn and 60 : 40 Fe–Mn alloys. These were melted in a crucible furnace, poured at 1210°–1240° C., and annealed to homogenize them at 900° C. for 14–16 hr. The cast rods were then cold rolled from 13 to 8 mm. dia., and drawn without annealing in 26 passes to 0·99 mm. dia. (total reduction 99·4%), and annealed at different temp. from 100° to 700° C. It is recommended that intermediate anneals be given after each 50% reduction, and the wire annealed at 400°–500° C. for 2 hr. to produce the lowest thermal coeff. of resistivity and min. thermoelect. effect against Cu (−0·40 μV./°C.). Mech. properties agree with those quoted in the literature for this alloy. 6 ref.—N. H. P.

[Nickel Brass] Zip-Fastener Wire. H. H. Symonds (*Metal Ind.*, 1951, 79, (6), 110).—Tests made on seven samples of Ni-brass zipper material of compn. approx. : Cu 58, Ni 18%, Zn balance, for hardness, grain-size, wire size, and compn. showed that fracture was primarily due to excessive hardness of the wire, contributory causes being undercutting of the elements during stamping and excessive Fe content in one of the samples.—J. H. W.

Physical-Property Measurements of Liquid-Metal Alloys : Literature Survey on Gallium. C. T. Ewing and R. E. Leed (*U.S. Atomic Energy Commission Publ.*, 1948, (NP–1197), 10 pp.).—Includes data on methods of purification, *d*, viscosity, surface tension, vapour pressure, elect. conductivity, and heat capacity. Ga is reported to dissolve, alloy, or react with most metals. Quartz appears to be the best substance for contg. Ga. Ref. are given to the properties of binary alloys of Ga with Zn, Cd, Hg, Sn, Pb, Al, Cu, Ag, and O.
—B. W. M.

*The AuCu Transformation in Ternary Alloys. Ernst Raub and Paul Walter (*Z. Metallkunde*, 1950, 41, (8), 240–243).—An X-ray study was made of the effects of ternary addn. of Cd 5, 7, 8, and 11 at.-%, Ti 7·2, Mg 6·6, Al 8·15, and Mn 12·9 at.-%, on the order-disorder transformation temp. of an alloy contg. Au and Cu in the atomic ratio of 1 : 1. The results show that the critical temp. of ordering is : (1) perceptibly lowered by addn. of Cd and Mn which, like Ag, are appreciably soluble in the binary alloy at its normal transformation temp., and (2) almost unaltered in the presence of Al, Mg, and Ti, which are only slightly soluble in the binary alloy at its normal transformation temp. 11 ref.—E. N.

*The Alloys of Gold with Cobalt and with Iron. Ernst Raub and Paul Walter (*Z. Metallkunde*, 1950, 41, (8), 234–238).—A report of a thermal, microscopic, X-ray, and dilatometric study of the systems Au–Co and Au–Fe. The results are described in detail and revised equilibrium diagrams are given. (1) *Au–Co*.—This is a simple eutectiferous system with the liquidus temp. falling from 1215° C. at 70 wt.-% Au to the eutectic at 995° C. and 90 wt.-% (72·9 at.-%) Au, and thereafter rising again to 1017° and 1038° C. at 95 and ~98 wt.-% Au, resp. The solid solubility of Au in Co is 2, 2·5, 1·8, and 1 at.-% at 1350°, 1200°, 995°, and 850° C., resp., the system thus being one having a retrogressive solid soln. curve. The Au-rich solid soln. contains 97·3 and 99·9 wt.-% Au at 1040° and 400° C., resp. The lattice const. of neither solid soln. obey Vegard's law. (2) *Au–Fe*.—This system contains no eutectic. There is a peritectic at 1168° C. with a compn. range of 8·1–43 at.-% Au, and a eutectoid at 903° C., 2·3 at.-% Au, with the eutectoid line extending to 45 at.-% Au. The solubility of Au in α-Fe after quenching from 1350°, 1170°, 1100°, 890°, and 500° C. is 4·6, 7·9, 6·1, 2·1, and 0·1 at.-%, resp. The saturated Au-

rich solid soln. contains 25·5, 32·0, 44·5, and 83·7 at.-% Au at 1168°, 1100°, 900°, and 400° C., resp. The α/γ transformation does not pass through a min., and the f.c.c. γ-Fe solid soln. is not stable below the transformation temp., and samples which have been very rapidly quenched show only the b.c.c. structure of α-Fe. 15 ref.—E. N.

*The Age-Hardening of Iron–Gold Alloys. Werner Köster and Eberhard Braun (*Z. Metallkunde*, 1950, 41, (8), 238–240).—A study is reported of the phenomena—changes in hardness and coercive force—which occur in an alloy contg. Fe 90 and Au 10 wt.-% on ageing. The alloy was prepared from carbonyl Fe powder and fine Au by melting under an atmosphere of H and cooling *in situ*, after which it was removed, cold rolled 60%, homogenized by heating for 4 hr. at 1100° C., and slowly cooled. Specimens were then soln.-treated at 1200° C., quenched—after which the hardness was 280 kg./mm.² —and aged at 400°–700° C. for periods up to 30 hr. On ageing at 400° C. the hardness rose continuously and did not attain a max. even after 30 hr.—then 420 kg./mm.² ; the coercive force increased only slightly. With increase of temp. the time taken to attain max. hardness decreased (16 hr. and 30 min. at 450° and 500° C., resp.) and the max. hardness attainable also decreased ; reversion phenomena set in when extended ageing times were employed. The coercive force behaved in a similar manner, except that the max. value was attained by ageing for 30 hr. at 560° C., or for shorter periods of time at higher temp. The activation energy for the age-hardening process is ~72,000 cal./mole, compared with a value of ~33,000 cal./mole for the similar process in Fe contg. Cu 1%. 9 ref.—E. N.

*Experiments on the Influence of [Calcium and] Hydrogen on the Grain-Size of [Chill-Cast] Magnesium and Magnesium Alloys. Walter Mannchen (*Z. Metallkunde*, 1950, 41, (8), 255–260).—Experiments are described which show that addn. of ~0·4% Ca to melts of pure Mg, and its binary alloys with Zn (1–5%) and Mn (1·8%) produce exceptional grain refinement and appreciable increases in the tensile properties of chill castings, provided that after such addn., and before casting, the melts are held for some time (20–60 min.) at 760° C. This "holding-period effect," as it is called (and the time necessary for its appearance), was found to be related to the H₂ content of the molten Mg ; when this was raised (to 30–40 c.c./100 g.) by the bubbling through of H₂, provided that the melt was held at const. temp. for a few min. after the H₂ was introduced, the metal could be poured immediately. This led to the conclusion that the effect is related to hydride formation, such hydride particles acting as crystn. nuclei. With this in mind, other hydride-forming elements—Se, Te, S, Pb, Bi, Zn, and Cl—were tried, and found to produce similar results when, with the exception of Zn, the addn. were of the order of 0·2–0·4% ; in the case of Zn much greater addn. (~2%) were necessary, however. In the case of Cl the effect appeared to be related to the Si content of the melt, and was a max. when ~0·04% Si was present. When Al was present the phenomenon did not occur. 9 ref.—E. N.

*Solid Solutions of Magnesium in Lead. N. S. Kurnakov, S. A. Pogodin, and T. A. Vidusova (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1947, 15, 74–79 ; *C. Abs.*, 1950, 44, 6373).—[In Russian]. Cf. *ibid.*, 1933, 6, 266. The limits of solid solubility of Mg in Pb were determined as 5·6 at.-% at 245° C., 4·1 at.-% at 220° C., and 2·2 at.-% at 160° C. (0·69, 0·50, and 0·26 wt.-%, resp.). The solubility of Mg in Pb is in agreement with the Schröder-Le Chatelier equation. By extrapolation, the saturation solubility of Mg in Pb was found to be 5·9 at.-% at the eutectic temp. (253° C.) and 1·0 at.-% at 100° C. (0·73 and 0·12 wt.-%, resp.). The hardness of aged Pb–Mg alloys was 15 B.H.N. at 0·5–0·7 wt.-% Mg, and 20 B.H.N. at 1·4 wt.-% Mg.

*Initial Magnetization of a Single Crystal of the Compound Mn₂Sb Along the Axes of Easy and Difficult Magnetization. Charles Guillaud, Roger Bertrand, and Roger Vautier (*Compt. rend.*, 1949, 228, (18), 1403–1405).—Measurements made between +16° and −78° C. are reported which can be represented best by Rayleigh's law $J = aH + bH^2$. The

directions of easy and of difficult magnetization are interchanged at −33° C.—N. B. V.

*The Alloys of Molybdenum and Tantalum. G. A. Geach and D. Summers-Smith (*J. Inst. Metals*, 1951–52, **80**, (3), 143–146).—An investigation of the binary system Mo–Ta has shown that these metals form a continuous series of solid soln. This is to be expected on theoretical grounds, as both metals crystallize with a b.c.c. structure and the atoms are very similar in size. No superlattice was detected and no anomalies occur between the true and X-ray *d*. Approx. m.p. 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.—AUTHORS.

*Intermetallic Compounds in the System Molybdenum–Beryllium. Samuel G. Gordon, James A. McGurty, Gilbert E. Klein, and Walter J. Koshuba (*J. Metals*, 1951, **3**, (8), 637–638).—A report of a study of the solid state reactions between Mo and Be at elevated temp. Solid diffusion at a Mo–Be interface was effected by sealing a Be rod inside a Mo capsule and heating at a temp. of 2000° F. (1100° C.) for 100 hr.; microscopic and X-ray detn. were made of the Mo–Be phases found and were compared with those obtained from solidified Mo–Be powder melts of known compn. The results : (1) verify the existence of the intermetallic compound $MoBe_2$; the purple-tinged tabular hexagonal crystals have the following unit cell const.: $a_0 = 4.433$; $c_0 = 7.341$ Å.; $Z = 4$; $\rho = 6.06$; $d = 5.92$ g./c.c.; space group, D^4_{0k} — $C6/mmc$; and (2) reveal the existence of a new intermetallic compound $MoBe_{13}$, which differs from other intermetallic compounds of the type MBe_{13} in being of prismatic hexagonal instead of cubic habit; single-crystal measurements show it to have the following const. : $a_0 = 10.27$; $c_0 = 4.29$ Å.; $Z = 4$; $\rho = 3.13$; $d = 3.13$ g./c.c.; probable space group $P42$. 6 ref.—E. N.

*The Value of the Spontaneous Magnetization of Binary Nickel Alloys as a Function of Temperature. J. J. Went (*Physica*, 1951, **17**, (6), 596–602).—[In English]. The relation of spontaneous magnetization, I_s, to temp., T, was studied for a large number of binary alloys of Ni with different amounts of Cu, Pd, Co, Fe, Mn, Cr, Mo, W, V, Al, Si, and Sn. The graph of I_s against T for pure Ni is more concave towards the T-axis than that for any of the alloys, except in the case of completely ordered alloys, e.g. slowly cooled Ni_3Fe. The results of all the experiments are contained in one diagram, in which the Brillouin function, $Q = I_T/I_0$, is plotted as a function of the concentration of the different alloying elements. For pure Ni the form of the I_s/T curve is explained by the occurrence of an order–disorder phenomenon of the magnetic moments, in which these moments can be arranged only parallel or anti-parallel. It is suggested that all the other I_s/T curves are the result of this order–disorder phenomenon and statistical fluctuations of dissolved atoms. In the case of Ni_3Fe two different order–disorder phenomena (one crystallographic, the other magnetic) act simultaneously. 11 ref.—J. S. G. T.

*Linear Magnetostriction of Homogeneous Nickel Alloys. J. J. Went (*Physica*, 1951, **17**, (2), 98–116).—[In English]. The saturation magnetic induction and saturation linear magnetostriction at 375°, 290°, 90°, 20°, and 0° K. (extrapolated) in fields up to 7000 Oe., Curie temp., and the number of magnetons/atom, of 41 homogeneous Ni alloys were studied and the results tabulated. Compn. of all the alloys are given. The homogeneous binary Ni alloys can be divided into three groups as regards their magnetostrictive behaviour. Group 1, comprising Ni alloys with various proportions of Al, Si, V, Cr, Cu, Mo, Sn, and W, in which the saturation induction decreases with increase of the alloying element. In this group the saturation linear magnetostriction, λ_s, increases continuously from the large negative value for pure Ni to zero. For all the different alloys λ_s is related to the number of Bohr magnetons/atom, β, by the relation $\lambda_s \times 10^6 = -70\beta + 6$. Magnetostriction is not very anisotropic for all these alloys. Group 2, in which the magnetic saturation increases with the amount of added element, consists of Ni alloys with Mn, Fe, or Co. For small

amounts of these elements, λ_s is given by the equation $\lambda_s \times 10^6 = 9\beta - 90$. Larger Mn contents give deviations from this linear relation, and the magnetostriction effect changes sign at about room temp. Deviations are also observed in the Ni–Co alloys with higher Co content. The magnetostriction is strongly anisotropic, λ being negative for all compn. in the preferred direction of magnetization, but in the difficult direction changes its sign from − to + in the neighbourhood of 1 Bohr magneton/atom. Hence Snoek's hypothesis that magnetostriction must be isotropic and zero for all alloys with 1 Bohr magneton/atom (*Nature*, 1949, **163**, 837 ; *Met. Abs.*, 1949–50, **17**, 17) is not confirmed. The decrease of the $+ \lambda_s$ for large Co contents is connected with the formation of a hexagonal phase. Group 3 consists of Ni–Pd alloys. In these the number of Bohr magnetons/atom is practically independent of the Pd content. Magnetostriction increases with increasing Pd content and satisfies all normal conditions. In ternary alloys consisting of Ni with elements of Group 1 and 2, or Group 2 and 3, the different components make independent and additive contributions to the magnetostriction and saturation induction. The most striking result is that all the magnetostrictive values, except for the Ni–Pd alloys are less negative than that of pure Ni. 13 ref.—J. S. G. T.

Nickel–Aluminium Alloy Combines Strength and Corrosion-Resistance. Philip O'Keefe (*Materials and Methods*, 1951, **33**, (4), 73–77).—Duranickel, a Ni–Al alloy contg. 4·4% Al and small quantities of other metals, has a high corrosion-resistance and a strength comparable to that of heat-treated alloy steel. Detailed tables are given of phys. const. and mech. properties and appn. are described.—J. W. C.

*Torque Curves and Other Magnetic Properties of Alcomax. K. Hoselitz and M. McCaig (*Proc. Phys. Soc.*, 1951, [B], **64**, (7), 549–559).—Using test specimens with coarse columnar grains and a [100] fibre texture, the magnetic anisotropy of Alcomax III has been studied by means of a torque magnetometer. According to the conditions of cooling through the Curie point, the anisotropy was of cubic or of tetragonal type. It was shown directly that the direction of the preferred magnetization was a [100] direction rather than the applied field direction, when the two did not coincide.—R. W. C.

*On the Magnetic Properties of Platinum–Iron Alloys.—II. A. Kussmann and G. (Grfn.) v. Rittberg (*Ann. Physik*, 1950, [vi], **7**, (3/4), 173–181).—Cf. Graf and K., *Physikal. Z.*, 1935, **36**, 544 ; *Met. Abs.*, 1935, **2**, 581. Measurements of the magnetic properties were made at room temp. on quenched and slowly cooled alloys over the whole compn. range, and the Curie points and magnetostrictive coeff. were also determined. Changes of saturation intensity and Curie point were found in the range 50–63 wt.-% Pt as the result of the formation of an Fe_3Pt superlattice. Very high values of magnetostriction were found in this range ($\Delta l/l = 180 \times 10^{-6}$ at ~62 wt.-% Pt). The best permanent-magnet properties were found at ~70 wt.-% Pt where the coercive force = 1900 Oe. and the remanence = 6000 gauss. Previous work showed the alloys with >90 wt.-% (72 at.-%) Pt to be paramagnetic, but in the present investigation a small range of alloys from 94 to 98 wt.-% Pt were found to be ferromagnetic at low temp. The Curie temp. of these alloys was ~−100° C. and the saturation magnetization reached a value about equal to that of Ni. There is a paramagnetic region between the two ferromagnetic regions.—J. W. C.

*On the Ferromagnetic Platinum–Manganese Alloys. M. Auwärter and A. Kussmann (*Ann. Physik*, 1950, [vi], **7**, (3/4), 169–172).—Alloys with 6–16 wt.-% Mn were prepared in a vacuum H.F. furnace from 99·9% pure Pt and 99·5% pure Mn, and were homogenized by annealing at 1200° C. for 10 hr. Magnetic measurements were made up to the Curie point, and dilatometric heating and cooling curves were plotted up to 950° C. Debye–Scherrer photographs were used for phase identification and lattice-spacing measurements. The results show the solid soln. extends to 14 wt.-% Mn, beyond which point there is a two-phase field. Superlattice lines were observed between 6 and 12 wt.-% Mn

(18·4 and 32·6 at.-% Mn) and the disordering temp. was found to be ~900° C. The max. saturation intensity ($4\pi J \cong 8000$ gauss) was attained at the compn. Pt_3Mn, and the ferromagnetism is therefore ascribed to the superlattice, which is of the $AuCu_3$ type. Curves of saturation intensity against compn. are given for temp. of —193° and +20° C. The Curie temp. increases continuously across the one-phase region. Neither the formation of the superlattice nor the appearance of ferromagnetism could be prevented by quenching.—J. W. C.

*The Electrical Resistance of Silver–Manganese Alloys at Low Temperatures. I.—The Resistance from 273° to 1° K. II.—The Resistance in a Magnetic Field. A. N. Gerritsen and J. O. Linde (*Physica*, 1951, 17, (6), 573–583 ; 584–595).— [In English]. [I.—] The elect. resistance of Ag–Mn alloys with concentrations of Mn up to ~20 at.-% was determined at 273°–1° K. The resistance decreases with temp. down to the lowest temp. and deviations from Matthiessen's law are large, anomalies occurring in the resistance/temp. relation at 0°–20° K. The resistance/temp. graphs for the most dil. alloys show, first a min. and then a max. with decrease of temp. As a possible explanation, G. and L. suggest that the energy of the unfilled d-level of the Mn atom may be a few cm.$^{-1}$ above the Fermi level of the s-electrons in the metal. 12 ref. [II.—] The elect. resistances of dil. Ag–Mn alloys were studied at temp. of 1°–20° K. in longitudinal and transverse magnetic fields up to 25 kOe. Below 10° K. the magnetic field reduces the resistance to less than the values in zero field for all alloys except that contg. 0·02 at.-% Mn. In a field of 20 kOe. there is a decrease of resistance for all concentrations up to 4·35 at.-% Mn at temp. <20° K. For any wire, the resistance in a field of 20 kOe. is a min. in the same temp. region where a decrease in zero field is found. The results do not appear to contradict the suggestion of the existence of a Mn d-level at a distance from the Fermi energy at liq. He temp. A remarkable decrease of resistance occurs in a magnetic field at the lowest temp., and differences of resistance in a transverse and in a longitudinal field appear to vanish at 0° K. 8 ref.—J. S. G. T.

*A Study of Some [Sintered] Alloys of Titanium. W. E. Kuhn, H. V. Kinsey, and O. W. Ellis (*Trans. Canad. Inst. Min . Met.*, 1950, 53, 54-67 ; in *Canad. Min. Met. Bull.*, 1950, 43, (454), 74-87).—An account is given of an investigation into the properties of certain binary alloys of Ti contg. up to ~10% of either Cr, Co, Fe, Mn, Mo, Ni, or W, made from powders. Experimental data on sintering, transverse and tensile strength, forgeability, relative resistance to deformation, creep-rupture properties, hardness, and oxidation-resistance are given, together with the results of an investigation of Ti compacts made from seven different lots of TiH_2. The metallography of sintered Ti compacts and binary Ti alloys is considered in detail, and the relationship between the light-etching areas and the properties of Ti compacts is discussed.—W. A. M. P.

*Frictional Properties of Tungsten Carbide and of Bonded Carbides. K. V. Shooter (*Research*, 1951, 4, (3), 136-139).— The frictional properties of WC and of four different bonded carbides are examined, using the Bowden–Leben friction apparatus with a sliding speed of 0·01 cm./sec. and a load of 4 kg. Cu, Cd, Fe, and Co sliders and flat surfaces were also introduced into the tests. The surface damage when bonded carbide and metal slide together is comparable with that found with WC itself, and also with that found for other softer metals. Repeated sliding stressing of bonded carbide by bonded carbide leads to a sudden rise in the coeff. of friction and a large increase in wear. This effect did not occur with WC.—F. A. F.

*The Uranium–Cobalt System. W. K. Noyce and A. H. Daane (*U.S. Atomic Energy Commission Publ.*, 1950, (AECD-2826), 6 pp.; also (CT-3242)).—The equilibrium diagram is similar in many respects to those of the U–Fe, U–Ni, and U–Mn systems. A eutectic is formed at 734° C. between the compounds U_6Co and UCo and at 1063° C. between UCo_2 and Co. U_6Co is formed by a peritectic reaction at 822° C. and UCo decomposes peritectically at 805° C.

From the microstructure and X-ray patterns given by alloys with >66 at.-% Co, it appears that there is a region of solid solubility of Co in UCo_2. U_6Co was found to be body-centred tetragonal with $a = 10·34 \pm 0·02$ and $c = 5·20 \pm 0·02$ Å. and contg. 24 U atoms and 4 Co atoms/unit cell. UCo is b.c.c. with $a = 6·343 \pm 0·001$ Å. and contains 8 mol./ unit cell. UCo_2 is f.c.c., having the Cu_2Mg structure and a lattice parameter of $6·9783 \pm 0·004$ Å.—B. W. M.

Bearing Alloys. S. Balicki (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1950, 2, (4), 327–338).—[In Polish]. A review of literature on the compn. and service properties of bearing materials. Three Russian Al-base alloys are said to be in use : Alcusil (with Cu 7–9, and Si 1–2·5%, balance Al), Al + 2·5% Ni, and Al + 6% Fe. A Zn-base alloy with 10% Al and 5% Cu is claimed to be used in U.S.S.R. in machine tools and on the railways to replace Sn bronze. This material should be suitable for steady work at low speeds (0·5–1 m./sec.) and max. pressures of 100 kg./cm.^{2}. 24 ref.—N. H. P.

Thermal Conductivity of Various Materials vs. Temperatures. —— (*Materials and Methods*, 1951, 33, (6), 107).—Materials Engineering File Facts No. 210. Graphs of thermal conductivity against temp. are given for a number of non-ferrous and ferrous alloys. Three parallel scales give the thermal conductivity in cal. or W./cm.^{2}/sec./°C./cm. or B.Th.U./ft.^{2}/ sec./°F./in.—J. W. C.

*Magneto-Resistance of Ferromagnetic Metals and Alloys at Low Temperatures. (Smit). See col. 267.

Alloys Isomorphous with the Iron–Chromium Sigma Phase. K. W. Andrews (*Research*, 1948, 1, (10), 478–479).— A letter. It is pointed out that phases of the sigma type that have been shown to exist in the Co–Cr and Fe–V systems as well as in the Fe–Cr system, appear to be based on equiatomic compn., suggesting that the formation of these phases depends on the onset of an ordered structure which immediately undergoes some kind of distortion.—N. B. V.

*New Supraconducting Borides and Nitrides. J. K. Hulm and B. T. Matthias (*Phys. Rev.*, 1951, [ii], 82, (2), 273–274).— A letter. The nitrides and borides of Nb, Ta, and Mo have been tested for supraconductivity. NbB, MoB, NbN, Mo_2N, and MoN become supraconductors at temp. below 4·4°–14·7° K. The results are particularly interesting theoretically, because Mo itself is not a supraconductor.—P. C. L. P.

*Diffusion in Multiple Phases and Its Metallurgical Application. H. Bückle and J. Descamps (*Rev. Mét.*, 1951, 48, (8), 569–578).—B. and D. applied a previously worked-out technique for the study of diffusion in multiple phases (B., *Metallforschung*, 1946, 1, 175, and *Recherche aéronaut.*, 1949, (12), 45 ; *Met. Abs.*, 1947–48, 15, 402 ; 1949–50, 17, 793) to that of a couple formed by very pure Al and an Al–Be alloy itself consisting of an aggregate of Be dispersed in an Al–Be solid soln. When two metals forming two different phases are made to diffuse reciprocally, the two phases cannot at any given temp. exceed certain concentrations corresponding to the max. solubility at that temp. The phases are in equilibrium with each other at those concentrations, and, this equilibrium always occurring at the plane of contact, the concentration will jump abruptly from one value to the other. This plane of sepn. will be displaced from its origin (e.g. the eutectic compn.) by an amount that varies with the temp. When this displacement is large it can be measured with a microscope, thus providing a simple means of measuring the diffusion. B. and D. show that there are two coeff. of diffusion for the two phases and that the displacement (ξ) is a function of the sq. root of the time and one coeff. of diffusion ; so that $\xi = 2\gamma \sqrt{D_1 t}$, where γ is a const. of proportionality, unknown, and depending on the max. and initial concentrations of the phases and on the other coeff. of diffusion. They derived the math. equations and tested these against the results of experimental detn. of the diffusion of the Al–Be couple after electrolytic polishing and various heat-treatments, measuring small local differences in compn. by micro-hardness detn. The experimentally determined values of the diffusion were somewhat higher than the calculated values by up to 25–30%, but in the circumstances this agreement must be considered satisfactory. 19 ref.—J. H. W.

*The Application of Chemical Thermodynamics to the Study of Metallic Alloy Formation. W. E. Wallace *et al.* (*U.S. Atomic Energy Commission Unclassified Rep.*, 1950, (**NYO-604**), 9 pp.).—A progress report on a study of the heat of formation of Na–K alloys, in which it was required to determine the heat of reaction of the pure metals with water. Since the hydroxides attacked the Monel calorimeter, errors were introduced, and it was necessary to plate with Ni to prevent attack. A summary of the programme of work in progress on Mg–Cd alloys is included.—B. W. M.

*Secondary Electron Emission from Mixed Metals. H. Salow (*Ann. Physik*, 1950, [vi], **5**, (6/8), 417–428).—Mixtures of Ag or Cu with Mg, Al, and Be were prepared by simultaneous evaporation of the two metals in a high vacuum and deposition on plates of Cu–Ni or Cr–Ni alloy. The surfaces were treated with O_2 at temp. from 550° to 750° C. for times of the order of 20 min. to 1 hr., and values of the secondary electron factor $\delta \approx 10$ were obtained for an accelerating potential of 500 V. The alloy surfaces were stable and were not poisoned by dry, CO_2-free gases. δ was found to decrease as the temp. increased, so that the alloys behaved like semi-conducting emitters. An explanation for this is given in terms of the formation of areas of insulating or semi-conducting crystal as a result of the treatment with O_2.—J. W. C.

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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.]

*Slip Bands and Hardening Processes in Aluminium. A. F. Brown (*J. Inst. Metals*, 1951–52, **80**, (3), 115–124).—Slip bands on Al increase in number during plastic deformation and, at the same time, further slip occurs within each band. At higher temp. 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 d 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 mech. equation of state can exist only at very small strains.—AUTHOR.

*[The Structure of] Beryllium Crystals. Gustav Jaeger (*Z. Metallkunde*, 1950, **41**, (8), 243–246).—The literature on the crystallography of Be is reviewed, and typical forms of Be crystals obtained by the electrolysis of fused $BeCl_2$–NaCl mixtures, at temp. below the m.p. of Be, are described and illustrated. They are flat, acicular, or columnar in shape, but all forms have a basic hexagonal structure. Direct measurements give $c/a = 1.569$, a value which is in good agreement with the average (1.568) of the X-ray detn. of other workers. Unlike the acicular and columnar crystals, which are extremely brittle, the plate-like crystals will withstand a limited amount of bending to and fro. 14 ref.—E. N.

Allotropy of Beryllium. A. U. Seybolt, Joseph S. Lukesh, and D. W. White (*J. Appl. Physics*, 1951, **22**, (7), 986).—A letter. Recent work by Sidhu and Henry (*ibid.*, 1950, **21**, 1036; *Met. Abs.*, 1950–51, **18**, 655), which purported to demonstrate the existence of an allotropic change in Be is criticized. S. and H. claimed that certain weak diffraction lines which they attributed to a distinct allotropic form could not be due to impurities, as spectrographic analysis had ruled these out. It is pointed out that BeO would not be revealed spectrographically, and about half of the diffraction lines in question agree with BeO lines in position and relative intensity. For this and other reasons it is concluded that the body of data showing Be to have no allotropic changes should be considered correct.—R. W. C.

*Observation of Magnetic Domains [in Cobalt] by the Kerr Effect. H. J. Williams, F. G. Foster, and E. A. Wood (*Phys. Rev.*, 1951, [ii], **82**, (1), 119–120).—A letter. Magnetic domains in Co have been explored by using magneto-optical effects. The direction of easy magnetization in Co is along the c axis, so domains are magnetized in either a positive or negative sense along this axis. Polarized light incident normal to the (0001) surface undergoes rotation of $\sim\frac{1}{4}°$ on reflection, in a sense depending upon the polarity of the domain. The domain patterns can be viewed if rotation in one sense is opt. compensated for. Photographs are reproduced illustrating the effects of magnetic fields on the domains in Co.—P. C. L. P.

*Crystallization of Copper in Contact with Mercury. Erwin W. Müller, Iwan N. Stranski, and Eugen Szabo de Bucs (*Z. Metallkunde*, 1950, **41**, (8), 226–227).—If ordinary Cu sheet is immersed in Hg for several days at room temp., rhombic dodecahedral crystals, ~ 0.1 mm. dia. and showing only (011) surfaces, are observed on its surface. If the temp. is raised, e.g. to 80° C., the number of crystals decreases; at 150° C. no crystals can be seen, but the surface of the Cu is finely porous and almost slimey, owing to the Hg distilling off. The rate of growth of the crystals is in the ratio 25 : 5 : 1 at temp. of 0°, 20°, and 40° C. The effect is not found with annealed, coarse cryst. material. The phenomena can be explained on the basis of differing rates of adsorption of the Hg on the various crystal planes of the Cu.—E. N.

*The Overgrowth of Copper by Its Oxide. Erich Menzel (*Ann. Physik*, 1949, [vi], **5**, (3/5), 163–180).—The growth and orientation of Cu_2O plates on spherical Cu single crystals previously etched, were investigated at temp. up to $\sim 1000°$ C. After heating at $\sim 1030°$ C. in a high vacuum, all traces of chemical corrosion on the single crystals disappeared and the structure was modified so that close-packed atomic directions formed circular terrace steps on the surface of the sphere. An air pressure of a few mm. Hg at 450° C. provides a thin coating which prevents the change on subsequent heating to high temp., so that the growth of the oxide layers could be related to the initial surface condition produced by etching. Opt. and X-ray methods were used for orientation detn. Only two different orientations of Cu and Cu_2O were observed; in the first the lattices were completely ‖ and in the second an octahedral pole of the Cu coincided with one of Cu_2O and the face diagonals were above each other so that an oxide dodecahedral pole occurred in the region of the Cu cubic pole. 7 different etching reagents were used, and it was found that the ‖ orientation occurred whenever dodecahedral faces of Cu had been exposed initially. In all other cases the antiparallel orientation was found.—J. W. C.

*Relations Between Initial and Final Orientations in Rolling and Annealing of Silicon Ferrite. B. F. Decker and David Harker (*J. Appl. Physics*, 1951, **22**, (7), 900–904).—The orientation relations in question were studied by rolling very coarse-grained strip, the initial orientations of the grains having been found. Grains near the orientation (110) [001] do not change much in orientation after 40% rolling reduction, but become greatly fragmented. Other grains remained almost undistorted (as judged by Laue photographs), but were rotated. A $\frac{1}{4}$-hr. anneal at 800° C. caused only the first group of grains to recrystallize; the recrystn. textures within the confines of the original grains were always close to the original orientation. (No explanation for this correspondence is offered.) The different behaviour of the two groups of grains is explained in terms of the orientations of the slip planes, leading to different deformation mechanisms. These experiments are held to explain why commercial Fe–Si strip, heavily rolled and with texture (001) [110],

recrystallizes to give the magnetically favourable texture (110) [001]. A few grains of the latter orientation are presumed to exist in the rolled strip, and only these grains are distorted enough to recrystallize (which they do without change of orientation).—R. W. C.

*On the Capacity for Slip of Metallic Structures of the Magnesium Group. H. Seifert (Ann. Physik, 1949, [vi], 6, 10–17).—The question of whether the elements of simple slip in metals of the Mg group obey the formal geometrical laws was studied. Complicated, secondary structural elements were shown to be important in the arrangement of lattice planes and lines. The dynamics of the slip movements were related to the anisotropy of the electron d, found by Fourier analysis. On this hypothesis the real glide elements are determined by the position of the nodes of max. electron d; but the lattice-energy explanation also remains possible. Mech. twin formation along the (10$\bar{1}$2) plane was analysed kinetically. 12 ref.—J. W. C.

*On Anodic Polishing of Nickel in Electrolytes Containing Chloride Ions. Sakae Tajima and Takemi Mori (Compt. rend., 1951, 233, (2), 160–162).—T. and M. studied the anodic behaviour of Ni in soln. of HCl and various chlorides at all concentrations between 1 g.-mol. and saturation. The current/potential curve has the form typical of the effect of polishing. The surfaces so obtained are passive. The electrolytes used for polishing and their potentials and c.d. were resp.: HCl, 2·8–5 V., 70–250 amp./dm.²; NH$_4$Cl, 5·5–14, 40–200; AlCl$_3$, 5·5–14, 60–200; MgCl$_2$, 9·5–16·5, 50–230. The effect of the temp. is the inverse of that observed with H$_2$SO$_4$, which polishes Ni equally well; thus at 50°–60° C. polishing is more difficult and requires a higher voltage. On the other hand, at ordinary temp., the efficiency of the anodic soln. is often better than in H$_2$SO$_4$. The pH varied little during the experiment with the AlCl$_3$ and NH$_4$Cl soln., but considerably with the MgCl$_2$ soln. Soln. of NaCl, CdCl$_2$, and ZnCl$_2$ were also investigated. All became alkaline during electrolysis and precipitated Ni(OH)$_2$, but the CdCl$_2$ soln. produced some polishing. Addn. agents, such as glycerine, widened the limits of the conditions of polishing, restrained the formation of hydroxide, and diminished the tendency to tracking on the anode. Ni polished in 3M-HCl is more noble than that polished in H$_2$SO$_4$, which indicates a state of passivity; also the surface does not displace Cu from a soln. of CuSO$_4$.—J. H. W.

*On the Etched Surfaces of Nickel Single Crystals. Shigeto Yamaguchi (J. Appl. Physics, 1951, 22, (7), 983–984).—A letter. The bounding planes of etch-pits produced by dil. Br in ethanol on the polished surfaces of Ni crystals were studied on crystals of different orientations by electron diffraction and microscopy. The general conclusion is that the (110) face, with a "coarse" atomic arrangement, is chem. the most active, while the closest-packed (111) face is the least active.—R. W. C.

*Ferromagnetic Domains in Bicrystals of Nickel. Ursula M. Martius, Kenelm V. Gow, and Bruce Chalmers (Phys. Rev., 1951, [ii], 82, (1), 106–107).—A letter. A bicrystal of 99·92% Ni, several mm. long, has been prepared, and ferromagnetic domain patterns obtained. It is pointed out that a study of the relationships between ferromagnetic domains and grain boundaries may provide additional information about the structure of the latter.—P. C. L. P.

*Calculation of the Density of States Curve for the 3d Electrons in Nickel. G. C. Fletcher and E. P. Wohlfarth (Phil. Mag., 1951, [vii], 42, (324), 106–109).—A letter. An approx. calculation of the shape of the 3d band of the transitional metals is reported, neglecting certain terms which are unlikely to affect the shape of the portion of the band with the highest energy. Tight-binding is assumed. The overall width of the band is 2·70 eV., and two peaks are shown in the number of electron energy states/atom/eV. at approx. 1·0 and 1·9 eV. above the bottom of the band. The deductions made about Ni include the degeneracy temp. for 0·3 vacant states/atom; the low-temp. electronic sp.-heat coeff.; and the degeneracy character at the Brillouin-zone centre.
—P. C. L. P.

*Grain Growth in a Texture [of Nickel–Iron Alloys], Studied by Means of Electron-Emission Microscopy. G. W. Rathenau and G. Baas (Physica, 1951, 17, (2), 117–128).—[In English]. Grain growth and secondary recrystn. in rolled f.c.c. Ni–Fe alloys were studied by electron-opt. observation, the polished surface of the alloys, activated by evaporation of Ba on to it, serving as cathode in an emission electron microscope. Magnification was effected by magnetic focusing. Grain growth in an imperfect cubic texture is a discontinuous process, almost all neighbouring grains quickly invading one grain or group of grains. Grain-boundary movement, at low-energy boundaries, occurs on a small scale. The high-energy boundary parts between a cubic crystal and its (near) twin move in a direction || the common (111) plane. The appearance of grooves at intercryst. grain boundaries is discussed in terms of surface-tension effects. The surface tension at the boundary between a secondary crystal and a cubic crystal which is to be absorbed is about twice that at the surface between two cubic crystal surfaces. The assumption that general grain growth causes crystals with high surface energies to disappear, partially explains why recrystn. textures appear to be similar to, but better-defined than, deformation textures.—J. S. G. T.

*Identification and Crystal Structure of Protactinium Metal. W. H. Zachariasen (U.S. Atomic Energy Commission Publ., 1951, (ANL-4632), 8 pp.).—Pa was prepared by reduction of PaF$_4$ with Ba and was found to have a tetragonal body-centred lattice with $a_0 = 3·925 \pm 0·005$, $c_0 = 3·238 \pm 0·007$ Å. and calculated $d = 15·37 \pm 0·08$. There are 2 atoms/unit cell and each atom has 10 nearest neighbours, of which 8 are at a distance of 3·212 and 2 at 3·238 Å. The radius of Pa for co-ordination No. 12 is 1·63 Å. PaO has the NaCl type of structure with $a_0 = 4·961 \pm 0·003$ Å. and calculated $d = 13·43$. Taking the metallic radii of Th and U as 1·80 and 1·54 Å. resp., the radii for Th, Pa, and U show the rapid decline with atomic number characteristic of filling the d-subshell, thus excluding the presence of 5f-electrons.—B. W. M.

*The Structure of Thorium Dicarbide by X-Ray and Neutron Diffraction. E. B. Hunt and R. E. Rundle (U.S. Atomic Energy Commission Publ., 1950, (AECD-3021), 13 pp.).—The tetragonal structure for ThC$_2$ proposed by von Stackelberg (Z. physikal. Chem., 1930, [B], 9, 437; Met. Abs. (J. Inst. Metals), 1931, 47, 387), is shown to be incorrect on the basis of good X-ray powder photographs and single-crystal data. The lattice found was C-centred monoclinic with $a = 6·53$, $b = 4·24$, and $c = 6·56$ Å., $\beta = 104°$, $d = 9·6$ g./c.c. and contg. 4 atoms/unit cell. The Th positions as determined by X-ray diffraction indicate a space group of C_2/c or Cc and satisfactory C positions have been found in C_2/c by neutron diffraction. The structure contains C$_2$ groups, but since the C–C distance is ~1·5 Å., the triple bonding is eliminated. The arrangement of the Th atoms around the C$_2$ group can be described as similar to that of C$_2$H$_6$, C$_2$H$_4$, and C$_2$H$_2$, indicating a marked covalent character for the Th–C bonds.
—B. W. M.

*An Investigation of the Structural Changes Accompanying Creep in a Tin–Antimony Alloy. W. Betteridge and A. W. Franklin (J. Inst. Metals, 1951–52, 80, (3), 147–150).—A Sn–5% Sb alloy has been used for a microscopical study of phenomena occurring during creep at room temp. In addn. 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 Al 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.—AUTHORS.

The Structure of Hard Metals. Paul Schwarzkopf (Powder Met. Bull., 1950, 5, (5/6), 68–79).—S. discusses the structure of hard metals—substances having similar structural character-

istics to the refractory carbides—and suggests that in establishing a theoretical basis for new development work, the nature of the interatomic forces in the crystals, or bond structure, should be emphasized more than the crystal structure. Pauling's resonating-valence-bond theory is considered to be a suitable starting point for the theoretical treatment of the bond structure of hard metals. 28 ref.—W. A. M. P.

The Structure of Electrodeposited Metals and Alloys. (Raub). See col. 303.

Recent Developments in Metallography. E. C. W. Perryman (*Metal Ind.*, 1951, **79**, (2), 23–26; (3), 51–53; (4), 71–73; (6), 111–113; (7), 131–133).—Read before the Leeds Metallurgical Society. P. outlines the recent advances in specimen prepn. and the apparatus available for metallography, including taper-sectioning, cold mounting, electrolytic polishing, phase-contrast microscopy, the reflecting microscope, the use of polarized light, and micro-hardness testing. He points out the particular appn. of these new techniques.
—J. H. W.

Twenty-Five More Years of Metallography. J. R. Vilella (*J. Metals*, 1951, **3**, (8), 605–619).—Howe Memorial Lecture, 1951. An appraisal of developments in microscopical metallography during the past 25 years : (1) the elimination of disturbed metal from the surface of metallographic specimens—by electropolishing—thereby revealing the true configuration of the structures ; (2) the development of the isothermal transformation diagram, which has helped in the understanding of the structures involved in (1) ; (3) the electron microscope, which has brought closer the definitive resoln. of the structures ; (4) developments in opt. microscopes—conical illumination, phase contrast, coated lenses, interference filters, colour photomicrographs, and photographic materials ; and (5) ultra-violet (3650 Å.) microscopy. 10 ref. and many excellent photomicrographs (of steels).—E. N.

Phase-Contrast Metallography. E. C. W. Perryman and Mary Lack (*Metallurgia*, 1951, **44**, (262), 97–102, 110).—The phase-contrast microscope is described and its met. appn. discussed. Examples are given illustrating the sensitivity of the microscope in revealing small differences in surface level. The possible value of the microscope in helping to solve research problems and to interpret electron micrographs is discussed.—F. M. L.

Examination of Metals Under Polarized Light. I.—Theory and Apparatus. II.—Applications. B. W. Mott and H. R. Haines (*Research*, 1951, **4**, (1), 24–33; (2), 63–73).—[I.—] A review is given of the theory as it developed from 1887, as applied to the major types of crystal structure ; the effects of anisotropic surface films and of etching are also considered. The apparatus required is discussed at length. 70 ref. [II.—] A review of the literature dealing with general appn. is followed by an account of the method as applied to U, Be, and Zr, to isotropic metals covered with an anisotropic film, to reflection from faceted isotropic surfaces and from non-metallic inclusions, and to the measurement of opt. const. for identification purposes. M. and H. conclude that there is a wide field of appn. in the study of the structure of metals and alloys, both for homogeneous phases and for identification and differentiation of inclusions and of various phases. 83 ref.—F. A. F.

The Anodic " Polishing " of Metals. K. Huber (*Chimia*, 1950, **4**, 54–62; *C. Abs.*, 1950, **44**, 6304).—[In German]. Electron micrographs are given of electrolytically polished surfaces of Zn, Cd, Cu, Mg, and Al. Experimental facts are reported on the electrolytic polishing of Zn in NaOH.

†**Anodic Polishing and Its Relation to Anodic Passivation.** (Huber). See col. 306.

Role of Sub-Structure Developed During Plastic Flow on the Mechanical Properties of Metals. E. R. Parker, T. H. Hazlett, and M. W. Nathans (*U.S. Atomic Energy Commission Unclassified Rep.*, 1950, (**COO-10**), 12 pp.).—A discussion of the literature dealing with dislocations, strain-hardening, recovery, and recrystn. and the appn. of the theories in the study of creep. When a stable crystallite size is ultimately attained during creep, the state of the material should then remain unchanged with further strain and the creep rate

should remain const. and independent of further strain. This was confirmed by the results of creep tests (*ibid.*, (**COO-4**) ; *Met. Abs.*, this vol., col. 260.—B. W. M.

Intrinsic Crystalline Structure and the Strength of Metals. W. A. Wood (*Phil. Mag.*, 1951, [vii], **42**, (326), 310–312).—A letter. W. points out that the objections raised by Paterson (*ibid.*, 1950, [vii], **41**, 410 ; *Met. Abs.*, 1949–50, **17**, 931) to the block theory of strength are not conclusive. It is suggested that it is necessary to distinguish between the " disoriented elements " which determine the mech. strength and the X-ray line broadening, and the ultimate intrinsic blocks, whose size variation with temp. has been discussed theoretically. Disoriented elements may contain several intrinsic blocks. The suggestion is supported by experimental evidence that at a given temp., the size of the disoriented elements varies with the strain rate.—P. C. L. P.

*****The Law of Constant Resolved Shear Stress in Crystal Plasticity.** F. R. N. Nabarro (*Phil. Mag.*, 1951, [vii], **42**, (325), 213–214).—N. gives an alternative derivation of the expression for the total force acting on a dislocation in a stressed crystal, previously obtained by Peach and Koehler (*Phys. Rev.*, 1950, [ii], **80**, 436 ; *Met. Abs.*, this vol., col. 31). The method distinguishes between motions of the dislocation which conserve the vol. of the crystal and those which generate interstitial atoms or vacancies. The conservative motions give rise to steady slip in accordance with the experimental result that slip begins when the resolved shear stress reaches a critical value which is independent of the other components of the applied stress.—P. C. L. P.

*****The Degree of Imperfection of Crystals.** N. W. H. Addink (*Rec. Trav. Chim.*, 1951, **70**, (2), 202–208).—[In English]. The apparent value of Avogadro's number N_0 (calculated from lattice-spacing and d measurements) for various crystals is used as an indication of the degree of their imperfection. Imperfections produce a higher apparent value of N_0, and for metals the degree of imperfection is ~0·06% by vol. Results of water-adsorption experiments with KCl are used to estimate the thickness of a mosaic block as ~6×10^{-4} cm.
—J. W. C.

*****Studies of Mechanically Finished Metallic Surfaces by Electron Diffraction and Electron Microscopy. I.—Lapping.** Noboru Takahashi and Toshio Asaeda (*J. Sci. Research Inst.* (*Tokyo*), 1951, **45**, (1240), 23–30).—[In English]. The effect of wet and dry lapping upon the surface of Al bronze, α-brass, phosphor bronze, and high-Cr, high-C steel was studied by electron diffraction and magnetic electron microscopy. The lapping operation comprises an initial (wet lapping) stage consisting of a minute cutting action followed by a later (dry lapping) stage, in which surface flow occurs. In wet lapping the surface remains cryst., and in metals and alloys having a f.c.c. lattice, e.g. Al, Cu, and Ni, fibre structures, [111] and [100], caused by cutting action, and inclined at ~6° to the direction of rubbing are produced. In dry lapping the surface generally becomes amorphous and the fibre structure developed is such that [110] is normal to the surface, indicating a compressive action. Chips become increasingly minute as lapping proceeds, and in the final stage have an average size of 0·02 μ.—J. S. G. T.

*****Anisotropy of Diffusion in Grain Boundaries.** M. R. Achter and R. Smoluchowski (*U.S. Atomic Energy Commission Publ.*, 1951, (**AECU-1219**), 4 pp.).—A. and S. discuss the results of measurement of the rates of diffusion of Ag along the grain boundaries of columnar Cu. Consideration of the diffusion phenomena at the junction of three grains indicates that diffusion along a grain boundary is marked along edge dislocations whenever the angle between the grains is >20° and is small ⊥ these dislocations. On the basis of Mott's grain-boundary model, in which the boundary is made up of islands of fit surrounded by areas of misfit, the diffusion would occur along the areas of misfit and would be always preferred in a direction ∥ the long axis of the areas of fit.—B. W. M.

*****Theory of Grain-Boundary Migration.** R. Smoluchowski (*U.S. Atomic Energy Commission Publ.*, 1951, (**AECU-1224**), 6 pp.).—In the light of recent work on Cu (Achter and S.,

ibid., (AECU-1219); preceding abstract) S. considers that for grain boundaries between grains differing in orientation by only small amounts, the boundary can be considered as built up of dislocations as pictured by Shockley and Read, but, for grains differing by >20° in orientation, Mott's theory of islands of fit and misfit is more applicable. An expression for the grain-boundary mobility is derived on the assumption that it involves a transfer of a group of atoms from the lattice of one grain to the disordered state at the grain boundary and a transfer of a group of atoms from the disordered boundary layer to the cryst. lattice of the other grain. In view of the uncertainty in some of the properties used in the expression, the agreement between the calculated and observed mobilities for Ag, Al, and brass was reasonable and indicates the necessity for further experimental data.—B. W. M.

*On the Causes and Mechanism of the Development of Chemical Inhomogeneity in the Crystallization of Alloys. V. M. Tageev (*Doklady Akad. Nauk S.S.S.R.*, 1949, 67, (3), 491–494).—[In Russian]. In order to study inhomogeneity in alloys, the centre of a 5-kg. steel sample was kept liq. for 5 hr. with the outer surface solidified. Tabulated results show an enrichment of the liq. phase in C, S, and P at the expense of the two-phase region, Si and Mn contents remaining unchanged. Addn. of Sn to the liq. centre resulted in its spreading throughout the two-phase region, which is found to represent a dendritic skeleton penetrated by interconnected canals filled with liquid rich in impurities. Impurities diffuse into the liq. phase where their concentration is lower. Thus, necessary and sufficient conditions for the development of chem. inhomogeneities are : (*a*) different solubilities of the impurities in the liq. and solid phases, and (*b*) prolonged contact of the two-phase and liquid regions. Inhomogeneity is increased by motion of the liquid during solidification.—A. G.

*Method of Investigation of Dendritic Inhomogeneity of Alloys. V. M. Tageev (*Doklady Akad. Nauk S.S.S.R.*, 1949, 67, (4), 675–677).—[In Russian]. Multicomponent systems can be successfully studied by a direct method proposed for the detn. of the compn. of the solid and liq. phases co-existing at a given temp. The method permits the construction of solidus and liquidus lines along the horizontal, and consists of heating a test-piece with a hole until the interdendritic liquid fills the hole, after which the liquid is rapidly cooled and chem. analysed. A cylindrical steel (C 1·08, S 0·022, P 0·017, and Mn 0·16%) specimen with holes at both ends was packed in quartz sand with addn. of 2% boric acid, heated to 1350° ± 10° C. for 6 hr. in graphite, and rapidly air-cooled. Chem. compn. of the liquid which filled the hole was C 1·98, S 0·087, P 0·027, and Mn 0·14%, and that of the solid C 1·03, S 0·040–0·041, P 0·014, and Mn 0·15%.—A. G.

The State of Matter Near the Melting Point. G. M. Bartenev (*Zhur. Eksper. Teoret. Fiziki*, 1950, 20, (3), 218–223; *Physics Abs.*, 1950, 53, 639).—[In Russian]. A derivation is given from general principles of the relation describing the liq.-cryst. state of pure substances near the m.p. Calculation shows that in a Zn single crystal the liq. phase within 3° C. of the m.p. represents ~1%. The calculation was carried through for the anomalous part of the sp. heat and of the vol. coeff. of expansion near the m.p. Agreement with measured results for Zn and Sn was satisfactory. The const. μ entering the formulæ and determined by experiment, represents phys. the critical molecular complex indicating the transition with a reduction of the number of particles in the system from the macro-characteristics of the phase to the micro-characteristics of the molecules. The dimensions of the critical molecular complex are 9000 atoms for Zn and 1500 for Sn. The max. number of independent centres of fusion or crystn. which may exist simultaneously in a system, is equal to the number of the critical molecular complexes forming the system. The ideal polycrystal corresponds to the simultaneous emergence of this max. number of centres in the liq. phase, and μ indicates the lower limit of the polydispersity of the polycryst. material.

*Kinetics of Precipitation in Supercooled Solid Solutions. G. Borelius (*J. Metals*, 1951, 3, (6), 477–484).—An American Institute of Metals Division lecture, dealing with work carried out at the Royal Institute of Technology, Stockholm, on the kinetics of the nucleation phenomena which occur when the cooling of a homogeneous solid soln. results in pptn. with the formation of a mixture of two phases having different concentrations. As a working hypothesis it has been assumed that fluctuations occur in the arrangement of groups of some hundreds of atoms on lattice points of the solid soln. where the alloy is supercooled below the solubility limit and into the range of two-phase equilibrium; on rare occasions these fluctuations give rise to a state which resembles the equilibrium state and thus forms a stable nucleus that is capable of growing by diffusion processes. The probability of nucleation will be connected with the thermodynamic properties of the solid soln., which in turn may be determined approx. from the equilibrium diagram. The theoretical aspects of this are discussed and considered in relation to the resistometric and calorimetric measurements which have been made during pptn. in alloys of Pb–Sn, Au–Pt, Au–Ni, Al–Zn, and Al–Cu, and in α-Fe contg. N and C. The results obtained so far seem to indicate that there are, besides the phase boundaries, two things which have a marked influence on the pptn. and which are fundamental enough to make a part of the phase diagram, namely the spinodal and, at least in some cases, the upper limits of the ranges of existence of various metastable states, as for instance the Guinier–Preston aggregates. It is emphasized, however, that in this first simple survey, no attention has been paid to such important complicating factors as the special influences of grain boundaries and cold working and the decreasing rate of crystal growth with decreasing supercooling; also, the calculations of the free energy from phase boundaries are only a first, rather rough, approximation.—E. N.

Order–Disorder Transformations. B. Kockel (*Ann. Physik*, 1950, [vi], 7, (1/2), 18–32).—Expressions for the energy, sp. heat, &c., of lattices contg. unequal numbers of A and B atoms are derived, assuming that only nearest-neighbour interaction influences the total energy and that the energy of two nearest neighbours is not influenced by the arrangement of the other atoms. [The last assumption is contrary to both the Bragg–Williams and Bethe treatments.] The distribution function $f(n)$ for the number n of AB pairs can be calculated rigidly only for a linear lattice in the general case, but is obtained for plane and three-dimensional lattices by a process of direct enumeration for lattices contg. small numbers of atoms. The results do not agree with the experimental sp.-heat curve. The above assumptions are found, however, to be sufficient for the derivation of the Weiss theory of ferromagnetism; the method is essentially the same as that used by Becker and Döring.—J. W. C.

*On the Problem of Order in Solid Solutions. Iris Runge (*Ann. Physik*, 1950, [vi], 7, (3/4), 129–146).—The plane sq. lattice is considered and the work of van der Waerden (*Z. Physik*, 1942, 118, 473), which suggested that in the partially ordered state atoms in wrong positions form small closed polygons, is developed. There are 11 differently shaped polygons with perimeters $\leqslant 10$ atomic dia. By considering the effect of place exchange of two neighbouring atoms, a balance equation is set up for each of these polygons for the equilibrium state in which the rate of prodn. of a polygon of given type from other types is equal to its rate of disappearance by absorption in other types. From these equations, expressions for the number of polygons of each type in the equilibrium state are obtained as power series in the Boltzmann factor x. Summation leads to a power series for the energy, which differs slightly from van der Waerden's results for powers from x^8 onwards, but for $x \leqslant \frac{1}{12}$ the difference is extremely small. Extension of the theory to the 3-dimensional case is briefly discussed.
 —J. W. C.

*On a Statistical Partial Question of the Problem of Order in 2-Dimensional Binary Solid Solutions. Iris Runge (*Ann. Physik*, 1950, [vi], 7, (5), 240–247).—Cf. *ibid.*, (3/4), 129; preceding abstract. The question of boundaries running right across the ordered crystal is considered. Where two

such boundaries cross they reunite so as to form a pair of hyperbolas. A statistical treatment is given of the frequency of these boundaries and numerical soln. are obtained for two particular cases.—J. W. C.

Possibilities of Studying Experimentally the Shape of an Atom in a Crystalline Lattice. S. T. Konobeevsky (*Izvest. Sekt. Fiziko-Khim. Anal.*, 1949, **19**, 19–25; *C. Abs.*, 1950, **44**, 6208).—[In Russian]. If atoms be considered as spherically symmetrical in the free state (as in the case of true metals), then any changes in the lattice will affect either the radial or the angular distribution. The second is the more likely. The effect of the neighbours on the atom is such that its symmetry decreases in order to conform with the symmetry of the lattice node. This induces anisotropy of the atomic factor in such a way that the denser directions within the atom reduce the brightness of X-rays reflected by planes the normal lines to which are ‖ these directions. A math. analysis is given, supporting the above considerations.

The Atomic Magnetic Moments of Transitional Metals and Their Alloys and the Interatomic Distances in Their Crystal Lattices. (Galperin). See col. 273.

Some Theorems on the Free Energies of Crystal Surfaces. Conyers Herring (*Phys. Rev.*, 1951, [ii], **82**, (1), 87–93).—Knowledge of the relative free energies of surface configurations of crystals is necessary in the study of the sintering of powders, thermionic emission, problems in absorption and catalysis, &c., although these fields involve important dynamical aspects. The simple problem of the equilibrium shape of a small particle is discussed using the Wulff construction. In larger crystals, the relative free energies of neighbouring configurations become important because movement of many atoms over relatively long distances might be necessary to attain the equilibrium shape. The conditions are deduced under which the free energy of a plane surface can be reduced by rearranging the atoms to give hills and valleys of dimensions large atomically but small macroscopically. The circumstances under which curved surfaces can have an energy min. are formulated, and possible crystal shapes are classified into 5 types. The difference in energy between sharp and rounded corners is discussed and applied to the sharp points formed on W in field-emission experiments, and their rounding-off on subsequent heating.
—P. C. L. P.

*Limitation of Electron Density by the Patterson Function. M. J. Buerger (*Proc. Nat. Acad. Sci.*, 1950, **36**, (12), 738–742).—Math. A further elaboration of the principles of vector sets, the appn. of which to Patterson projections allows, in principle, electron-d distributions to be directly obtained. A new function is introduced in this connection, and its properties explained. Its chief use is in " roughing out " certain features of the electron-d distribution.
—R. W. C.

*X-Ray Determination of Particle Size, Lattice Distortion, and Irregular Lattice Displacement Using an Interference Counter-Tube. R. Fricke and K. Heinle (*Z. Elektrochem.*, 1951, **55**, (4), 261–273).—The appn. of apparatus comprising an interference counter-tube and an interference goniometer to the study of particle size, crystal-lattice distortion, and irregular crystal-lattice displacement by X-ray analysis, is described. Attention is drawn to a cause of X-ray line broadening not hitherto discussed, and results obtained with mech. cold-worked and sintered Ni powder indicate a considerable increase of particle size, particularly in sintered samples.—J. S. G. T.

The Diffuse Scattering of X-Rays by a Partially Ordered Crystal.—I. Takeo Matsubara (*X-Sen* (*X-Rays*), 1949, **5**, 102–109; *C. Abs.*, 1950, **44**, 6226).—[In Japanese]. A general theory of the diffraction of X-rays by a crystal having irregularities in its atomic arrangement has been developed. The intensity formula is obtained as a sum of 3 terms, the first representing a normal Laue pattern, the second a diffuse pattern similar to that due to a gas, and the third an anomalous Laue pattern. The last term, which is due to the correlations between the irregularities possessed

by any two lattice points, was calculated for (1) the irregularities due to lattice vibrations, and (2) those due to the order–disorder arrangements of the atoms in an alloy of the type *AB* having a simple cubic structure. Some points concerning the intensity of the anomalous scattering are discussed in connection with the change of the degree of order at the transition point.

Diffuse X-Ray Scattering by Disordered Binary Alloys. William J. Taylor (*Phys. Rev.*, 1951, [ii], **82**, (2), 279–280).—A letter. T. derives the dependence of the diffuse X-ray scattering upon the short-range order parameter. The expressions are applied to Cowley's measurements on Cu_3Au (*J. Appl. Physics*, 1950, **21**, 24; *Met. Abs.*, 1949–50, **17**, 929).—P. C. L. P.

*Geiger-Counter Measurement of the Diffuse Scattering of X-Rays by Crystals and the Effect of Plastic Deformation on It. Gustav Wagner and Albert Kochendörfer (*Ann. Physik*, 1949, [vi], **6**, 129–145).—A sensitive counter was fitted to an X-ray spectrometer and used for measurements on Zn single crystals and Al and Ag polycryst. specimens, using monochromatic Cu K_a radiation. With the Zn crystal, the diffuse X-ray scattering measured in various parts of the $h = k$ plane of the reciprocal lattice showed no significant change as a result of extensions up to 32%, though some irregular variations of ~10% occurred. Similarly, the diffuse scattering from polycryst. Al was unaffected by cold rolling to as much as 99% reduction. For comparison the integrated intensities of the (004) and (002) reflections for polycryst. Ag were measured after compressions up to 65%, and the ratio of these were found to be const. at 0·119 ± 0·004. It is concluded that cold work produces no detectable effects on either the diffuse scattering or on the normal diffraction lines.—J. W. C.

*The Determination of Crystal Structures from X-Ray Data Without a Knowledge of the Phases of the Fourier Coefficients. Dan McLachlan, Jr. (*Proc. Nat. Acad. Sci.*, 1951, **37**, (2), 115–124).—The method, explained here from first principles, yields an electron-d projection of an imaginary crystal created by a special form of addn. of as many mutually displaced Patterson projections as there are atoms in the structure. The resulting electron-d projection contains certain " coincidences," as a result of which a proportion of the peaks coincide with the positions they would occupy in an electron-d projection of the true structure. As an illustration, the method is applied to a compound of known structure.—R. W. C.

A Modification of the Cohen Procedure for Computing Precision Lattice Constants from Powder Data. James B. Hess (*Acta Cryst.*, 1951, **4**, (3), 209–215).—[In English]. It is argued that Cohen's statistical procedure is faulty in respect of the weighting of measurements of line position depending on their position, and consequentially in respect of the least-squares procedure followed. The weighting is discussed anew in the light of recent evidence on the precision of measurement of different lines, and an appropriate least-squares procedure is developed by applying modern statistical theory. The resulting procedure is compared with Cohen's for some sp. samples, and it is shown that the new method often yields slightly different most probable values and that the estimated precision attained is often better by a factor of two or three. Expressions are derived for various film geometries, and the method of derivation for different crystal systems is explained.—R. W. C.

Portrait Shadow-Casting [of Electron-Microscope Replicas]. Delbert E. Philpott (*J. Appl. Physics*, 1951, **22**, (7), 982–983).—A letter. A new technique involving shadow-casting from 2 directions is described, and some electron micrographs of replicas so treated are reproduced. Improved appearance and elimination of excessive contrast are claimed.—R. W. C.

X-Ray Analysis of Crystals. (Bijvoet, Kolkmeyer, and MacGillavry). See col. 332.

The Interpretation of X-Ray Diffraction Photographs. (Henry, Lipson, and Wooster). See col. 333.

Practical Electron Microscopy. (Cosslett). See col. 333.

5 — POWDER METALLURGY

*The Sintering of [Aluminium and] Aluminium Alloys. A. von Zeerleder (*Z. Metallkunde*, 1950, **41**, (8), 228–231).— When oil-free Al powder in the form of thin flakes (<1 μ thick and up to 100 μ dia.) is cold pressed at 20–50 kg./mm.², sintered at 500°–600° C. and then extruded at this temp. under a pressure of 50–100 kg./mm.², the resulting product, called " SAP " (Sintered Aluminium Powder), can be hot or cold worked into sheets, bars, rods, and forgings. It has mech. properties which are equal to those of heat-treated Al alloys such as Avional and Y alloy, but, unlike these, it remains almost unaffected by prolonged heating at 500° C. for periods up to 1000 hr.; e.g. after such treatment SAP has P.S. and T.S. of 30 and 33 kg./mm.², resp., elongation 5%, and hardness 90 kg./mm.². This retention of mech. properties after annealing is due to the oxide films present on the original flake material being retained throughout all the stages of manufacture, and thereby preventing the onset of recrystn. Alloyed SAP can be produced if the powder alloying elements are added in the initial stages of manufacture and if the heating time before hot pressing is sufficiently long to allow complete diffusion to take place; such materials, however, are only slightly susceptible to age-hardening.—E. N.

Brass Powder Parts Aid Fuse Output. ―― (*Iron Age*, 1951, **168**, (7), 108).—The difficulties connected with the high speed prodn. of precision fuse components from brass powders of the 70 : 30 type have largely been overcome, and it is expected that this will result in a widening of the range of appn. of powder metallurgy to ordnance and civilian requirements.
—J. H. W.

*Sintered Porous Graphite-Bronze Bearings. A. Krupkowski and W. Cegielski (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, **1**, (2), 163–179).—[In Polish]. An investigation was made into the effect of compn., sintering, and impregnation conditions on properties such as porosity, hardness, plasticity, compressive strength, and dimensional changes during sintering, using electrolytic Cu powder of 250 mesh, and slightly finer Sn powder. Preheating for 2·5 hr. at 400° C. before sintering at 800° C. increased the porosity and reduced the radial shrinkage, but it also reduced the hardness and compressive strength. The porosity could be increased by adding 3% of talcum, NH_4Cl, or $(NH_4)_2CO_3$, or by sintering in H. The latter method is recommended as giving the best compromise between porosity and dimensional stability, and mech. properties. As reported, oil-porosity varied from 15 to 30%; B.H.N. = 25–45; compressive strength = 30–90 kg./mm.²; radial linear contraction was up to 2·5%. 15 ref.—N. H. P.

Oil-Pocket Metal-Powder Bearings Improve Self-Lubricating Qualities. John Haller (*Materials and Methods*, 1951, **33**, (5), 80–81).—Bearings and bushes are made from sintered powders with interior cavities which serve as lubricant reservoirs. The grease is impregnated into the metal by static pressure and fills the reservoir as well as the pores in the metal itself, thus prolonging the self-lubricating properties five to six times.—J. W. C.

Shell Rotating Bands Successfully Made from Iron Powders. H. R. Clauser (*Materials and Methods*, 1951, **33**, (4), 61–65).— Porous Fe rotating bands for projectiles, made from sintered Fe powder, were originally developed in Germany to save Cu. Tests in America indicate that in addn. to this important advantage Fe bands are superior in performance, particularly for high-velocity projectiles.—J. W. C.

*A New Type of Stainless Powder. George Stern (*Proc. Metal Powder Assoc.*, 1950, 49–56; discussion, 57–58).— S. gives an account of the properties of two new grades of stainless steel powder. Data on chem. analysis, green strength, and the phys. properties of compacts pressed from both powders are given, and data are also included on the phys. properties of one of the powders after infiltration with Cu and Ag. It is claimed that both powders have excellent mouldability, and that their green strength is far superior to any other powders of similar compn.—W. A. M. P.

*On the Pressing of Nickel Powder. H. Lipson (*Powder Met. Bull.*, 1950, **5**, (4), 52–57).—The pressing moduli of Ni powders of different screen sizes have been determined for single- and double-action dies. The moduli do not vary appreciably with particle size, except when there is a marked difference in apparent *d* for a certain size of powder. The difference in die-wall friction between single- and double-action dies is apparent from the different moduli for the two types of pressing.—W. A. M. P.

*A Study of Some [Sintered] Alloys of Titanium. (Kuhn, Kinsey, and Ellis). See col. 281.

*The Sintering of [Tungsten Carbide–Cobalt] Hard Metals. Walther Dawihl and Karl Schröter (*Z. Metallkunde*, 1950, **41**, (8), 231–233).—A report of a microscopic investigation of the structure of WC–Co mixtures sintered at various temp. The results show that WC is appreciably soluble in solid Co at temp. as low as 1150° C., i.e. when there can be no possibility of fusion taking place. The solubility increases with increasing temp., and fully sintered structures can be obtained by heating a 6% WC–Co mixture at 1250° C. for 30 hr. Thus, there is no fundamental difference between materials produced at temp. below or above the m.p. of the eutectic (1400° C.), both products being the result of soln. of WC in Co, in the solid and liq. states, resp. The system WC–Co is quasi-binary, with a eutectic at 22·5% WC; Co can retain 10 and 7% WC in solid soln. at 1400° and 1300° C., resp.; <1% Co is soluble in WC at temp. between 1300° and 1500° C.—E. N.

*Frictional Properties of Tungsten Carbide and of Bonded Carbides. (Shooter). See col. 281.

Hard Metal in Industry. G. J. Trapp (*Trans. Manchester Assoc. Eng.*, 1947–48, 187–206; discussion, 207–214; and (abridged) *Mech. World*, 1948, **123**, (3192), 332–335).—After defining hard metals and giving a short note on their development, T. describes their prodn. and appn.—S. R. W.

The Structure of Hard Metals. (Schwarzkopf). See col. 286.

The Use of Metal Powder Parts as Friction Material. L. P. Kane (*Proc. Metal Powder Assoc.*, 1950, 59–75).—K. gives an account of the manufacture and properties of metal-powder friction materials, with some examples of their use in clutch and braking systems.—W. A. M. P.

*The Effect of Porosity of Powdered Metal Products on Their Mechanical Properties. Critical [Strength] Properties of Porous Metal-Ceramic Materials. M. Yu. Balshin (*Doklady Akad. Nauk S.S.S.R.*, 1949, **67**, (5), 831–834).—[In Russian]. Strength of porous metal-ceramic materials, related to the weakest contact cross-section A_c forming a fraction α of the total cross-section is unaffected by porosity, whilst deformation is proportional to the change in stress referred to A_c. In his books, B. has previously shown that $\dfrac{d\sigma}{\sigma} = m\dfrac{d\theta}{\theta}$ and $\dfrac{\sigma}{\sigma_{max.}} = \dfrac{\alpha}{\alpha_{max.}} = \theta^m$, where σ is the nominal compacting pressure, θ relative *d*, $\sigma_{max.} = \sigma$ for θ = 1 when also α = $\alpha_{max.}$ = 1, and *m* = const. These relations also hold for sintered compacts, though the values of *m* are different, as shown by logarithmic plots of B.H.N., Young's modulus, % elongation, U.T.S., compacting pressure, and Y.P. against θ for unsintered and sintered Cu compacts. Test data show that in the formula $\sigma_{bn} = \sigma_{bk}\theta^m$, where σ_{bn} = U.T.S. of sintered compact of relative *d* θ, and σ_{bk} = U.T.S. of the annealed cast metal, *m* varies from 3 to 6, but tends to the limit *m* = 3 for compacts produced by best techniques, as demonstrated by the results of numerous investigators.—A. G.

*Pressing and Sintering of Mixtures of Metallic and Non-Metallic Powders. W. Rutkowski and H. Rutkowska (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, **1**, (2), 111–125).—[In Polish]. Experiments were carried out on the pressing and sintering of electrolytically prepared powders of Fe, Cu, Fe–Cu (1 : 1), Fe–Mn (1 : 1), Cu–Mn (1 : 1),

and Fe–Cu–Mn (1 : 1 : 1) using pressures of 1–15 tons/cm.², followed by sintering for 3 hr. in H at 800° (Cu), 1200° (Fe), and 1100° C. (all others). Experiments were also made on mixtures of either carbonyl or electrolytic Fe with steatite or Bakelite in proportions from 99 : 1 to 50 : 50. Curves are given relating d and shrinkage (or expansion) to applied pressure. The relation between d (s) and compacting pressure (p) is expressed by $\lg s = a \lg p + \lg b$, a and b being experimentally determined const. Other curves show d and actual pressure distribution in tall compacts ($h/D = 2 \cdot 5/1$) pressed in a single-action hydraulic press. 14 ref.—N. H. P.

Powder Metallurgy Can Help Conserve Critical Metals. A. J. Langhammer (*Materials and Methods*, 1951, 33, (2), 134, 136, 138, 140).—The uses and advantages of powder-met. methods of prodn. are outlined.—J. W. C.

Flexible Design and Lower Costs Achieved by Powder Metallurgy. Charles L. Bigelow (*Western Metals*, 1951, 9, (8), 39–41).—Some appn. of powder metallurgy are indicated.—T. G.

Looking Forward in Powder Metallurgy. H. W. Greenwood (*Machinery* (Lond.), 1951, 79, (2021), 236–237).—Cu-impregnated cast Fe and Fe compacts, Cu-coated Fe powders, and Ti-bonded ceramic materials are among the developments discussed.—J. C. C.

Informal Open Discussion of Powder Metallurgy. —— (*Proc. Metal Powder Assoc.*, 1950, 5–24).—A verbatim account of an informal discussion at the Annual Meeting of the Metal Powder Association at Detroit, Mich., in 1950. —W. A. M. P.

*Preparation of Metal Carbides.—I. G. F. Hüttig, V. Fattinger, and K. Kohla (*Powder Met. Bull.*, 1950, 5, (3), 30–37).—An account is given of some experimental work on the effect of the surrounding gas atmosphere on the prepn. of WC and Mo_2C from W metal, WO_3, MoO_3, and C black, and experimental data on the reaction rates for the following atmospheres are tabulated : (1) H with propane ; (2) H vacuum ; (3) H with HCl and propane ; (4) H with HBr and propane ; (5) H with HI and propane. The effects of the H halide content of the atmosphere on the prepn. of WC, Mo_2C, and TiC are discussed, and some experiments on the effect of H halide addn. to the atmosphere on the sintering of metal powders are briefly described. 9 ref.—W. A. M. P.

*Oxidation of Cemented Carbides. R. Kieffer and K. Kölbl (*Powder Met. Bull.*, 1950, 5, (3), 38–39).—Photographs obtained during a systematic study of the oxidation mechanism of cemented carbides are reproduced.—W. A. M. P.

*Method of Mechanically Pulverizing Metal by Cutter Mill. Toshihiko Okamura, Kôji Inagaki, and Yoshimichi Masuda (*Sci. Rep. Research Inst. Tôhoku Univ.*, 1950, [A], 2, (5), 809–820).—[In English]. The pulverizing of Pb, Cd, Sn, Ag, Cu, Fe, cast steel, and imitation Au (Cu alloy) in a cutter mill, pulverizing the material mainly by the shearing forces of the cutter, was studied. The relation between work done in the mill and pulverizing time, the effects of mech. property and size of initial material, changes of particle-size distribution and of shape of powder produced, are discussed. The pulverizing action of the mill comprises two processes, viz. (1) initial pulverization down to 75 mesh, followed by (2) powdering due to friction between the particles, down to, e.g., 300 mesh.—J. S. G. T.

On the Principles of Electrodeposition of Metal Powders. (Wranglén). See col. 305.

*Thermochemical Coated Metal Powders. J. E. Drapeau (*Proc. Metal Powder Assoc.*, 1950, 28–35).—The thermochem. principles of coating one metal powder with one or more other metals are briefly discussed and future indust. appn. are considered. The Cu coating of cast-Fe powder by mixing finely divided Cu_2O with the cast Fe in a neutral atmosphere at an elevated temp. is described, and an account is given of the thermochem. coating of three cast-Fe powders from different sources with Cu, Cu and Ni, and Ni, using Cu_2O and NiO. Data on the phys. properties of the coated material are included.—W A. M. P.

The Rapid Determination of Powder Density. G. H. B. Lovell (*Trans. Brit. Ceramic Soc.*, 1951, 50, (1), 1–9).—Describes apparatus for the rapid and accurate detn. of powder d. A specified vol. of a liquid is run into an evacuated flask of known vol. and contg. a specified weight of sample. By measuring the total vol. of liquid and solid, the vol. of the sample is obtained.—D. K. W.

6 — CORROSION AND RELATED PHENOMENA

*The Fundamentals of Corrosion of Metals Under Stress. I.—Aluminium. P. F. Thompson (*Commonwealth Australia, Dept. Supply Development, Aeronaut. Research Lab. Rep.*, 1950, (SM-143), 50 pp.; and (also) *Commonwealth Australia, Council Sci. Indust. Research, Aeronaut. Research Rep.*, 1950, (ACA-49), 22 pp.).—Potential/time measurements were made on pairs of sheet test-pieces of commercial Al, one unstressed and the other under increasing tension. They were immersed in tap- and sea-water, without and with aeration. Inflections in the curves denoted rupture in the surface film ; these increase in magnitude until fracture occurs. Re-formation of the film is rapid, and the curve for the potential is characteristic of the liquid of immersion, as the potential reverts to its original value. Superimposed shock loading increases the amplitude of the inflections. On work-hardened Al, increasing stresses have very little effect until fracture is approached. Corrosion may involve grain-boundary material being dissolved, followed by differential aeration corrosion in stress-opened cavities.—S. M.

*Experiments on the Reaction of Aluminium–Magnesium Alloys with Steam. A. J. Swain (*J. Inst. Metals*, 1951–52, 80, (3), 125–130).—The reaction of Al–Mg alloys contg. up to 25% Mg with pure steam has been investigated over a range of temp. from 450° to 700° C. A max. reactivity, dependent on the compn., was found at temp. between 550° and 625° C. The significance of these results in relation to metal/mould reaction in the Al–10% Mg alloy is discussed briefly.—AUTHOR.

*Oxidation of a Thin Single-Crystal Film of Copper. Shunji Shirai (*J. Phys. Soc. Japan*, 1947, 2, 81–83 ; *C. Abs.*, 1949, 43, 6880).—[In English]. The structures of pure and oxidized Cu films were studied by the transmission method of cathode-ray diffraction. The Cu film, approx. 40 μ thick, was prepared on the cleavage surface of NaCl, maintained at 300° C., by the evaporation method. The films were then oxidized by heating for 10 min. in an elect. furnace in air at temp. between 220° and 280° C. (Cu_2O), exposing to air while floating on water for several hr. (Cu_2O), and by heating at 280° C. or higher in an elect. furnace for 10 min. (CuO). Numerous orientations of the Cu_2O and CuO crystals relative to Cu were found. The Cu_2O crystals gave 3–4 orientations, the latter number occurring when oxidized over water. Impurities or changes in the temp. of the water produced no changes. The results on the CuO crystals were inconsistent, the number and kinds of arrangements changing in different experiments. Variations in the manner of oxidation produced no significant changes. Films oxidized while attached to the substrate gave less perfect orientations.

*The Overgrowth of Copper by Its Oxide. (Menzel). See col. 284.

*Corrosion of Condenser Tubes. I.—Electrochemical Characteristics of Copper, Zinc, and Some Brasses. II.—The Effects of Cuprous Ions, Sodium Chloride, and Air-Bubble Impacts on the Electrode Potential. III.—Effects of Biological Factors and Crystal Structure [of Oxides], with Concluding Remarks. P. F. Thompson (*Commonwealth Australia, Council Sci. Indust. Research, Div. Aeronautics, Rep.*, 1948, (SM-113), 100 pp.).—The degree of aeration at metal surfaces has profound effects on their electrochem., and hence corrosion, behaviour. [I.—] Micro-electrode potential/time

and potential/temp. measurements were made with Cu, Admiralty brass, Muntz metal, and Al brass, in fresh and salt water, with aeration and de-aeration, together with regulated abrasion. NaCl depresses the potentials of Zn, Cu, and their alloys and also affects their potential under aeration. Corrosion examples are illustrated by stereographs. [II.—] Further micro-electrode potential measurements were made to determine : (1) the effect of Cu˙ ions on the corrosion of brass by Cu deposition (dezincification) ; (2) the effect of NaCl soln. on the oxidizing potentials of dissolved O ; and (3) the effect of air-bubble impacts on the potentials of brasses. Where aeration is limited, Cu˙ ions in sea-water have an oxidizing potential sufficient to corrode brass. The electrode potential of Cu|sea-water is not affected by 0–315 g./l. NaCl and 6·2–1 c.c./l. O. Brasses become cathodic when air bubbles impinge on them, so that corrosion should occur away from these points. [III.—] Sulphides or ammonia from decomposing organic matter have a negligible effect on electrode potentials, although bacterial sludge may screen O from access to metal surfaces sufficiently to affect differential aeration potentials. Corrosion may attack oxides preferentially at grain boundaries. The nature of the oxides (*not* the crystal structure) and season-cracking are discussed at length. Differential aeration is also an important factor at drawing defects in the tubes. High velocity and turbulence may not be important factors in corrosion.—S. M.

Nylon—A Factor in the Stress-Corrosion of Aluminium Bronze. E. J. Silk (*J. Metals*, 1951, 3, (8), 600).—A description of stress-corrosion-cracking which occurred in a clamp ring of Al bronze (Cu 94·5, Al 5·2, Pb 0·001, Fe 0·04, and As 0·20%) in contact with a nylon part in service in city tap-water. The assembly consisted of the nylon part held to a stainless-steel part by means of Si bronze screws through the clamp ring ; the last-mentioned failed in a brittle manner at two points diametrically opposite and at right angles to the screw holes. As the Al bronze was found to be met. free of defects, it was concluded that the nylon part was the contributing factor to the failure, a possible explanation being that the nylon may have contained some residual diamines which did not polymerize during manufacture, thereby releasing NH_3 when in contact with H_2O.
—E. N.

*Intercrystalline Corrosion of Copper–Silicon Alloys by Atmospheric Oxygen. Otto Heusler (*Z. Metallkunde*, 1950, 41, (8), 261–264).—A study of Cu–Si (15–40%) alloys shows that on exposure to the atmosphere they undergo intercryst. attack due to oxidation. The effect, which is appreciable at room temp. (20° C.)—the alloys eventually crumble to a powder—increases with increasing temp., and at 300° C. spontaneous combustion takes place owing to the exothermic nature of the reaction. Small addn. (0·1–0·5%) of : (1) As, Bi, Ge, Pb, Sb, Sn, and Tl accelerate the reaction, as does the presence of moisture, while (2) Be and Zn suppress it, and (3) Ag, Al, Ca, Cd, Co, Cr, Fe, Hg, Mg, Mn, Mo, Ni, P, Se, Ti, V, and W have no appreciable effect. H. can offer no explanation of the phenomenon resulting from the small addn. 5 ref.
—E. N.

*The Corrosion of Silver by Potassium Cyanide Solutions and Oxygen. Vibeke Lund (*Acta Chem. Scand.*, 1951, 5, (4), 555–567).—[In English]. The variation with time of the O pressure over, and the concentration of O^{--} ions in, a KCN soln. in a closed, shaken flask also contg. O and an immersed Ag sheet, was followed quant. The Ag is at first autoxidized, and the resultant Ag^{++} ions are attacked by the CN^- ions. H_2O_2 is generated at the same time. At a later stage, the H_2O_2 itself contributes to the attack on the Ag, and some of it is catalytically decomposed. The rate of attack depends on the rate of shaking and on the surface condition of the Ag. The later stages of the attack were eliminated if the Ag was first given a high chem. polish by immersion in a soln. of KCN + excess H_2O_2.
—R. W. C.

The Electrochemical Character of Corrosion and Other Reactions in a Metal|Solution System : The Potential/pH

Diagram of the System Silver|Solution According to M. Pourbaix. (Nagel). See col. 305.

*Chemical Reactions of Intermetallic Phases. I.—Corrosion of the Intermetallic Compounds Ag_5Al_3 and Mg_2Pb. E. E. Cherkashin, F. A. Derkach, and S. M. Przhevolotskaya (*Zhur. Obshch. Khim.*, 1949, 19, (5), 798–804 ; *C. Abs.*, 1949, 43, 8341).—[In Russian]. (1) Ag_5Al_3 (Ag 87, Al 13%), melted under a $CaCl_2$ + NaCl flux, superheated to 200°–300° C. above the liquidus temp. and slowly cooled, has a porous structure and corrodes in the atmosphere very rapidly to form a dark powder. The same alloy, produced under the same flux, but superheated by only 20°–30° C. and quenched, has a considerably denser structure and corrodes less rapidly. In dry air, neither of the alloys corrodes. Alloys prepared under the same conditions as the first sample, but under a $BaCl_2$ + NaCl flux or under illuminating gas, do not corrode even in the atmosphere. Corrosion consists in a reaction with H_2O : $2Ag_5Al_3 + 18H_2O \rightarrow 10Ag + 6Al(OH)_3 + 9H_2$, and is strongly accelerated by the hygroscopic flux occluded during the crystn. (2) Among Mg–Pb alloys, in the range 2–70% Mg, Mg_2Pb corrodes most rapidly in the atmosphere. The corrosion/time curve for this alloy shows an inflection at a relative increase of weight close to 0·266, corresponding to the completion of the reaction : $Mg_2Pb + 4H_2O \rightarrow Pb + 2Mg(OH)_2 + 2H_2$, which evidently represents the first stage of the process. Beyond that point, the finely divided Pb is no longer protected by the H_2 evolved, and is slowly oxidized to the final product $PbO.2MgO.3H_2O$. The relative gain of weight of the alloy with 15% Mg (excess of Pb) at the inflection point was only 0·16, instead of the theoretical 0·199, possibly owing to inhibition by the eutectic α-phase. On the other hand, the alloy with 21·7% Mg showed, at the inflection point, a relative gain of weight of 0·280, higher than the theoretical value 0·253. This may be linked with the longer time (20 days) necessary for this alloy to reach completion of the first stage of the corrosion process, as contrasted with 12 days for Mg_2Pb (18·8% Mg) and the 15% Mg alloy. (3) The reactivity with H_2O increases in the order Mg_2Si, Mg_2Ge, Mg_2Sn, Mg_2Pb, parallel with the increasing proportion of metallic bond and decreasing proportion of ionic bond. The 4th-group elements in these intermetallic compounds play the role of electron-transferring agents from Mg to H^+ ions. (4) The rate of corrosion of intermetallic phases can be used for purposes of physicochem. analysis. Thus, the relative increase of wt. at a given moment, e.g. after 120 hr., has a max. at the compn. Mg_2Pb, and thus reveals the existence of a compound.

*An Investigation of the Longevity of Galvanized Roofing Sheets and Wires. D. J. Swaine (*Chem. and Ind.*, 1951, (37), 764–767).—A detailed investigation is reported of the compn., structures, and corrosion products of galvanized roofing sheets exposed to rural and indust. atmospheres in Australia for periods of 50–60 years, and of galvanized fencing wire exposed to rural conditions for 90–100 years, the data being compared with analyses, &c., of modern galvanized products. The superior corrosion-resistance of the older materials is considered to be mainly due to their heavier coatings.
—J. R.

*Oxidation of Cemented Carbides. (Kieffer and Kölbl). See col. 296.

Stainless Steel [and Non-Ferrous Alloy] Castings, Valves, and Fittings Beat Corrosion in Paper Industry. N. S. Mott (*Paper Trade J.*, 1951, 132, (4), 31–32, 34, 36).—Conditions obtaining in various paper-making processes are briefly described with ref. to corrosion of metal parts. Monel metal gives very good service in soda mills where hot and strong caustic soln. are used, and Cr–Ni steel (20 : 10 or 29 : 9) is recommended for certain parts of sulphite digesters. Bronze or Cu alloy should not be employed in sulphite cooker systems, owing to the possibility of electrolytic reaction with the steel parts at high temp. Pipe-lines, valves, and fittings for conveying Cl for pulp bleaching should be of Hastelloy C. Mech. properties, chem. compn. and A.S.T.M. designations of corrosion-resistant cast alloys used are tabulated. 12 ref.
—J. R.

Corrosion Problems in the Railroad Industry. Ray McBrian and L. C. Atchison (*Corrosion*, 1950, **6**, (9), 286–289; discussion, 289).—Some corrosion problems in the railroad industry are briefly reviewed and discussed.—W. A. M. P.

Corrosion in Industry. I. S. Hunt (*Rep. 6th Sci. Congr.*, **1947**, in *Trans. Roy. Soc. New Zealand*, 1949, **77**, 131–138; *C. Abs.*, 1950, **44**, 6375).—Metal corrosion is reviewed and examples are given of a number of types, illustrated with photomicrographs. Pitting of a boiler feed-water tube was attributed to dissolved O, and a different type of pitting on a boiler tube was found to be due to Cu deposited on the tube, after soln. elsewhere in the system. Only one type of metal should be used in such a system. Pitting of a cast-Fe pot used for distilling HCl was attributed to a poor-quality casting. The protective oxide film on stainless steel is destroyed by hot gases contg. S oxides. Corrosion due to dezincification of a brass condenser tube in salt-water is illustrated; a material not subject to dezincification should have been used, such as Al bronze. Bimetallic couples are a frequent cause of corrosion. An example is given of an unusual type of Cu-clad heat-exchanger tube, which showed characteristic bimetallic corrosion; a heavy Sn or Zn coating would have been preferable. Failure of the protective coating on a steam-heated Sn-coated Cu vessel was attributed to overheating; this converted the protective layer of Sn–Cu eutectic into a higher Cu alloy, which was cathodic both to the eutectic and to the Cu sheet, causing pitting to continue until the sheet was perforated.

*The Stability of Analytical Weights, Particularly in Chemical Laboratories. P. H. Bigg and F. H. Burch (*Brit. J. Appl. Physics*, 1951, **2**, (5), 126–131).—A variety of 100-g. weights were exposed to corrosive laboratory conditions, mostly for 6 months, and their weight changes (mostly increases) measured. The most const. were the weights made of: (1) 25 : 20 austenitic stainless steel highly polished, (2) Nichrome (Ni 80, Cr 20%) highly polished, and (3) plated brass weights where the plating (Pt, Rh, or Cr) was deposited on a thick undercoat of Ag or Ni. Plated weights without undercoats and unpolished steel weights deteriorated badly. The polished steel weights were the best (change over 3 months <0·1 mg.). A warning against the use of Pb as a weight-adjusting material is given. Some other points relevant to the stability of weights are discussed.—R. W. C.

The Corrosion Situation : Past, Present, and Future. U. R. Evans (*Chem. and Ind.*, 1951, (34), 706–711; also *Engineering*, 1951, **171**, (4446), 443–444).—A lecture to the British Iron and Steel Research Association. Aspects of non-ferrous interest include discussion of protective Al_2O_3 films on alloys, Zn and Al spraying, and the development of Al brass marine condenser tubes. Treatment of ancient bronze and other castings attacked by " bronze disease " (intensive local corrosion following destruction of the protective patina) is described in some detail.—J. R.

Fundamentals of Galvanic Corrosion. A. D. Lauderbaugh (*Proc. Amer. Gas Assoc.*, **1950**, 900–912).—An elementary description of the generation of galvanic currents, corrosion caused by stray currents, and the appn. of cathodic protection.
—G. T. C.

*Evans [-Type Local] Cells : A Reply to Masing's Observations on the Theory of Corrosion by Local Cells. K. Wickert and H. Wiehr (*Werkstoffe u. Korrosion*, 1950, **1**, (8), 299–302).— Cf. Masing, *Arch. Metallkunde*, 1949, **3**, 343; *Met. Abs.*, 1950–51, **18**, 40. W. and W. have carried out further experiments with the N_2–O_2 cell previously described (*ibid.*, 1947, **1**, 251; *Met. Abs.*, 1947–48, **15**, 511)—a form of Evans differential-aeration cell—modified so that its external and internal resistances are ~33 and 65 Ω, resp. Measurements of the loss in weight at the anode and at the cathode show that where their surfaces have been roughened with a file the metal loss at the cathode is ~4 times as great as that at the anode; only when the surfaces have been polished with 0000 emery—and thereby passivated—does the loss at the anode (which is then somewhat greater than when filed) exceed that at the cathode (which becomes zero). If the internal resistance of the cell is increased to 940 Ω and roughly filed electrodes are used, the losses at the cathode and anode become, resp., greater and considerably less than those obtained when the internal resistance is low. It is concluded that the results show that any formulations of a general nature about Evans-type local cells can only result in contradictions if the surface activity of the anodic and cathodic surfaces is not taken into account. Since the corrosion is regarded as a phenomenon occurring within atomic domains, the theoretical nature of the process can be deduced correctly only if the properties of those areas are taken into consideration. 7 ref.—E. N.

*The Significance of the Activity of Metal Surfaces in Electrochemical Reactions. Kurt Wickert (*Metalloberfläche*, 1950, [A], **4**, (12), 181–186).—After reviewing the nature of de la Rive-type and of Evans-type local cells, experiments with a N_2–O_2 cell—a modified form of Evans differential-aeration cell—are described (cf. preceding abstract). The results are considered to contradict the observations of U. R. Evans that : provided the resistances of the cell and the external circuit are low, and that all parts of the aerated electrode are well supplied with O, the corrosion at the aerated electrode (the cathode) is (often) less than that at the unaerated electrode (the anode). W. found that the metal loss at the cathode depended principally on the surface activity of the electrodes—a factor ignored by Evans—and only to a slight extent on the internal resistance of the cell; only when passive electrodes were used did the loss at the anode exceed that at the cathode; in all other cases the loss at the cathode was approx. four times as great as that at the anode. It was also found that initially passive Fe cathodes became corroded after long periods of aeration, thus showing that when O is present in NaCl soln. it has an activating and not a passivating effect on Fe electrodes. Although all the results reported were obtained with mild steel (0·05% C) electrodes, similar results have been found when using electrodes of Al, Ca, Mg, and Sn. 6 ref.—E. N.

Corrosion Testing. —— (*Mech. Eng.*, 1951, **73**, (3), 224–225).—A new sea-water corrosion-testing station under the Kure Beach Project was formally opened at Wrightsville Beach in 1950. Testing facilities have been extended to observe the behaviour of metallic and organic protective coatings. Some of the advantages of this new site are quoted as : (1) a continuous supply of full-strength sea-water, uncontaminated by indust. wastes, oil films, or other polluting agents ; (2) a relatively wide range of sea-water temp. 45°–85° F. (7°–29° C.) ; and (3) a long season of growth of a large number of marine organisms. Specimens are fastened to Monel racks by means of Monel machine screws, and galvanic action is eliminated by the use of Bakelite insulators.
—H. Pl.

Management's Attitude Toward a Corrosion Control Programme. R. B. Wilson (*Proc. Amer. Gas Assoc.*, 1950, 328–331).—The corrosion control programme of a large gas and elect. undertaking is described in so far as it is applied to buried services. It is claimed that experience over several years shows such a programme to be fully justified.—G. T. C.

7 — PROTECTION

(Other than by Electrodeposition.)

Surface Protection of Light Metals by Organic Film Formers [Lacquers]. G. Hoffmann (*Metalloberfläche*, 1951, [A], 5, (4), 61–62).—A study is reported of the surface protection afforded to light-metal alloys by various cellulose and synthetic lacquer films. The coatings were applied to the properly cleaned metal surfaces, air dried or stored and then subjected to continuous and to alternating immersion in aq. soln. of : (1) NaOH 0·1%, $pH = 10$, and (2) NaCl 0·2 + CH_3COOH 0·1%, $pH = 4$. The results show that : (i) synthetic resin lacquers have the best adhesion and imperviousness ; (ii) in alkaline media, stoved phenol-aldehyde lacquers are superior to air-drying cellulose lacquers ; (iii) it is more difficult to obtain satisfactory protection against alkaline media than against acid media ; and (iv) the best protection is afforded by lacquers based on polyvinyl chloride and alkyl styrol.—E. N.

Tube Galvanizing Plant. —— (*Metal Ind.*, 1951, 79, (4), 65).—A brief description of the magnetic-roll equipment at the works of Stewarts and Lloyds, Ltd., Northampton.
—J. H. W.

*Prevention of Corrosion of Galvanized Iron by Glassy Metaphosphates. F. E. Worsnop and A. Kingsbury (*Chem. Eng. and Min. Rev.*, 1950, 42, (5), 173–176).—Zn and Ca metaphosphates were found suitable for protection of galvanized Fe against attack by distilled water. The weight of the Zn coating was ~2 oz./ft.², and it contained 0·3% Sn and 0·8% Pb. The phosphate being tested was either dissolved directly in the corrosive medium (distilled water) or added as a relatively strong soln. in distilled water. A method for preparing the metaphosphate glasses is given.
—T. A. H.

Phosphate Coatings. —— (*Australasian Plating and Finishing*, 1951, 1, (3), 9–13).—A brief review.—T. A. H.

Cathodic Protection of an Active Ship in Sea-Water.—II. K. N. Barnard (*Corrosion*, 1951, 7, (4), 114–117).—Cf. B. and Christie, ibid., 1950, 6, 232 ; *Met. Abs.*, this vol., col. 201. H.M.C.S. *New Liskeard* was examined 11 months after Mg anodes had been fitted to each bilge keel. Very little corrosion was found on the underwater hull and it was also relatively free from fouling growth. 20–35 amp. were required for this protection, and 435 lb. of Mg were consumed in the 11-month period. The measurement of "hull potential" is shown to be an effective criterion as to how much current should be applied at a given time.
—W. A. M. P.

Use of Magnesium Ribbon for Cathodic Protection [of Gas Pipes]. T. W. O'Brien (*Proc. Amer. Gas Assoc.*, 1950, 819–822).—Experience is described of the use of Mg ribbon ploughed in alongside a bare steel gas pipe which passes through various types of soil. Reduced leak frequency is taken to indicate the effectiveness of the cathodic protection.
—G. T. C.

Cathodic Protection on the 840-Mile Texas Gas Transmission Corporation 26-Inch Pipe Line from Texas to Ohio. Robert J. Kuhn (*Proc. Amer. Gas Assoc.*, 1950, 831–840).—The appn. of cathodic protection, using rectifiers, is described in general

terms, and some data are presented on the reduction of leak frequency as a result of this protection.—G. T. C.

On Safety Standards Affecting the Electrical Protection of Underground Systems from Corrosion. V. S. Kalman (*Elektrichestvo*, 1948, (7), 55–58 ; *Elect. Eng. Abs.*, 1950, 53, 333).—[In Russian]. Experiments carried out on a typical urban underground system of pipes, drains, and cables are used to show that the existing Soviet standard rules concerning elect. protection from corrosion must be modified to reduce the danger of accidental corrosion of neighbouring pipes and drains. The underground system used for the experiment is described, and graphs are given to support the conclusions.

*The Role of Current Distribution in Cathodic Protection. Homer D. Holler (*J. Research Nat. Bur. Stand.*, 1951, 47, (1), 1–6).—A method is developed for determining the c.d. variation over an electrode surface in a uniform medium. The procedure is to determine the relation between c.d. and polarized potential, using a ref. electrode and a bridge method, and then to calculate the c.d. at various points from the potentials measured to the same points. The resistances of the paths from anode to cathode were calculated for a laboratory test cell arranged to give non-uniform distribution over the cathode, and the effect of back e.m.f. demonstrated. Suggestions are made for a field procedure for determining current distribution in large cells. When the c.d. polarized-potential relation is not known in this case, it may be determined using a guard-ring principle to give a uniform current over a measured area. The c.d. distribution may be important in cathodic protection ; if the medium is non-uniform, however, the potential criterion for protection must be used alone.—J. W. C.

Metal Spraying in Protective Atmospheres. H. Reininger (*Metalloberfläche*, 1951, [A], 5, (3), 42–43).—R. considers that spraying in atmospheres of inert or reducing gases, as a means of preventing oxidation, can only be effective if the whole operation is carried out in a special chamber from which air is excluded. The method is, however, uneconomical and never completely satisfactory owing to the unavoidable oxidizing action of the flame.—E. N.

Metallizing to Prolong Life of [Oil-] Refinery Equipment. Joe E. Young, Jr. (*Petroleum Refiner*, 1951, 30, (7), 118–120).—The characteristics and advantages of metallizing are briefly described, and details are given, step by step, of how to use metallizing for building up a worn part. Brief information is also presented on the spraying of an acid-resisting alloy, Colmonoy No. 6. The deposit has to be fused after appn. Metallizing for protecting flat surfaces in plant is described and information included on what precautions are necessary to ensure satisfactory results.—G. T. C.

Some Methods of Protection Against Metallic Corrosion. T. A. Evans (*J. Inst. Prod. Eng.*, 1951, 30, (2), 62–78).—A review of the mechanisms of corrosion of metals is followed by a description of the various means of protection, including inhibitors and coatings.—M. A. H.

Hot-Dip Galvanizing. (——). See col. 335.

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8 — ELECTRODEPOSITION

The Universal Cold Chrome Process. T. J. Hollinrake (*Australasian Plating and Finishing*, 1951, 1, (3), 5, 7 ; discussion, (4), 13, 15).—Brass and Cu- and Zn-base die-castings are plated directly with Cr to ~0·00016 in. thickness. Steels must first be plated with another metal, Zn being usually preferred. Bath efficiency is ~40%. Compn. of the soln. is not given.—T. A. H.

Conservation of Nickel in Plating. —— (*Metal Ind.*, 1951, 79, (7), 128–130).—Gives general and sp. suggestions of a practical nature by the British Non-Ferrous Metals Research

Association Research Committee for preventing waste in Ni plating. It is recommended that accepted standards of thickness of Ni deposits should continue to be upheld and that substantial economies can be effected by reducing drag-out losses.—J. H. W.

Choose the Right Nickel-Plating System. —— (*Steel*, 1951, 128, (23), 94–96, 117, 118, 121, 124, 127, 130, 132).—A general review is given of the properties and uses of Ni plating, and of the baths and techniques used in the plating of various items.—E. J.

Development, Production, and Manufacture of Electro-Tinplate. W. E. Hoare (*J. Inst. Prod. Eng.*, 1951, 30, (3), 104–131).—A description of the stannous phenolsulphonate, halide, and alkaline stannate baths, with detailed descriptions of modern plants exemplifying the use of each bath, is followed by a brief comparison.—M. A. H.

Tin Research Institute Announce New Electroplate. —— (*Product Finishing*, 1951, 4, (6), 64–65).—A note on a new Sn–Ni electrodeposit (cf. *Met. Abs.*, this vol., col. 138) and on the opening of the new Research Laboratories of the Tin Research Institute.—H. A. H.

*The Structure of Electrodeposited Metals and Alloys.** Ernst Raub (*Z. Elektrochem.*, 1951, 55, (2), 146–151).—The influence of polarization and the presence of impurities in the electrolyte upon the properties and state of electro-deposited metals, particularly Ag and, to a smaller extent, Au, Cu, and alloys of these metals, Pb, and Fe, was studied. Electrolyte impurities adsorbed by the cathode, and incorporated in the deposited metal, produce marked lattice distortion, leading to considerable alteration to all structure-sensitive properties of the deposited metal. By the simultaneous discharge of two cations, alloys are produced in which phases similar to those characterizing the appropriate equilibrium diagram are present. Under some conditions, unstable types of crystals may be deposited at room temp., and the phase boundaries may be much displaced from those of the equilibrium diagram. Markedly supersaturated solid soln. can be produced by electrolysis as a consequence of the simultaneous incorporation in the deposit of non-metallic impurities, and these solid soln. can differ considerably in their properties from those of supersaturated crystals produced by recrystn.—J. S. G. T.

The Hull Cell : Use in the Control of Electroplating Processes. J. W. Cuthbertson (*Metal Ind.*, 1951, 79, (5), 87–90).—The Hull cell has a trapezoidal cross-section. At the ends are placed the electrodes, so that the c.d. over the cathode varies smoothly and in a known manner. The cell is calibrated, i.e. the distribution of the effective c.d. is found experimentally by plating a cathode, cutting the deposit into strips, and estimating the weight of deposit on each. C. describes the use of this cell and gives details of its appn. in the control of the stannous fluoborate and the Sn–Ni baths. 5 ref.—J. H. W.

pH **and Buffering.** —— (*Metal Finishing*, 1951, 49, (6), 80).—Engineering data sheet. The *pH* values for various acids and bases commonly used in metal-cleaning and plating operations are included in a table. In addn. some of the more common buffer mixtures and the range of *pH* over which they offer max. buffering capacity are presented.—G. T. C.

Synthetic Surface [-Active] Agents in the Engineering and Plating Industries. —— (*Australasian Plating and Finishing*, 1951, 1, (4), 5–6).—A brief review.—T. A. H.

Exposition of Principles Governing Electroplating. H. E. Arblaster (*Australasian Plating and Finishing*, 1951, 1, (4), 9–11, 21).—A brief review.—T. A. H.

Neutralizing Industrial Waste Liquids. —— (*Steel*, 1951, 128, (15), 74–77, 95, 98, 101, 104, 108, 111, 112).—Report of the Fifth Annual Industrial-Waste Conference sponsored by the Purdue University and the Indiana State Board of Health. Describes various treatments for waste liquids, including effluents from pickling and Ni-, Cu-, and Cr-plating baths.
—E. J.

Electroplating for the Metallurgist, Engineer, and Chemist. (Mohler and Sedusky). See col. 336.

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9 — ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

Electrochemistry and the Science of Metals. Roberto Piontelli (*J. Inst. Metals*, 1951–52, 80, (3), 99–107).—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.—AUTHOR.

*Electrochemical Studies of Copper–Platinum Alloys.** Brigitte Sarry (*Z. anorg. Chem.*, 1950, 261, 187–196 ; *C. Abs.*, 1950, 44, 6304).—In the cell : alloy|0·983 M-CuSO$_4$|2N-H$_2$SO$_4$| Hg$_2$SO$_4$|Hg in a N atmosphere, the potential approximated to that of pure Pt down to alloys contg. 16·5 at.-% Pt. At lower Pt concentrations the potential approached that of pure Cu. In the presence of air the potentials slowly approached those of the O electrode. The (O) decomposition of 2N-H$_2$SO$_4$ had a lower potential on alloys contg. >16·5% Pt than on those contg. less Pt. The decomposition potential with the 2% Pt alloy was unusually low. The trends in the potentials showed analogies with the catalytic activity of the alloys.

*Hydrogen Overvoltage at Individual Cadmium [Single] Crystal Surfaces.** E. Albrecht (*Z. Elektrochem.*, 1951, 55, (2), 173–174).—The H overvoltage at the (0001) basal plane and at the rounded surface region of an individual Cd single crystal was studied, voltage current graphs being obtained when the crystal was used as cathode in conjunction with a large Pt anode in N-H$_2$SO$_4$. In all cases, log c.d. plotted against voltage gave straight-line graphs. The average value of b in Tafel's equation, giving the slope of the straight lines, is 0·076 for the rounded region and 0·129 for the (0001) basal plane. The results, taken altogether, suggest that the curved region of the crystal was insufficiently cleaned or that impurities were imperfectly removed therefrom.
—J. S. G. T.

*Investigations on the Electrodeposition of Manganese.** [—I.–II.]. M. Perec (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, 1, (2), 131–138 ; 1950, 2, (2), 93–99).—[In Polish]. [I.—] P. first outlines the theoretical bases of the electrodeposition of Mn. In the presence of NH$_3$ or of NH$_4$ salts the cathodic polarization curves are displaced in the direction of greater negative potentials. This shows that the concentration of the simple ion in the soln. decreases and enables the ratio of concentration of complex ion to simple ion to be calculated. The effect has been determined of varying : (a) the concentration of NH$_4$ salts, (b) the concentration of Mn salts, and (c) time of electrolysis. Observation of the effect of inhibitors on the course of electrolysis indicates that they are not adsorbed on the Mn, but they reduce the number of dendrites formed during deposition. The disadvantage of inhibitors lies in the fact that they cause an increase in the S content of the Mn deposit. The presence of inhibitors also leads to a periodic structure in the deposit, with alternate layers of pure Mn and metal rich in S ; this is due to the periodic variation of the ratio of the S ion to the Mn ion. [II.—] The effect of Fe^{2+}, Fe^{3+}, Pb^{2+}, Ni^{2+}, and Cu^{2+} ions on current efficiency has been studied. It was found that after 5 hours' electrolysis 1·5 mg./l. Co^{2+} or 2·5 mg./l. Ni^{2+} had reduced the current efficiency by 50%. Co proved the most deleterious impurity ; at concentrations of 2 mg./l. it dissolved nearly all the Mn. Potential measurements on electrodes with partially dissolved Mn showed that at $pH = 8$, Co is deposited on the cathode and a cell is set up in which Mn dissolves, being the anode. Cathode polarization curves taken in a soln. contg. (NH$_4$)$_2$SO$_4$ (130 g./l.) and Na$_2$SO$_3$ (0·35 g./l.), using smooth cathodes of acid-resisting steel on which powdery Co had been deposited, showed that in this form Co markedly reduces the H over-voltage. It is inferred from this that in baths contg. small amounts of Co, H–Mn cells are set up in which Mn acts as the anode. The H electrode reduces the polarization of the cathode and prevents deposition of Mn ; H is vigorously evolved instead. A rapid polarographic method for the detn.

of Co in soln. used for the electrolytic prodn. of Mn has been worked out.—N. B. V.

*Adsorption of Hydrogen by Platinum [Electrodes] as Determined by Polarization-Capacity Measurements. A. Eucken and B. Weblus (*Z. Elektrochem.*, 1951, **55**, (2), 114–119).—The occurrence of well-marked electrode-capacity max. at metallic electrodes has been observed by Schaper (Dissert., Göttingen, **1949**). Measurements obtained by using a small Pt anode and a large non-polarizable Pt cathode, with molecular H as depolarizer, indicate the existence of a marked max. capacity of the anode at $+130$ mV. and a smaller max. capacity at $+250$ mV. These results indicate that Pt surfaces possess two different types of strongly adsorbing regions. The significance of these results in relation to the prodn. of H overvoltage is discussed, and quant. expressions for the adsorption isotherms are derived.
—J. S. G. T.

Hydrogen Overvoltage on a Platinized Platinum Electrode: Contribution to a Theoretical Study. René Audubert and Maurice Bonnemay (*Compt. rend.*, 1950, **230**, (12), 1171–1173). —It is shown that abnormally large values of α (the transfer coeff. of H discharge) observed experimentally for the H overvoltage on platinized Pt can be explained by supposing that the discharge depends, at any rate in a certain range of c.d., on the process of adsorption of the atoms on the surface of the electrode.—N. B. V.

*On the Study of Periodic Oscillations Obtained During the Anodic Solution of Silver. Claude Chalin (*Compt. rend.*, 1951, **233**, (5), 362–364).—C. briefly summarizes the history of the mechanism of periodic oscillations of current and voltage during the anodic electrolysis of Ag in a soln. of KCN. He gives the results of his own experiments in measuring the impedance of a Ag wire anode in an aq. soln. of KCN (17 g./l.) with a Pt cathode.—J. H. W.

The Electrochemical Character of Corrosion and Other Reactions in a Metal|Solution System : The Potential/pH Diagram of the System Silver|Solution According to M. Pourbaix. Kurt Nagel (*Z. Elektrochem.*, 1951, **55**, (2), 144–145).—N. contends that to obtain a clear and complete knowledge of the nature of the reactions occurring at a metal|electrolyte interface it is necessary to take into account all possible reactions controlling potentials at phase boundaries, and analyses the results given by Delahay, Pourbaix, and Van Rysselberghe for the potential/pH diagrams of the system Ag|soln. (*J. Electrochem. Soc.*, 1951, **98**, 65 ; *Met. Abs.*, 1950–51, **18**, 719) in which not only univalent Ag^+ ions, but also Ag^{++}, AgO^+, H^+, OH^- ions, solid oxide phases Ag_2O, Ag_2O_2, and Ag_2O_3, and gas phases H_2 and O_2 must be taken into account, from this point of view. The analysis distinguishes three regions in the diagram, viz. a region of immunity, an intermediate one of corrosion, and one of passivity due to oxide formation.—J. S. G. T.

*A Study of Metallic Electrodes Prepared by Sublimation. Benjamin C. Bradshaw (*J. Chem. Physics*, 1951, **19**, (8), 1026–1027).—Electrodes prepared by vacuum condensation of Zn on Pt wires were practically strain-free. The e.m.f. of the Zn crystals was $\sim$100 μV., and it is considered that the electrodes would be highly suitable in reversible appn.
—J. R.

On the Principles of Electrodeposition of Metal Powders. Gösta Wranglén (*Acta Polytech.*, 1950, **3**, (2), 40 pp.).—[In English]. Cf *Met. Abs.*, 1950–51, **18**, 381. It is shown that two factors are responsible for the prodn. of powdery and spongy metal deposits at the cathode, depletion of the cathode film, and hydroxide formation in the cathode film. If the polarization curve of the metal is steep, depletion is likely to be responsible and is favoured by low concentration, high c.d., low temp., absence of agitation, and the presence of neutral salts. Generally speaking, the same factors favour H-ion depletion and therefore hydroxide formation, so that the two factors may each contribute to the powder formation. In some cases, however, e.g. Zn and Cd in neutral soln., the conditions of metal and H overvoltage have a pronounced influence on the cathode film pH, so that powder formation may be favoured by low c.d. and

high temp. Detailed information is presented on the prodn. of Cu, Fe, Ni, Zn, and Cd powders by electrolysis, and a series of photographs is included of powders produced under different conditions. Brief reference is made to the prodn. of powdery deposits from fused salts by electrolysis. Other theories of the mechanism of metal-powder formation by electrolysis are briefly examined and criticized. 59 ref.
—G. T. C.

†Anodic Polishing and Its Relation to Anodic Passivation. Kurt Huber (*Z. Elektrochem.*, 1951, **55**, (2), 165–169).—Literature relating to anodic polishing of metals is reviewed, and the following conclusions are drawn : (1) under polishing conditions the anode metal surface is oxidized and rendered passive ; (2) the covering deposit so produced is removed at once, as otherwise the dissoln. of metal, necessary for polishing to continue, would cease. Diffusion processes in the liq. anode film control the rate of removal of the oxide layer, and its reprodn., and hence also the c.d. Thus results the characteristic independence of the current, necessary for the polishing process, of the applied voltage. A study of the voltage/current curve for any combination of electrodes and electrolyte will indicate what properties the electrolyte must have to be useful for polishing. 49 ref.—J. S. G. T.

*A Critical Discussion of H. H. Uhlig's Electron Configuration Theory of Passivity. Theo Heumann (*Z. Elektrochem.*, 1951, **55**, (4), 287–295).—The oxide-film theory of the passivity of Fe and other metals is briefly discussed, and attention is directed to Uhlig's electron-configuration theory of passivity (*Trans. Electrochem. Soc.*, 1944, **85**, 307 ; see *Met. Abs.*, 1945, **12**, 87) which attributes passivity to the unfilled electron d-shell of the transition metals. Filling-up these d-shells with 10 electrons, brought about either by alloying or by adsorption layers, removes the passivity. The known phys. properties of metals and alloys indicate that this process of electron-exchange is, save in the case of Ni, very improbable ; even should electron exchange occur, it is likely to take place in the opposite direction to that required. In heterogeneous alloys constitutional characteristics must be taken into account.—J. S. G. T.

*Investigations on Local-Cell Models Regarding Current-Density Distribution and *Modus Operandi* (at Electrode Surfaces), Based Upon Measurements of Overvoltage. K. Dürkes (*Z. Elektrochem.*, 1951, **55**, (4), 280–287).—Investigations on the relation between potential and c.d. at electrodes in soln. of electrolyte, carried out with apparatus comprising local-cell models of the type described by Jaenicke and Bonhoeffer (*Z. physikal. Chem.*, 1944, [A], **193**, 301 ; *Met. Abs.*, 1947, **14**, 473) lead to the deduction that certain conclusions relating to the polarization of platinized Pt electrodes need to be revised. D. finds that, at considerable concentration-polarization, limiting current can exist at small cathodes and at the boundary of large platinized Pt cathodes. Also, the c.d. at large cathodes is limited by processes occurring at the cathode boundaries. Recombination of electrolytically deposited H atoms to form molecules occurs as a process at the boundary regions. It is shown that Ohm's law affords a fair approx. to the elect. processes taking place in the local-cell type of apparatus.—J. S. G. T.

*Evans [-Type Local] Cells : A Reply to Masing's Observations on the Theory of Corrosion by Local Cells. (Wickert and Wiehr). See col. 299.

*The Significance of the Activity of Metal Surfaces in Electrochemical Reactions. (Wickert). See col. 300.

†Studies of Anodic and Cathodic Polarization of Metals. R. Piontelli (*Z. Elektrochem.*, 1951, **55**, (2), 128–143 ; discussion, 143).—Theoretical and experimental researches on the anodic and cathodic polarization of metal electrodes, carried out at the Politecnico, Milan, in recent years, are comprehensively reviewed. 21 ref.—J. S. G. T.

†Electrochemical Fundamentals Relating Particularly to the Metal|Solution Electrode. Erich Lange (*Z. Elektrochem.*, 1951, **55**, (2), 76–92).—After an introduction stressing the need for rationalizing, simplifying, and clarifying, possibly by international discussion, the multiplicity of ideas, notations, and symbols used in electrochemistry more especially

as applied to the metal|soln. interface, L. discusses, briefly, some fundamental principles relating to elect. fields, the definitions of " potential " and " tension ", electrode potentials and other electrochem. notions essential to the discussion of single-, two-, and three-phase systems. 20 ref.—J. S. G. T.

The Significance of Inhibitors and Poisons in the Study of Electrochemical Processes at Metal[|Liquid] Interfaces. Hellmut Fischer (*Z. Elektrochem.*, 1951, 55, (2), 92–97).—F. briefly discusses : (1) the effect of inhibitors upon concentration-, blocking-, crystn.-, poisoning-, and transport-overvoltages in electrochem. processes ; (2) the reduction and oxidation of inhibitors ; (3) inhibitor sensitivity ; (4) the mechanism of the coating of electrode surfaces by an inhibitor due to current flow ; and (5) changes in the course of crystal growth. The presence of inhibiting centres (and c.d.) affects not only the form and orientation of crystal deposits, but produces also corresponding changes in the sub-crystallites of the crystal mass. Changes of inhibition produce different forms of crystal growth. Thus, very weak inhibition favours the prodn. of field-oriented sub-crystallites and isolated aggregates of these sub-crystallites growing parallel to the lines of current flow (the so-called field-oriented isolation type of crystallites, F.I.). Weak inhibition by furthering growth in breadth and checking formation of nuclei, favours the formation of the basal-oriented reproduction type of crystal (B.R.). Conditions for the prodn. of field-oriented prisms, plates, needles, &c., the so-called field-oriented texture type (F.T.), and the non-oriented dispersion type (U.D.) of crystals are briefly referred to. 22 ref.—J. S. G. T.

*Determination of the Exchange Velocity at Equilibrium Potential by Polarization Measurements Using D.C. and A.C.** H. Gerischer (*Z. Elektrochem.*, 1951, 55, (2), 98–104).—The establishment of a potential-determining equilibrium at an electrode ensues as the result of the interplay of two opposed chem. processes concerned with the exchange of elect. charges across phase boundaries. The velocities of these processes determines the behaviour of the electrode when polarized, and hence one can deduce from measurements of polarization resistance at an electrode near its equilibrium potential conclusions regarding the velocity of exchange. Conditions to be satisfied by electrodes used for this purpose, by the use of D.C. and A.C. in such experiments, and results obtained using a $Hg|Hg^{++}$ electrode in a soln. of $Hg_2(ClO_4)_2$ contg. a large excess of $HClO_4$, and at a Pt electrode in a " Redox " (reduction–oxidation) system comprising Fe^{++} and Fe^{+++} ions in KNO_3 soln., are given.—J. S. G. T.

On the Hydrogen Overvoltage. (Mlle) Geneviève Sutra (*Compt. rend.*, 1950, 230, (7), 644–645).—Theoret. S. regards it as very probable that, for all metals, adsorption of protons takes place until a certain surface concentration is reached. Discharge of H begins when the cathode is saturated with it ; at this point a relation exists between the concentration of protons inside and those outside (k), which varies with the affinity of the metal concerned for protons.—N. B. V.

On the Hydrogen Overvoltage. (Mlle) Geneviève Sutra (*Compt. rend.*, 1950, 230, (8), 749–750).—On the basis of her previous work (*ibid.*, (7), 644 ; preceding abstract), S. deduces that the H^+ ions adsorbed on the cathode set up a field sufficient to extract electrons from the metal in the cold, and that therefore some relation should exist between the H overvoltage (η) and the work-function (χ). Comparison of published values of η and χ failed to reveal any such correlation however.—N. B. V.

Hydrogen Overvoltage in the Course of Electrolysis. (Mlle) Geneviève Sutra (*Compt. rend.*, 1950, 230, (9), 842–844).—Overvoltage is particularly important in the case of cathodes having a low solubility for H. The slow discharge of H at

the cathode is considered as a reversible process, and it is shown that the overvoltage is independent of the pH of the electrolyte and also of the pressure. Overvoltage (η) is related to the temp. (T) by the equation : $\frac{1}{\eta} \cdot \frac{d\eta}{dT} = \frac{b}{T^2 \ln S}$, where S is the solubility of H (as protons) in the cathode. When the current is increased, polarization occurs, which is regarded as the adsorption of ions on the metal cathode and extraction of electrons from the metal in the cold by the ions. The work of extraction is a function of the dielect. const. of the electrolyte and addn. of ethyl or methyl alcohol consequently reduces η. Overvoltage is greater on a polished metal surface than on a rough surface of the same metal, since adsorption of ions is easier on the latter. For polished Pt and platinized Pt the difference is 0·1 V.—N. B. V.

*†**The Kinetics of Hydrogen Evolution as a Function of the Electrode Material.** J. O'M. Bockris (*Z. Elektrochem.*, 1951, 55, (2), 105–111).—The comparative ease with which H evolution at a clean metal|electrolyte interface can be studied in the case of the metal Hg has caused most of the few available investigations of the kinetics of H evolution to be carried out with the aid of Hg ; the effect of the cathode material upon the evolution has not been considered. Contemporary knowledge in this latter field is presented and critically reviewed so far as the sparse reliable data permit. Recent measurements of overvoltage reveal the existence of three groups of cathode metals characterized by different mechanisms of H evolution, viz. (1) Hg, Pb, Tl, Sn ; (2) W, Ni, Mo, Nb, Ba, Au, Fe ; (3) platinized Pt, Pd. The chief factor controlling the measured overvoltage in group 1 is the heat of adsorption of H ; in group 2 it is the difference of the heats of adsorption of O and H at the metal surface ; sufficient evidence is not available to determine the principal factor in group 3. 25 ref.—J. S. G. T.

The Significance of the Nature of the Electrode Metal in Electrolytic Reduction and Oxidation Processes. M. von Stackelberg (*Z. Elektrochem.*, 1951, 55, (2), 120–121).—Illustrative examples are given of the influence of the metal used as electrodes in electrolytic reduction and oxidation processes upon the speed of the process and the nature of the end-product, principally in connection with the electrolytic reduction and oxidation of organic compounds. Thus, nitrobenzene is reduced at Pb and Hg cathodes to aniline, but at a Ni cathode, which has a much lower H overvoltage, it can be reduced only as far as phenylhydroxylamine.
—J. S. G. T.

On the Determination of the Reaction Mechanism [in Electrolytic Processes] by the Formation of Redox Potentials. Klaus J. Vetter (*Z. Elektrochem.*, 1951, 55, (2), 121–127). [*Note by Abstractor* : Redox is a contraction, which is being increasingly used to signify reduction and oxidation.] Transfer-, diffusion-, reaction-, and resistance-polarization in electrolytic processes are discussed as a preliminary to the introduction of reduction and oxidation potentials into a theory of the reaction mechanism controlling such processes, and a general equation for the relation of c.d. to potential is derived. Theoretical deductions agree with experimental results obtained by Vetter and Manecke for the Mn^{1+}/Mn^{3+} Redox system (*Z. physikal. Chem.*, 1950, [A], 195, 337) and for the cathodic polarization of a Pt electrode in a HNO_3/HNO_2 Redox system (*ibid.*, 1950, [A], 194, 199).
—J. S. G. T.

*Calculations of the Energy Relationships During the Discharge of Ions at Electrodes.** Roger Parsons (*Z. Elektrochem.*, 1951, 55, (2), 111–114).—A thermodynamic treatment of the energy relations during the elect. discharge of ions at an electrode leads to a value of the heat of activation of the process which agrees reasonably well with experimental values. 40 ref.—J. S. G. T.

11 — ANALYSIS

*Observations on the Arc Spectra of Aluminium Electrodes in Air and in Hydrogen. Raymond Ricard and Alexis Dufour (*Compt. rend.*, 1951, **233**, (5), 370–372).—R. and D. photographed the spectra of arcs struck between Al electrodes in air and in H. They concluded that spectrographic observations of commercial Al under the usual conditions (arc in air) do not give the greatest sensitiveness for detecting the different impurity atoms; a better result is obtained by arcing successively in air and in H, an easy proceeding with a suitably constructed discharge tube.—J. H. W.

*Effect of the Physical Condition [on the Spectrum] in the System Sb–Sn. J. M. Lopez de Azcona and A. Camuñas Puig (*Rep. 13me Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, 1950, 32).—Abstract only. By examination of the spectra of metallic electrodes sparked in an oven at temp. up to 300° or 500° C., the evolution of the various phases of the binary system Sb–Sn can be traced.
—E. van S.

*An Empirical Titrimetric Method for the Determination of Beryllium. J. H. McClure and C. V. Banks (*U.S. Atomic Energy Commission Publ.*, 1950, (ISC-109), 12 pp.).—The Craig method (*J. Soc. Chem. Ind.*, 1911, **30**, 184) for titrating Al was modified for the detn. of Be. The Be soln. is adjusted to a pH of 8·5, NaF or KF is added, and the liberated OH⁻ is titrated with a standard acid soln. A precision better than 2 parts/1000 is obtainable.—B. W. M.

*Estimation of Beryllium in Chromium–Beryllium Alloys. E. J. Lumley (*Commonwealth Australia, Dept. Supply Development, Aeronaut. Research Lab. Rep.*, 1950, (SM-183), 9 pp.).—Preliminary removal of Cr by filtering $Be(OH)_2$ precipitate from soln. of Cr^{6+} reduces the amount to be removed by Hg-cathode electrolysis. $Be(OH)_2$ precipitated by NH_3 or α-picoline gave results of ~5% accuracy.—S. M.

*Separation of Chromium from Vanadium by Extraction of Perchromic Acid with Ethyl Acetate. Robert K. Brookshier and Harry Freund (*Analyt. Chem.*, 1951, **23**, (8), 1110–1113). —Cr is separated from V by extraction with three successive portions of ethyl acetate of a soln. contg. 0·02 moles/l. H_2O_2 at $pH = 1·7 \pm 0·02$ and $\ngtr 20°$ C. and then determined with s-diphenylcarbazide.—F. M. L.

*Isolation and Determination of Cobalt as Nitroso-R-Salt Complex by Chromatographic Ion-Exchange. John A. Dean (*Analyt. Chem.*, 1951, **23**, (8), 1096–1097).—Co can be determined in steels, &c., Cu, and Ni alloys by dissolving the sample in suitable acid, filtering, if necessary, the resultant soln. and, if much HNO_3 or HCl is present, taking to fumes with $HClO_4$, diluting, and then adding nitroso-R-salt and passing the soln. down a column packed with $HClO_4$-washed Al_2O_3. The Co nitroso-R-salt complex forms a red band at the top of the column and unused salt a yellow band below it. The unused salt is first eluted with M-HNO_2 and the Co complex then eluted with M-H_2SO_4 and the Co determined in known manner.—F. M. L.

*Colorimetric Determination of Copper in Aluminium Alloys. S. A. Shrager and George Norwitz (*J. Appl. Chem.*, 1951, **1**, (8), 354–355).—Cu in Al alloys can be determined in 7 min. as follows. To a 1-g. sample in a 500-ml. Erlenmeyer flask, add 30 ml. HCl. After reaction ceases, add 7 ml. HNO_3, boil for 1 min. to complete soln., and add 300 ml. H_2O and 100 ml. NH_4OH. Dilute to the 500-ml. mark, stopper, shake, permit the precipitate to settle for $\geqslant 2$ min., and filter through 12·5-cm. No. 41 filter-paper set in place with H_2O. Collect filtrate in 125-ml. Erlenmeyer flask, discard first 50-ml. portion, and read the colour on the second 50-ml. portion at 580 mμ in a photoelect. colorimeter set to zero with H_2O. Convert colorimeter readings to Cu % by ref. to curve prepared from samples of known Cu content, correcting for Ni, if present (1% Ni = 0·05% Cu). Some of the tabulated results are compared with detn. by the umpire method of the Aluminum Co. of America. 20 ref.—J. R.

*Determination of Copper as Cuprous Thiocyanate and a Separation from Arsenic and Zinc. Imre Sarudi (née Stetina)

(*Z. anal. Chem.*, 1950, **130**, 301–303; *C. Abs.*, 1950, **44**, 6336).—Contrary to former findings, an excess of thiocynate is necessary to complete the reaction $2Cu^{++} + 2CNS^- + SO_3^{--} + H_2O = Cu_2(SCN)_2 + SO_4^{--} + 2H^+$; for 0·1 g. Cu, at least 19 ml. of 0·1 N-thiocynate soln. should be used. The precipitate can be washed with cold water, dried at 100°–180° C. and weighed. To separate Cu from Zn proceed in the same way, first adding excess SO_2 soln., then neutralizing with NH_4OH, dissolving any precipitate in SO_2 soln. and then adding excess thiocyanate. If As^{5+} is present, more SO_2 is required. If the precipitate is greenish, add more SO_2. To determine As, it is best to use another sample, distil off $AsCl_3$, and titrate the distillate with I. To determine As in the filtrate from the $Cu_2(SCN)_2$ precipitate, oxidize the SO_3^{--} with HCl + Br, add 0·5 g. of HCO_2Na to remove excess Br, after 15 min. add 1 g. KCNS and precipitate $As_2S_3 + S$ with H_2S at room temp. Dissolve the filtered precipitate in NH_4OH, oxidize with perhydrol, and determine the As as $Mg_2As_2O_7$ after pptn. as $MgNH_4AsO_4$. Zn can be determined as $ZnNH_4PO_4$ after removing the $Cu_2(SCN)_2$ precipitate; in this case, KCNS, rather than NH_4CNS, should be used to precipitate the Cu. The results of ~25 experiments were excellent.

*Analysis of Brass by Direct Spectroscopy. A. Hans (*Rep. 13me Congr. Groupement Avancement Méthodes Anal. Spectrograph. Produits Métallurg.*, 1950, 75–79).—The direct spectroscopic analysis equipment at the Centre National de Recherches Métallurgiques at Liège has now been used for the estimation of Zn in brass in the range 12–40%. A simple comparison of the relative intensity of Zn and Cu lines cannot allow for the variation of the Cu content; but by the use of a critically damped spark source and the proper selection of a Zn line, the calibration can be made giving a smooth curve when the Zn line voltage multiplied by 1% of the Cu concentration is plotted against % Zn. A precision measured by a standard deviation of ~1% is claimed, but the wave-lengths measured are not specified.
—E. van S.

*The Determination of Germanium. I.—Titration of Mannito-Germanic Acid. II.—Absorptiometric Determination with Phenylfluorone. III.—Determination in Flue Dust, Coal, and Coke. H. J. Cluley (*Analyst*, 1951, **76**, (906), 517–522, 523–530, 530–535; discussion, 535–536).— [I.—] After a preliminary sepn. as sulphide, Ge is determined by titration, with standard NaOH, of mannito-germanic acid by the same method as that used for the detn. of B. As does not interfere with the titration provided that it is in the trivalent state. [II.—] Small amounts of Ge can be determined absorptiometrically in HCl soln. using phenylfluorone. The method is more sensitive than the Mo blue method and is advantageous, since it can be used directly on the distillate obtained from a sepn. of Ge based on distillation of $GeCl_4$. [III.—] The appn. of method II to the detn. of Ge in flue dusts, coal, and coke is described.
—F. M. L.

*The Determination of Indium in Beryllium Compounds. G. W. C. Milner (*Analyst*, 1951, **76**, (905), 488–492).—In is extracted from a H_2SO_4 soln. of Be at $pH = 3·5$–$4·5$ into a chloroform soln. of 8-hydroxyquinoline. The chloroform is boiled off the extract, H_2SO_4 added, and the soln. taken to fumes; organic matter is destroyed by drop-wise addn. of HNO_3 and final addn. of $HClO_4$, and the soln. is then taken to dryness. The salts are taken up in HCl, any Fe and/or Mo extracted into ether, and the In then determined polarographically in an electrolyte comprising HCl contg. Na formate, hydrazine, and starch.—F. M. L.

*The Micro-Estimation of Iron with Triphenylmethylarsonium Thiocyanate. F. P. Dwyer and N. A. Gibson (*Analyst*, 1951, **76**, (906), 548–550).—1–10 γ Fe may be determined colorimetrically by the extraction into o-dichlorobenzene of the red complex formed by Fe with triphenylmethylarsonium thiocyanate.—F. M. L.

*Rapid Titrimetric Determination of Magnesium in Aluminium Alloys by the Phosphate Method. Hans Ulrich v. Vogel (*Z. anal. Chem.*, 1950, **130**, 304–314; *C. Abs.*, 1950, **44**, 6338).—Digest 1 g. of alloy shavings in 15 ml. of 25% NaOH, add 5 ml. of 3% H_2O_2, and boil. Dilute with hot water to 150 ml., filter, and wash the residue with hot water. Dissolve the residue in 30 ml. of hot $6N$-HNO_3, with a little H_2SO_3 if necessary to dissolve MnO_2. Add 50 ml. of a reagent prepared from H_2O 8 l., $7.5N$-HNO_3 2 l., tartaric acid 2 kg., and $(NH_4)_2HPO_4$ 600 g. Add also 65 ml. of conc. NH_4OH. After suitable washing dissolve the $MnNH_4PO_4.6H_2O$ and $MgNH_4PO_4.6H_2O$ precipitate in 20–50 ml. of standard H_2SO_4 and titrate the excess with standard NaOH soln. to a methyl-red end-point. This leaves $H_2PO_4^-$ in soln., and the H_2SO_4 required is a measure of the Mn + Mg contents. The Mn is determined in a separate sample by the per-sulphate method after dissolving the Al with NaOH. For alloys of Al, Zn, and Mg, 200 g. citric acid can replace the 2 kg. of tartaric acid in preparing the precipitant soln.

*The Determination of Selenium and Tellurium in Copper. F. D. L. Noakes (*Analyst*, 1951, **76**, (906), 542–547).—The sample of Cu is dissolved in HNO_3 and the soln. boiled to expel nitrous fumes. After neutralization with NaOH until it is only just acid, hydrazine hydrochloride is added to decompose any remaining nitrous oxides and Se and Te precipitated by addn. of $SnCl_2$ and conc. HCl. The precipitate is filtered off and washed and then Se and Te separated and determined, grav. or vol., by known procedures.
—F. M. L.

*The Photometric Determination of Small Amounts of Titanium with 8-Hydroxyquinoline. K. Gardner (*Analyst*, 1951, **76**, (905), 485–488).—Small amounts of Ti can be determined absorptiometrically by extraction of the Ti from a H_2SO_4 soln. at $pH = 3$ contg. H_2O_2 into a chloroform soln. of 8-hydroxyquinoline and measuring the colour of the extract by means of a Spekker absorptiometer.
—F. M. L.

*Colorimetric Determination of Uranium with 1-Ascorbic Acid. G. G. Smith, E. F. Orlemann, A. A. Smaales, C. D. Rothenberger, and W. R. Grimes (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-2101), 16 pp.).—U^{6+} in weakly acid soln. buffered with CH_3COO^- produces, with a large excess of 1-ascorbic acid, a yellow-brown colour whose intensity is a linear function of U concentration up to 15 mg. U/50 ml. The colour intensity is independent of pH in the range 4.5–4.8, but adjustment of the pH to 4.0 ± 0.1 is recommended to minimize the danger of slow pptn. of hydrous ferric oxide. Addn. of pyridine along with acetate buffer minimizes the interference of Cu^{2+}. The colour intensity is a function of the concentration of CH_3COO^- and of 1-ascorbic acid, so the amounts of these reagents must be closely controlled. The method is accurate to $\pm 2\%$ in contaminated soln. Cl^-, NO_3^-, and ClO_4^- are without appreciable effect on the colour produced, but anions such as PO_4^{3-}, F^-, SO_4^{2-}, &c., which are capable of producing stable complexes with U in weakly acid soln., interfere by lowering the colour intensity. Moderate amounts of Fe, Cd, Mn, Na, Ca, Sn, Mg, and Th are without effect, but significant amounts of Cr, V, Ni, Zn, Mo, and Al interfere.
—B. W. M.

Determination of Small Amounts of Uranium by the Polarographic Method. S. B. Kilner (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-2059), 1 p.).—K. summarizes a method of determining down to 10 γ/c.c. U in a $Fe(OH)_3$ precipitate.—B. W. M.

Szilard–Chalmers Process for Uranium. D. S. Breslow and J. W. Hamaker (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-2409), 3 pp.).—Describes the successful sepn. of U^{237} from U^{239}, using salicylaldehyde-o-phenylenediamine uranyl complex.—B. W. M.

*Colorimetric Determination of Vanadium with Benzoylphenylhydroxylamine. Sudhir Chandra Shome (*Analyst*, 1951, **23**, (8), 1186–1188).—Benzoylphenylhydroxylamine is a sensitive reagent for the colorimetric detn. of V. Fe and Al interfere.—F. M. L.

*Rapid Electrolytic Method for the Determination of Zinc in Aluminium–Zinc Alloys. George Norwitz (*J. Appl. Chem.*, 1951, **1**, (8), 353–354).—Zn in Al–Zn alloys is rapidly and accurately determined by dissoln. of a 1-g. sample with 25 ml. NaOH soln. (15%) in a 300-ml. electrolytic beaker on a steam bath; 30 ml. H_2O, 20 ml. H_2SO_4 (1 : 1), and 5 ml. H_2O_2 (30%) are added and are heated to boiling, the boil being continued for some min. to destroy the peroxide. After cooling to room temp., 25 ml. tartaric acid (25%) are added, and the soln. is made just alkaline to methyl red by NH_4OH (1 : 1); tartaric acid is then added until the soln. is just acid to methyl red, when a few drops excess acid are added. After dilution to 190 ml. and electrolysis for 1 hr .at 2 amp./dm.^{2}, using Pt-gauze cathodes and Pt-spiral anodes, the cathodes are immersed in H_2O and then in alcohol, dried at 105° C. for 3 min., and cooled. The deposit is weighed as Zn + Cu + Fe, the Zn content being found by difference. Ni, Pb, and Sn interfere with the detn., but these are not usually found in Al–Zn alloys. Results obtained on two representative alloys are compared with detn. by the A.S.T.M. ZnO method. 11 ref.—J. R.

*Potentiometric Determination of Zinc in Magnesium Alloys. L. Ya. Polyak and F. M. Shemyakin (*Zavod. Lab.*, 1950, **16**, (1), 24–28; *C. Abs.*, 1950, **44**, 6340).—[In Russian]. Zn can be titrated potentiometrically by $K_4[Fe(CN)_6]$ in the presence of Mg, Mn, and Al with accuracy. Procedure : dissolve 1 g. sample in 25 ml. $6N$-H_2SO_4, filter off any residual Cu, add 1.5 ml. conc. H_2SO_4, 30 ml. saturated K_2SO_4, dilute to 100–120 ml., heat to 75° C., add 10–12 drops ferricyanide soln. (6 g./l.) and titrate at this temp., while stirring, with $0.025N$-$K_4[Fe(CN)_6]$ soln., using smooth Pt and calomel electrodes. If much Fe is present, add oxalate or pyrophosphate to remove Fe^{+++} after addn. of enough $KMnO_4$ to give a stable colour.

*Tannin as a Reagent for Zirconium. A. Purushottam and Bh. S. V. Raghava Rao (*Rec. Trav. Chim.*, 1951, **70**, (7), 555–558).—[In English]. With the exception of Sn and Ti, almost all elements commonly associated with Zr were found to be separated by single or double pptn. of the Zr from 0.1–$0.15N$-HCl soln., using tannin. Small quantities of Zr can also be estimated by this method.—J. W. C.

*Spark Analysis of Semi-Micro Samples. W. J. Wark (*J. Opt. Soc. Amer.*, 1951, **41**, (7), 465–468).—For the spectroscopic examination of small samples by spark-emission methods, samples of powders, metal sponge or foils are pressed into the end of a C rod 0.24 in. in dia. in a steel mould. The fractured surface of a 10-mm. piece of rod is used, since the adhesion of the sample is better ; after the specimen has been moulded, it is held by a cylindrical clip on to the end of a rod for use as a specimen. Remarkably low C-sample spectrum ratios are claimed, and reproducibility is good, as shown by figures quoted. Extension of this method to arc spectra may be possible.—E. VAN S.

The Use of Fluorescence in Qualitative Analysis. Charles E. White (*J. Chem. Education*, 1951, **28**, (7), 369–372).—Qual. tests for the detection of Al, Zn, Na, Be, Tl, and Th by fluorescence in ultra-violet light are described. The construction of a visual comparator is illustrated.—J. R.

Preparation of Filtered Samples for Radioactive Assay. T. H. Oddie and K. R. Lynn (*Commonwealth Australia, X-Ray and Radium Lab.*, *Tracer Elements Rep.*, 1951, (1), 6 pp.).—Apparatus, method, precautions, and recommended precipitates are given for the Geiger counting assay of Ag, Ba, Br, C, Ca, Cd, Cs, I, Mo, P, Pd, rare earths, Ru, S, Sr, Te, Zn, and Zr. Corrections for self-absorption must be made.—S. M.

Magnetic Mercury Cathode [for Analytical Separations]. E. J. Center, R. C. Overbeck, and D. L. Chase (*Analyt. Chem.*, 1951, **23**, (8), 1134–1138).—A magnetic circuit which provides rapid countercurrent stirring at the Hg-electrolyte interface and continuously cleans the surface of ferromagnetic elements in a Hg cathode cell is described.—F. M. L.

*The Stability of Analytical Weights, Particularly in Chemical Laboratories. (Bigg and Burch). See col. 299.

Errors in Weighing Caused by Electric Charges Developed in Silica Crucibles on Heating. A. W. Armstrong (*Analyst*, 1951, **76**, (906), 551–522).—In a note, it is stressed that errors in weighing of silica crucibles after heating in an elect. muffle may pass undetected when using an aperiodic balance, since there are no erratic beam movements to indicate the presence of an elect. charge.—F. M. L.

Zirconium [Crucibles] for Sodium Peroxide Fusions. G. J. Petretic (*Analyt. Chem.*, 1951, **23**, (8), 1183–1184).—Zr crucibles have been found very satisfactory for carrying out Na_2O_2 fusions.—F. M. L.

12 — LABORATORY APPARATUS, INSTRUMENTS, &c.

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

Characteristics and Behaviour of Bonded Wire-Resistance Strain-Gauges in Thermal Coefficient of Expansion Measurements. I.—SR-4 Paper-Bonded A-7 and Bakelite-Bonded AB-19 Gauges. M. R. Goodman (*U.S. Atomic Energy Commission Publ.*, 1950, (ORNL-706), 41 pp.).—By using bonded wire-resistance strain-gauges, the thermal expansion coeff. of metals can be measured to an accuracy $> \pm 0.1 \times 10^{-6}$ in the range $32.5°$–$60.0°$ C. The effects due to creep and zero shift in the gauge have to be differentiated from those of the specimen, and suggestions are made to eliminate these adverse properties to enable greater accuracy to be obtained.
—B. W. M.

A Radiographic Method of Dilatometry [of Metals]. A. H. Smith, N. A. Riley, and A. W. Lawson (*Rev. Sci. Instruments*, 1951, **22**, (3), 138–140).—A method of dilatometry using X-rays has been developed and applied to measurements of compressibility of Al, Fe, and Cu subjected to pressures up to 10,000 atm.—E. J.

***Low-Temperature Performance Test on Rupture Diaphragms.** W. L. Richardson and G. S. Storer (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-2268), 7 pp.).—Describes an apparatus for determining the effect of temp. down to $-100°$ F. ($-75°$ C.) on the bursting pressure of rupture diaphragms of Al and Ag.—B. W. M.

A Discussion of the Transpiration Method for Determining Vapour Pressure. J. V. Lepore and J. R. Van Wazer (*U.S. Atomic Energy Commission Publ.*, 1948, (AECD-1188), 8 pp.).—L. and Van W. correlate the variables of the transpiration process to find a guide for the optimum design of apparatus and the method of interpretation of experimental results.—B. W. M.

Thermodynamics of Metallic Solutions : Technical Progress Report No. 3, Scope I. [Measurement of Vapour Pressures of Metals]. J. Chipman and G. Scatchard (*U.S. Atomic Energy Commission Publ.*, 1950, (NYO-568), 2 pp.).—A spectrographic method was developed to measure the partial vapour pressures of metals over their alloys. Light from a spark source of the metal is passed through a vessel contg. the alloy in equilibrium with the metal vapour and the relative absorption of the two spectral lines, only one of which is absorbed by the vapour, measures the partial pressure of the metal. The apparatus is capable of high accuracy. A general theory for the correlation of activity coeff. in ternary alloys based on the assumption that positive ions in the metal react mainly with electrons and are relatively independent of one another, is under development.—B. W. M.

An Automatic Magnetic Balance for the Study of Ferro- magnetic Materials. R. F. S. Robertson and P. W. Selwood (*Rev. Sci. Instruments*, 1951, **22**, (3), 146–152).—An instrument is described for plotting, automatically, magnetization/temp. curves for ferromagnetic substances.—E. J.

***The Use of the Geiger–Müller Counter X-Ray Spectrometer in an X-Ray Laboratory, with Special Reference to Automatic Recording.** C. Wainwright (*Brit. J. Appl. Physics*, 1951, **2**, (6), 157–163).—The scope of this kind of apparatus in a laboratory where varied *ad hoc* problems have to be solved is considered, some instances being discussed, and a comparison with film methods is sketched. A detailed treatment is given of the statistical basis of automatic recording and of the effects of varying the counter time const.—R. W. C.

Graphical Method for the Simultaneous Determination of the Optical Constants and Thickness of Thin Metallic Deposits. Damien Malé (*Compt. rend.*, 1950, **230**, (14), 1349–1351).—Graphical methods previously described by M. (*J. Phys. Radium*, 1945, [viii], **6**, 40S and by Cotton (*ibid.*, p. 40S and *Ann. Physique*, 1947, **12**, 209) are applied to results obtained by Goos (*Z. Physik*, 1937, **106**, 606 ; *Met. Abs.*, 1938, **5**, 422) and Murmann (*ibid.*, 1936, **101**, 643 ; *Met. Abs.*, 1937, **4**, 29) on thin films of Au and Ag deposited on quartz.—N. B. V.

An Alignment Chart [Nomogram] for Computing the Thickness of Evaporated Films. A. L. Schoen and R. H. Davis (*J. Opt. Soc. Amer.*, 1951, **41**, (5), 362–363).—A nomogram is illustrated for calculating the thickness of a metallic film deposited from its vapour, in terms of the distance from source to sample, mass of metal evaporated, d of the metal, angle which the surface makes with the direction of the source, and a const. characteristic of the source depending on its shape. Names of the common metals are inserted on the d scale.—E. van S.

Development and Application of High-Vacuum Technique. M. Pirani (*Research*, 1950, **3**, (12), 540–542).—A general review. 48 ref.—F. A. F.

Fundamental Research in Physical Metallurgy : Progress Report. J. H. Hollomon and D. Turnbull (*U.S. Atomic Energy Commission Publ.*, 1951, (SO-2012), 5 pp.; also (RL-487)).—A couple consisting of single crystals of Cu and α-brass was prepared by electroplating the Cu at a slow rate in a warm electrolyte on a brass single crystal. This couple is being used for an investigation of the Kirkendall effect at $800°$ C. Tests are in hand to examine the possibility of using a mass spectrometer for vapour-pressure measurements which would be of value in the detn. of activity coeff. in alloys and of solid surface energies.—B. W. M.

13 — PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

***Application of Centrifugal Force to the Investigation of the Strength of Metal Systems. I.** I. I. Kornilov (*Doklady Akad. Nauk S.S.S.R.*, 1949, **67**, (5), 843–846).—[In Russian]. Cf. *Zavod. Lab.*, 1949, **15**, 76 ; *Met. Abs.*, 1950–51, **18**, 297. Tensile and bend (and also creep-strength and oxidation-resistance) tests can be carried out by mounting, inside a thermostatically controlled elect. furnace, 12, 24, or more cylindrical test-pieces some 3–5 mm. in dia., with one end clamped to a rotating shaft. Loading is varied by adjusting the speed of rotation of the shaft or by adding weights to the free ends of the specimens. The angle of bend or elongation of cooled specimens is measured by the usual methods at intervals of 15 min. to 100 hr. or more, the entire tests lasting from a few hr. to a few months. The apparatus is particularly suitable for the detn. of compn./strength diagrams, and can be employed for tests on : (1) low-m.p. metals and light metals and alloys from room temp. to m.p., (2) non-ferrous metals and alloys at $300°$–$900°$ C., and (3) high-m.p. metals and alloys at $500°$–$1200°$ C.—A. G.

***A Study of Fatigue in Metals by Means of X-Ray Strain Measurement.** John A. Bennett (*J. Research Nat. Bur. Stand.*, 1951, **46**, (6), 457–461).—Experiments were made to

investigate the claim of Glocker and his co-workers (*Z.V.d.I.*, 1941, **85**, 793; *Met. Abs.*, 1943, **10**, 186) that as fatigue damage increased there was a change in the value of the surface stress measured in a static test with a given load. The specimens used were made from S.A.E. 4130 steel and had a grain-size of 10 on the A.S.T.M. scale. The lattice strains when a series of static bending moments were applied to the specimens were determined by measuring the changes in X-ray-diffraction angles for X-ray beams incident ⊥ and at 45° to the surface. These static tests were made on specimens in the normalized state and after various amounts of fatigue stressing. The slope of the curve relating the bending moment to the lattice strain in the most highly stressed filaments was found to decrease with increased amounts of fatigue stressing, but the decrease was of the same order of magnitude for specimens stressed above and below the fatigue limit, and no further change took place when a specimen stressed for 51×10^6 cycles below the fatigue limit was stressed for ~75% of its life above the limit. B. concludes that this change cannot therefore be due to fatigue damage. His results indicate that even after millions of cycles of fatigue stressing plastic deformation occurred under a static stress less than the fatigue stress.—J. W. C.

*N.P.L. Investigates Rockwell Hardness Test. C. E. Phillips and A. J. Fenner (*Machinist* (*Eur. Edn.*), 1951, **95**, (32), 1183–1189).—P. and F. point out that considerable differences can arise between the results obtained from different Rockwell testing machines, and they investigated the test conditions specified in B.S. 891 : 1940, using first an experimental rig and then a special machine here described. They found that : (1) it is probable that many of the conical diamond indenters used in industry do not conform to the tolerance limits of radius and angle laid down in B.S. 891 ; (2) the hardness numbers obtained in the Rockwell C-Scale test are a function of the profile of the spherical tip of the indenter ; (3) there is no clear indication that small departures from the nominal angle of 120° have any important influence on the hardness number ; (4) errors in either of the applied loads have an important effect on the C-scale number, particularly with soft metals, but that this is less critical on the B-scale ; and (5) under a C-scale test load of 150 kg. unrecoverable creep of test material of low hardness took place over a period above 15 min., leaving doubt as to when the test should be regarded as complete.—J. H. W.

Comparability of Results of Hardness Measurements. J. Wozniacki (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1950, **2**, (3), 197–211).—[In Polish]. A review consisting virtually of a slightly condensed translation of pp. 13–70 of O'Neill's " Hardness of Metals and Its Measurement " (London, **1934**). Math. expressions and practical examples relating to comparability of Brinell, Meyer, and Ludwik cone hardness are reproduced from the book. According to Tammann and Krupkowski, yield stress $S = P/(\pi R^2)$, where P is the load on the indenter, and R the radius of indentation. Vitman and Zlatin (*Zavod. Lab.*, 1947, **13**, 990) claim, however, that $S = P/(\pi R^2) \times 1\cdot8$. 14 ref.—N. H. P.

*Triangular Diamond Pyramids for Micro-Indentation Tests. P. Grodzinski (*Research*, 1951, **4**, (5), 239–240).—A letter. Test results are presented which show that there is no particular advantage in the triangular diamond pyramid as an indenter.—F. A. F.

Damaged Impressions in Microhardness Testing. P. Grodzinski (*Research*, 1951, **4**, (3), 140–141).—A letter. A Vickers pyramid indentation may crack in the corners when hard and brittle materials are being tested, even at low loads. Impressions from a double cone indenter made under similar conditions show no sign of cracking.

—F. A. F.

Measurement of Mechanical Damping of Metals. Cord Petersen (*Arch. techn. Messen*, 1949, (164), V9115–9116; *Appl. Mechanics Rev.*, 1951, **4**, 225).—P. reviews various methods of measuring internal friction, after giving a general discussion of damping, indicating how it is evidenced, the usual measurement unit (log decrement), and the cause of damping. Factors which must be considered in its measurement and factors of which it is often a function are mentioned. Six measuring procedures are reviewed, from static measurement of the damping loop to heat-gain methods. Description of the action of sp. machines is restricted to six kinds, including the optical-loop recording machine of Lehr. *Bibliography.*

The Measurement of Strain. J. H. Lamble and S. S. Gill (*Trans. Liverpool Eng. Soc.*, 1947, **68**, 3–24; discussion, 25–37).—The various methods of measuring strain—mech., elect., and acoustic methods and by the use of brittle lacquers —are summarized. The main advantages and disadvantages of each method are given, and the principal requirements of any extensometer are briefly considered. In the discussion, — Meyer describes briefly an appn. of an elect. resistance strain-gauge, together with a 3-tube cathode-ray oscillograph or an Al alloy Isolator arm of a high-voltage airblast circuit-breaker. 43 ref.—S. R. W.

Non-Destructive Testing of Engineering Materials and Parts. R. C. McMaster and S. A. Wenk (*Materials and Methods*, 1951, **33**, (2), 81–96).—Materials and Methods Manual No. 67. A comprehensive review covering radiography, including xeroradiography and fluoroscopy, ultrasonic testing, magnetic-particle testing, penetrant methods, electromagnetic induction, and the recently developed triboelect. methods.—J. W. C.

Non-Contacting Thickness Gauge Using Beta Rays. C. W. Clapp and S. Bernstein (*Elect. Eng.*, 1950, **69**, (4), 308–310; *Elect. Eng. Abs.*, 1950, **53**, 333).—The general principles are given of the instrument, which utilizes Sr^{90} as source and an ionization chamber as detector, with a window of Al 0·001 in. thick. The source has a half-life of 25 years and there is no γ-emission, thus necessitating min. shielding for protective purposes. A motor-driven chopper is used in the beam to produce a 90-c./s. current which can be amplified by a simple A.C. amplifier. A second source of 90 c./s. is used as a ref. comparison and the resultant converted to D.C. by a metal-rectifier circuit. Appn. is to various metallic and non-metallic materials, which may be moving.

Metal Wall-Thickness Measurement from One Side by the Ultrasonic Method. N. G. Branson (*Elect. Eng.*, 1951, **70**, (7), 619–623).—The principles of ultrasonic detn. of metal thickness are discussed, and the construction of requisite apparatus is described and illustrated. Flaw detection by ultrasonics is also dealt with.—J. R.

*Detection of Laminations in Plates by the Ultrasonic Method. L. Kozłowski and M. Kurek (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, **1**, (2), 99–103).— [In Polish]. The detection of laminations in a non-ageing boiler plate (C 0·12, Si 0·19, Mn 0·46, Al 0·07%) by means of a Hughes Supersonic Flaw Detector, type MK-III, is described. The presence of a lamination is revealed by a drastic reduction of amplitude of the cathode-ray oscillogram. Results were confirmed by subsequent deep etching of the sectioned plate. 7 ref.—N. H. P.

Ultrasonics and Its Technical Application [in Metallurgy]. Heinz Thiede (*Z. Erzberg. u. Metallhüttenwesen*, 1951, **4**, (6), 212–221).—After discussing the prodn. of ultrasonic waves by magnetostrictive and piezo-elect. transducers, T. briefly reviews their tech. appn., including non-destructive testing, the coagulation of aerosols, and in the melting and casting of metals and alloys.—E. N.

Method of Detecting Cracks and Defects on the Surface of Castings by Fluorescence. —— (*Fonderie*, 1951, (65), 2539–2540).—The cracks are impregnated for up to 15 min. with the fluorescent soln. washed in warm water and air dried. The soln. is driven into the cracks by the appn. of a very finely divided powder and the surface examined under a Hg lamp.—J. H. W.

14 — TEMPERATURE MEASUREMENT AND CONTROL

*High-Temperature Thermal Analysis Using the Tungsten/ Molybdenum Thermocouple. [Determination of Freezing Point of Chromium]. H. T. Greenaway, S. T. M. Johnstone, and Marion K. McQuillan (*J. Inst. Metals*, 1950–51, 80, (3), 109–114).—A technique which enables thermal analysis to be carried out at temp. up to 2000° C. has been developed. H.F. induction heating is used, and the temp. is measured by means of a W/Mo thermocouple, for which a calibration curve is given. The method has been applied to the detn. of the f.p. of pure Cr, which is shown to be 1845° ± 10° C., and the liquidus curve for the solid soln. of Mn in Cr.

—Authors.

A Simplified Method for the Calibration of Thermoelements. G. Meyer and J. W. Franken (*Rec. Trav. Chim.*, 1950, 69, (3), 515–519).—[In English]. A modification of the potentiometric measurement of thermocouple e.m.f. is described, the standard cell being replaced by a Pt/Pt–10% Rh thermocouple placed in the vapour of boiling water. Measured e.m.f. are referred to this arbitrarily chosen standard, after appn. of the correction for atmospheric pressure.—J. W. C.

Forging Ahead at Utica. [Temperature-Recording Equipment]. F. N. Darmara and Warren Walker, Jr. (*Indust. Heating*, 1950, 17, (11), 1920–1922, 1924, 1926).—An illustrated account of the temp.-recording and indicating equipment installed at the works of the Utica Drop Forge and Tool Corp. for the heat-treatment of forged turbine blades for aircraft jet engines. The actual forging and heat-treatment processes are outlined.—D. M. L.

Sensitive Temperature Controller. —— (*Engineering*, 1951, 172, (4464), 203–204).—Describes a new temp. controller which operates on the differential expansion of a quartz rod and a heat-resisting steel rod.—D. K. W.

15 — FOUNDRY PRACTICE AND APPLIANCES

Water-Cooled Engine Block Gravity Die-Cast [in Aluminium Alloy]. —— (*Steel*, 1950, 127, (6), 86, 87, 108).—The prodn. procedure of a foundry making a complicated permanent-mould casting in Al alloy, is described.—E. J.

Close-Tolerance [Aluminium-Alloy] Die-Castings Cut Auto Transmission Cost. —— (*Steel*, 1950, 127, (22), 96–97).— A number of close-tolerance, intricate Al-alloy die-castings for automobile prodn. are described and illustrated.—E. J.

Aluminium-Alloy Gravity Die-Castings as Affecting the Production Engineer. H. C. Cross (*J. Inst. Prod. Eng.*, 1949, 28, (2), 62–81).—A general review covering practical manufacture; comparison with sand castings and pressure die-castings; selection of material; design; prodn.; inspection; and finishing.—M. A. H.

A Die for a Clack-Valve Body in Aluminium Bronze. —— (*Fonderie*, 1951, (64), 2447–2449).—An Al bronze clack valve chill-cast in a vertical mould contained severe piping and numerous pinholes, caused by intense local heating. A redesigned cast-Fe mould used horizontally resulted in a satisfactory casting that stood up to a pressure of 10 kg./cm.².

—J. H. W.

Casting Phosphor-Bronze Crowns on Cast Iron Wheels. —— (*Fonderie*, 1951, (67), 2582–2583).—Describes the cleaning, pickling, ZnCl₂ immersion, melting, and casting of P-bronze or 88/12 crowns on cast Fe wheels so as to ensure good adhesion of the bronze on the cast Fe.—J. H. W.

The Casting of Inconel-Tubing Heaters in High-Copper Alloy. F. J. Lambert (*U.S. Atomic Energy Commission Publ.*, 1950, (Y-583), 25 pp.).—The short life of Inconel-tube heaters embedded in high-Cu alloy heaters was investigated and was found to be due to voids in the castings which were most marked around the Inconel tubes. The report deals with experiments designed to improve the quality of the castings, and recommendations are made for tube-surface prepn. and casting techniques.—B. W. M.

Magnesium Castings Aid Britain's Industrial Recovery. G. B. Partridge (*J. Inst. Prod. Eng.*, 1949, 28, (7), 334–349).— A review of Mg extraction, casting alloys, founding, inspection and defects, heat-treatment, and properties.—M. A. H.

How to Handle Magnesium-Zirconium Casting Alloys. L. J. G. van Ewijk (*Modern Metals*, 1951, 7, (3), 48–49).— The technique employed in a Dutch foundry for handling Mg–Zr alloys is described.—R. J.

20 mm.-Fuse Noses Zinc Die-Cast at High Speed. —— (*Steel*, 1951, 129, (7), 78, 79).—Using a 12-cavity die, fast plunger-type machines cast 2700 components/hr. in Zn alloy.

—E. J.

The Theory of Degassing of Molten Metals. Joseph A. Verö (*Acta Techn. Acad. Sci. Hungar.*, 1950, 1, (1), 130–155).— [In Russian]. Sievert's law of gas solubility and the available experimental data on the solubility of gases in pure metals were used to develop quant. relations between the compn., pressure, and temp. of the furnace atmosphere and the amount of dissolved gas in a metal. It is shown that these relations could be extended also to deal with four principal methods of degassing: presolidification, standing, bubbling a neutral gas, and appn. of vacuum.—V. K.

Continuous Casting of Non-Ferrous Metals. M. Schneider and E. Zalesiński (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1950, 2, (1), 49–70).—[In Polish]. Continuous casting methods for non-ferrous metals are reviewed, with special ref. to the Junghans, V. L. W., and Asarco processes, and the advantages and disadvantages of each are discussed. The possibility of applying continuous casting to steel is considered. Theoretical calculations on the solidification of ingots have been made by means of formulæ given by Roth (*Aluminium*, 1943, 25, 283; *Met. Abs.*, 1947, 14, 273), and the effects of continuous casting on the microstructure and mech. properties of the ingot are described.—N. B. V.

Centrifugal Casting for Critical Components. Herbert Cooper (*Machine Design*, 1950, 22, (11), 131–132).—The advantages of centrifugal casting as a prodn. technique for ring-form components are indicated, and some typical aircraft-engine castings are used as illustrations. The text refers mainly to castings in stainless steels.—D. M. L.

How to Select Non-Ferrous Alloys for Investment Castings. Rawson L. Wood and Davidlee Von Ludwig (*Materials and Methods*, 1951, 33, (1), 78–82).—The factors which produce good castability are reviewed, and particular ref. is made to Al, Mg, and Cu alloys.—J. W. C.

The Fields of Utility of Investment Castings. R. L. Wood and D. Von Ludwig (*Mech. Eng.*, 1951, 73, (3), 191–197).—W. and ¼Von L. group investment castings under four categories: (1) castings which cannot be produced by any other process, e.g. a turbine bucket made from a heat-resistant alloy which cannot be forged economically; (2) castings previously made by other processes; simplification in design, increase in rigidity, reduction in weight, ability to choose a rare suitable material, as well as reduction in cost, led to the choice of the investment process; (3) castings which served as trial material during design evolution; such parts were ultimately mass-produced as forgings or die-castings, the investment castings being used to run performance tests on the new design without incurring the excessive expense and risk of the tools and dies needed to make similar tests on forged or die-cast parts; and (4) castings being investment cast but which represent an uneconomic appn. of the process owing to incorrect design or material selection.—H. Pl.

Mercury Patterns for Precision Castings. [Mercast Process]. —— (*Machinery* (Lond.), 1951, 78, (1999), 396–397).— Cf. *Met. Abs.*, this vol., col. 61. Photographs are given of

the operations involved in producing and investing patterns of frozen Hg for precision castings.—J. C. C.

Ejecting Mechanism for Die-Casting Die. C. Bossmann (*Machinery* (*Lond.*), 1951, **78**, (1998), 354*l*).—A method of providing a short injection stroke relative to the travel of the die is described.—J. C. C.

Cleaning Die-Casting Dies by the Liquid Blast Process. —— (*Machinery* (*Lond.*), 1951, **78**, (1996), 283–285).—Periodic cleaning of dies by the " Hydro-Finish " process maintains a high standard of surface finish in the castings.
—J. C. C.

Inspection and Storage of Foundry Lifting Equipment. W. M. Halliday (*Found. Trade J.*, 1950, **89**, (1772), 191–192).—A general review.—J. E. G.

Foundry Sand : Getting Down to Fundamentals. Clifford E. Wenninger (*Amer. Foundryman*, 1951, **19**, (1), 45–48).—The main feature of the available data on properties of foundry sands is a high degree of empiricism, and consequently, a limited amount only of the fundamental knowledge. The arguments are put forward for changing the present situation and supplanting the empirical by fundamental studies.
—V. K.

Grey Iron Shrinkage Related to Moulding-Sand Conditions. Clyde A. Sanders and Charles C. Sigerfoos (*Amer. Foundryman*, 1951, **19**, (2), 49–55).—Cylindrical specimens of grey Fe were cast vertically, sectioned, and examined for the extent of top surface shrinkage (piping). The tendency to piping is increased by fire clay bond, western bentonite, moisture, and low ramming *d*, while it is decreased by sea coal, wood flour, oil bonds, and high ramming *d*. 34 ref.—V. K.

Infra-Red Oven Used for Core Drying in Brass Foundry. —— (*Indust. Heating*, 1950, **17**, (10), 1811–1812).—A brief illustrated account of a small 18-kW. infra-red drying oven for drying sand cores after wash-dipping. The cores are completely dry approx. 2 min. after dipping. The tunnel is 6 ft. long, and the cores are conveyed through the heater at 3 ft./min.—D. M. L.

Core-Oil Evaluation Method. A. E. Murton, H. H. Fairfield, and B. Richardson (*Amer. Foundryman*, 1951, **19**, (4), 85–89).—The procedure for comparative testing of core sand bonds and an improved elect. resistance core-drying oven are described.—V. K.

Statistical Quality Control Manages Foundry Operations. Robert J. Feltrin (*Amer. Foundryman*, 1951, **19**, (1), 40–42).—Examples of appn. of statistical analyses to foundry prodn. problems are given.—V. K.

Production and Methods Control : Application to Foundry and Patternshop. T. H. Wood (*Found. Trade J.*, 1950, **89**, (1790), 529–533).—Paper presented to the South African Branch of the Institute of British Foundrymen. A general account of the appn. of prodn. control to foundry and patternshop.—J. E. G.

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17 — FURNACES, FUELS, AND REFRACTORIES

G.W.B. Twin-Chamber Coil-Annealing Furnace [with Forced-Air Circulation] for Aluminium Alloys. —— (*Machinery* (*Lond.*), 1951, **78**, (1993), 144).—J. C. C.

Copper Tubes Spot-Annealed in Deoxidizing Atmosphere. W. W. Swarthout (*Steel*, 1951, **128**, (11), 84).—Equipment for the annealing of Cu tubes is described.—E. J.

[Copper] Wire-Annealing Furnace Fits Low Headroom. W. J. Krailing (*Iron Age*, 1951, **168**, (2), 93–95).—K. describes the special furnace construction and handling equipment for wire- and rod-drawing a wide range of Cu-base alloys in a factory with ample floor space but only 12 ft. head-room.—J. H. W.

Electric Furnace Melting of Copper at Baltimore. Peter R. Drummond (*J. Metals*, 1951, **3**, (6), 468–472).—A detailed description is given of the construction and operation of a heavy-duty, 3-phase, 50-ton capacity, arc furnace—rated at 20 tons/hr.—and its auxiliary, 5-ton capacity, L.F. induction holding furnace, for the final melting of refined Cu. The ingots produced contain ~0·01% O, and are without con-tamination from fuel.—E. N.

Arc Melting of Titanium Metal [and the Mechanical Properties of the Resulting Ingots]. S. F. Radtke, R. M. Scriver, and J. A. Snyder (*J. Metals*, 1951, **3**, (8), 620–624).—The construction and successful operation of an automatic, continuous-casting arc furnace—of the von Bolton type—is described. It has a non-consumable water-cooled, rotating cathode, a D.C. arc, and an inert atmosphere of A or He. While designed sp. for melting commercially pure iodide Ti sponge—which it does at a rate of 0·6–1·0 lb./min., with a power input of 64–100 kW.—and casting into ingots 5¾ in. dia. and up to 60 in. long, it can be employed for melting other reactive metals. A comparison of the properties of arc- and induction-method Ti indicates that where high ductility, formability, and toughness are required, arc melted metal is preferable. 14 ref.—E. N.

The Choice of the Proper Refractory for the Casting of High-Melting Electropositive Metals [Titanium]. L. Brewer (*U.S. Atomic Energy Commission Publ.*, 1950, (UCRL-762), 20 pp.).—B. discusses the thermodynamic requirements for suitable refractory materials, with special ref. to Ti. Tables are included giving the heats of formation of various halides and sulphides, which show that Ti is relatively a more power-ful reducing agent in fluoride and oxide systems than in those with iodides or sulphides. B. concludes that in com-pounds in which the anion is large compared with the cation, an increase in stability may be expected in going down the Periodic Table or in moving to the left. When the anion is only slightly larger than the cation, a reversal may be expected. It is recommended that tests on the suitability of various materials are best carried out by heating intimate mixtures of metal powder and powdered container rather than heating the metal in a fabricated crucible. The most promising refractories for Ti appear to be Mo_3Si, $Ta_5Si_{1\pm0·2}$, TaB, ZrB_4, UB_2, CeB_4, CeS, TaS, and $TaC_{0·8-1·0}$.
—B. W. M.

Vacuum Melting Titanium and Zirconium. A. C. Haskell, Jr. (*Steel*, 1951, **128**, (25), 77–79).—An abridged version of a paper given at a symposium sponsored by the Niagara Falls Section of the Electrochemical Society. A description is given of melting equipment used for producing ingots up to 25 lb. in Ti and up to 40 lb. in Zr. Contamination is discussed.—E. J.

[Discussion on a Paper by F. C. Evans on :] Modern Melting Practices. —— (*Found. Trade J.*, 1951, **90**, (1798), 177–179).—Cf. *Met. Abs.*, 1949–50, **17**, 763).—J. E. G.

The Operation of Small Salt Baths. S. W. Hugo (*Machinery* (*Lond.*), 1951, **79**, (2020), 198–201).—Operating notes are given on the starting-up and emptying of small immersed-electrode salt baths.—J. C. C.

Heating by High-Frequency Fields. I.—Induction Heating. E. C. Witsenburg (*Philips Tech. Rev.*, 1949, **11**, (6), 165–175).—W. outlines the principles underlying the use of H.F. elect. currents in heating both conducting and non-conducting bodies. A formula for the efficiency of a work coil used in the induction heating of a simple cylinder is evaluated, and from this is deduced the most efficient frequency to employ for a cylindrical specimen of given size. Valve generators, and some typical appn. for this type of heating (surface-hardening, annealing, brazing, soldering, and melting) are discussed.—D. K. W.

Heating by Means of High-Frequency Fields. II.—Capacitive Heating. M. Stel and E. C. Witsenburg (*Philips Tech. Rev.*, 1949, **11**, (8), 232–240).—Cf. *ibid.*, (6), 165 ; preceding abstract. The theoretical aspects of capacitive heating are discussed, and H.F. generators for this type of heating are considered. Important appn. are also dealt with.—D. K. W.

Heat-Application Fundamentals. M. H. Mawhinney (*Indust. Heating*, 1951, **18**, (6), 978–980, 982, 984, 986, 988, 990, 992, 994, 996).—M. outlines the gradual appn. of heat-transfer theory to furnace design and traces the development in furnace forms since 1914. The provision of formulæ for furnace design is discussed, and simplified formulæ are given allowing the computation of heat loss, fuel requirements, and the coeff. of convection. Heat balance in furnaces is discussed in some detail. The subjects of furnace capacity and burner capacity are treated in considerable detail.—D. M. L.

Cooling Rates in Modern Furnaces. M. H. Mawhinney (*Indust. Heating*, 1950, **17**, (10), 1708–1710, 1712, 1714, 1870–1872, 1874, 1876, 1878).—M. points out that very little information exists on the cooling rates of indust. furnaces and stresses that math. calculation of cooling rates is very difficult. The cooling rates depend on the size and shape of the furnace, the type of refractory lining, the charge in the furnace, and the open area of flues or furnace vents. Batch- and continuous-type furnaces are separately treated, and data are tabulated for the different furnaces to allow the effects of the various factors to be analysed. For batch-type furnaces the general conclusions drawn are that furnace shape is of great importance; that the presence of a charge lowers the initial rate of cooling, but has little effect on the time required to cool a furnace to 10% of its original temp. and that the entry of air into the furnace has a great effect on the cooling rate. For continuous furnaces, M. outlines a calculation designed to give figures for the length of cooling zone required to cool the charge to a desired temp. at various prodn. rates.—D. M. L.

Preventing Fuel Explosions in Gas- and Oil-Fired Industrial Furnaces. I.—Gas-Fired Furnaces. II.—Oil-Fired Furnaces. J. B. Smith (*Indust. Heating*, 1950, **17**, (10), 1729, 1730, 1732, 1734, 1840–1842, 1844; (11), 1950, 1952, 1954, 1964).—[I.—] The danger to life and property due to fuel explosions are stressed, and details are given of methods for reducing the fuel explosion hazard in indust. furnaces. It is stressed that well-trained " burner men " are essential to supplement the protection given by automatic safety controls. Recommendations are made for the installation of the automatic controls and for procedures for lighting and maintaining the furnaces. The establishment of maintenance and testing procedures is essential. [II.—] S. describes the causes of explosions in oil-fired furnaces and outlines suitable preventive measures. Details are given of a number of recommended safety practices applicable to direct internal-fired types of furnaces such as muffle multi-burner types and to direct external-fired recirculation furnaces. The protection of combustion–air piping against explosions is discussed.
—D. M. L.

The Preparation of Alumina Bodies by the Slip-Casting Method. W. Watt and J. P. Roberts (*Metallurgia*, 1951, **43**, (260), 307–308).—Micronized Al_2O_3 is ball-milled to obtain the desired particle size and then leached with HCl. After settling, the supernatant liquor is decanted and the Al_2O_3 washed to a pH of 2·5–3 and sufficient water then left to give a slip contg. 70–78% Al_2O_3. After putting under a good water pump for 20–30 min. to remove air bubbles, the slip is cast into a plaster-of-Paris mould and the casting allowed to stand for 24 hr. It is then removed and fired at 1000° C., after which it may be machined with WC tools; final firing is at 1720°–1820° C. after which machining may be done with diamond tools.—F. M. L.

The Production of Sintered Alumina Ware. T. G. Carruthers (*Metallurgia*, 1951, **44**, (262), 103–107).—An account is given of the techniques used at Leeds University to produce Al_2O_3 articles. Details are given of the grinding of Al_2O_3, prepn. of slip and moulds and casting of the slip for producing slip-cast articles and prepn. of materials and dies for producing articles by pressing. The method of firing is fully described.
—F. M. L.

New Zirconia Refractory Material Useful at Temperatures to 4600° F. (2550° C.). —— (*Materials and Methods*, 1951, **33**, (3), 81).—Appn. of fused stabilized zirconia, now being produced commercially in a wide variety of shapes, are summarized.—J. W. C.

Heat-Resistant Ceramic Coatings Permit Substitutes for Critical Materials. Dwight G. Bennett (*Materials and Methods*, 1951, **33**, (3), 65–67).—A review of some of the uses of ceramic coatings on materials for service at high temp.
—J. W. C.

High-Temperature Refractories. J. F. Stirling (*Pract. Eng.*, 1951, **23**, (588), 596–597).—A general review is given of special refractories for precious metals, &c.—E. J.

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19 — WORKING

Electric Equipment for Hot-Mill Line at Rogerstone Works of the Northern Aluminium Co., Ltd. D. Bolton (*B.T.H. Activities*, 1951, **22**, (3), 80–87).—Cf. *Met. Abs.*, 1950–51, **18**, 399. Describes the working and layout of the mill, with particular ref. to power supplies, mill drives, &c.
—D. K. W.

Electric Equipment for High-Speed Cold Strip-Rolling Mills at Rogerstone Works of the Northern Aluminium Co., Ltd. A. S. Wallis (*B.T.H. Activities*, 1951, **22**, (4), 99–105).—Cf. *Met. Abs.*, 1950–51, **18**, 399. A comprehensive account of the mill drives, auxiliary elect. equipment, &c.—D. K. W.

*The Extrusion of Beryllium. A. R. Kaufmann, *et al.* (*U.S. Atomic Energy Commission Publ.*, 1950, (AECD-2883), 29 pp.).—A method for the extrusion of Be into rods and tubing is described. The extrusion temp. is from 1500° to 1900° F. (815°–1040° C.) and the billet is clad in soft Fe to prevent contact with the dies. The extruded material is stronger and more ductile than the cast metal, and this is considered to be due to the grain refinement and to the preferred orientation which is shown by X-ray studies. A rod reduced by 98·4% in area gave a U.T.S. of 53,000 lb./in.² and an elongation of 0·9%. The U.T.S. of extruded flat stock was three times as great in a direction ‖ the extrusion as that in a perpendicular direction.—B. W. M.

How Revere Checks Machinability [of Copper Alloys]. Harry Schagrin (*Steel*, 1950, **126**, (25), 84–86).—The detn. is described of the machinability of various Cu alloys by a test which measures and interprets minute movements of the cutting tool.—E. J.

*Hot-Forging Tests on Some Copper-Base Alloys. A. L. Simmons (*Australian J. Appl. Sci.*, 1951, **2**, (1), 76–88).—Hot-forging tests were used to determine whether the behaviour of α-brasses could be predicted in hot rolling. Apart from the standard nickel silvers, cartridge and 65 : 35 brass are of equivalent plasticity at the resp. rolling temp. of 870° and 780° C.; 10% low-Zn nickel silver is a little more difficult, and 70 : 30 cupro-nickel is still more difficult to deform. Cu is the most plastic. This agrees with large-scale practice, but the test does not indicate the likelihood of hot shortness in the alloys tested.—S. M.

Pratt and Whitney Draws Tough Alloys to Close Limits. Robert M. Treat and Herbert Chase (*Steel*, 1950, **127**, (17), 68–70).—Discusses the manufacture by hydraulic presses of aircraft parts in alloys such as a 80 : 20 Ni–Cr.—E. J.

Metal Spinning : Now a Mass-Production Tool. Dan Reebel (*Steel*, 1951, **128**, (23), 88–91, 114).—The advantages of spinning are enumerated, and, in a table, factors are given which indicate the adaptability to spinning of a wide range of metals and alloys. Examples of various spinning operations are described.—E. J.

Work and Power in the Rolling Process. Z. Wusatowski (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1950, **2**, (3), 213–267).—[In Polish]. A comprehensive review of the literature relating to the evaluation of roll torque and power consumption in hot and cold rolling. About nine-tenths of the paper summarizes English, American, German, &c., work, as at 1948, by Ekelund, Underwood, Cook and Larke, Ford, Siebel, Emicke, and others. Several diagrams

of power consumption/rolled ton are quoted from books by Tselikov. These refer to heavy steel slabs (16·6 tons), 6·6-ton blooms, rails, angles, rounds, &c., and one of them to the hot rolling of 5-5½-in. Duralumin slabs on a 4-high mill with 28-in. work rolls. 51 ref.—N. H. P.

*The Effects of Speed, Temperature, and Kind of Rolls on Spreading and Elongation in Hot Rolling. Z. Wusatowski and R. Wusatowski (*Prace Badawcze Głównego Inst. Met i Odlewnictwa*, 1950, 2, (2), 111–122).—[In Polish]. Formulæ for calculating lateral spread and elongation in the hot rolling of steel are given as $\beta = acdf\gamma^{-W}$, and $\lambda = 1/(acdf) \times \gamma^{(W-1)}$, in which $W = 10^{-1.269\delta} \times \varepsilon_W{}^{0.566}$, $\beta = b_2/b_1$, $\lambda = l_2/l_1$, $\gamma = h_2/h_1$, $\delta = b_1/h_1$, $\varepsilon_W = h_1/D$; h_1, h_2, b_1, b_2, l_1, l_2 = thickness, width, and length before (1), and after (2) the pass; D = roll dia., a = coeff. of temp. = 1 above 950° C. and 1·005 for 840°–950° C.; c = coeff., depending on rolling speed, d = coeff. depending on steel quality, f = coeff. depending on roll material and finish = 0·997–1·025. The author's experiments were confined to rolling steel billets 55 × 55, 30 × 30, 30 × 15, and 20 × 20 mm. in 700, 370, 300, 280, and 260 mm. rolls at speeds of 1·5–16·5 m./sec. 12 ref.—N. H. P.

Calculation of Roll Pressure in the Rolling Process. Z. Wusatowski (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, 1, (3), 233–290).—[In Polish]. A review, based substantially on Underwood's and Ford's articles (*Sheet Metal Ind.*, 1945–1949), which discusses theoretical and semi-empirical methods of roll-pressure calculation in hot and cold rolling, the effects of roll flattening and tension in strip rolling, friction, work-hardening, rolling temp., and rate of deformation. Values for power consumption in the hot rolling of steel joists, channels, and angles on 3-high mills are quoted from Filasiewicz ("Walcownictwo," Gliwice: 1947). 176 ref.—N. H. P.

New Methods for Control of Thickness in Rolling Strip Metal. W. C. F. Hessenberg and R. B. Sims (*Research*, 1951, 4, (2), 92–93).—A letter. Tests made on an experimental rolling mill show that the relation between the roll-separating force exerted by the metal being rolled and the smallest distance between the rolls is practically linear. The mill can be regarded as a stiff spring of formula $F = M(S - S_0)$, where F is the roll-separating force, S the smallest distance between rolls when the force is acting, and S_0 is the "roll setting," which is the value of S when $F = 0$; M is a const. S is nearly equal to h, the thickness of the rolled strip; the above equation is then $F = M(h - S_0)$, which when differentiated becomes $\left(\dfrac{\delta F}{\delta S_0}\right) \cdot \dfrac{1}{h} = -M$. These equations provide the bases for two new automatic thickness regulators. In the first, since changes in h result from changes in F, F is held const. by applying more or less tension to the strip either before or after passing through the rolls. In the second, h is controlled by adjusting the roll setting so that $\delta S_0 = -(1/M) \cdot \delta F$. Details of the working of the thickness regulators are to be published.—F. A. F.

Hot Machining of Many Metals Improved by Arc Heating. E. T. Armstrong and A. S. Cosler (*Materials and Methods*, 1951, 33, (1), 69–73).—Preliminary tests indicated that many high-temp. alloys (e.g. Vitallium) have much better machinability at high temp. Arc heating was found best for controlled heating of the surface being machined. The arc is struck between the work and a C electrode and a H.F. high-voltage spark is superimposed to prevent the tendency of cold metal to quench the arc. The position of the arc is stabilized by a magnetic field. The importance of selection of correct power d for a particular alloy is emphasized; modification of the microstructure is then negligible.—J. W. C.

How Tool Life is Affected by Heat-Resisting Alloys. —— (*Automotive Ind.*, 1951, 104, (7), 52, 80, 83).—Machining alloys such as Inconel "X" or S-816, used in the manufacture of gas-turbine buckets and discs, presents great difficulties because they: (1) exhibit high shear strength during cutting, (2) work-harden during cutting, and (3) are highly abrasive when in contact with conventional sintered-carbide tools. They should be machined, where possible, in the soln.-treated condition, and carbide tools of the cast-Fe grade should be employed. A graph is given showing the relationship between the tool life and cutting speed for several alloys.—Z. M.

History and Development of Wire-Drawing Machinery. R. S. Brown (*Trans. Manchester Assoc. Eng.*, 1947–48, 245–252; discussion, 252–259).—A brief survey of the development of wire-drawing machines, and the wire-drawing die. Mention is made of characteristic changes in phys. properties of the wire during the process.—S. R. W.

On the Correct Shape of Wire-Drawing Dies. M. Schneider (*Prace Badawcze Głównego Inst. Met. i Odlewnictwa*, 1949, 1, (3), 213–231).—[In Polish]. A review of the wire-drawing theories of Lewis, Sachs, Siebel, Lunt and MacLellan, and others. A Russian theory by Sobolev and Sokolov (*Stal*, 1947, (8), 734) is discussed in detail. Their formula is: $\sigma_z/k = 1 - \exp[-\gamma q/(1 - q)]$, in which $q = (A_0 - A)/A_0$ and $\gamma = (\mu + \tan \alpha)/[(1 - \mu \tan \alpha)\tan \alpha]$, where σ_z = drawing stress, k = yield stress; A_0 and A = initial and final sectional area; μ = coeff. of friction, and α = die semi-angle. For $\alpha < 30°$ and light drafts, $\gamma \simeq 1 + \mu \tan \alpha + \mu \cot \alpha$, and with further simplifications $\sigma_z/k = q/(q - 1) \times (1 + \mu \tan \alpha + \mu \cot \alpha)$. This latter form is considered inaccurate by S. Formulæ are deduced for the ratio bearing length : reduced dia., expressed as a function of reduction of area and die angle. Nomograms simplifying the calculation are given. 13 ref.—N. H. P.

The Mass Production of Coins. J. H. Partridge (*J. Inst. Prod. Eng.*, 1951, 30, (3), 94–103).—A description of some of the processes involved in the manufacture of coins in the Calcutta Mint is given.—M. A. H.

[Manufacture of Ammunition at] The Royal Ordnance Factory, Radway Green. —— (*Engineer*, 1951, 191, (4975), 732–733).—A brief description of the manufacture of 0·303-in. Mk. VII and 20-mm. cannon ammunition.—D. K. W.

20 — CLEANING AND FINISHING

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

Processing of "Land Rover" [Aluminium] Components. —— (*Indust. Finishing* (*Lond.*), 1951, 3, (36), 907–908, 911; also *Product Finishing* (*Lond.*), 1951, 4, (6), 54, 55, 106).—A short description of the Spra-Bonderite "170" plant used for the pretreatment of Al motor-car sections.—H. A. H.

Pickling of Aluminium Bronze. —— (*Fonderie*, 1951, (65), 2540).—The compn. of the pickling bath, temp., time, and resulting appearance in the chem. pickling of Al bronze are tabulated.—J. H. W.

Rinsability and Buffering Action of Alkaline Cleaners. —— (*Metal Finishing*, 1951, 49, (6), 62–63).—Recent proposals by the American Society for Testing Materials in connection with methods of testing the rinsing of cleaners,

a test for buffering action, and a series of definitions relating to metal cleaning, are summarized and comments invited.
—G. T. C.

Preparing Zinc-Base and Aluminium-Base Die-Castings for Finishing. D. F. Seymour (*Materials and Methods*, 1951, 33, (3), 68–69).—Before the appn. of a painted, enamelled, or electroplated surface finish to a die-casting, it is necessary to clean and condition the surface to improve its corrosion-resistance and appearance. The methods used are outlined.
—J. W. C.

Improved Finishing Machine for Zinc-Base Die-Castings. —— (*Automotive Ind.*, 1951, 104, (9), 58).—The machine is briefly described and illustrated. Rotating tubs charged

with an abrasive mixture perform the polishing of the parts mounted on a special fixture. The complete cycle lasts <2 min. and is entirely automatic, except for loading and unloading.—Z. M.

A Practical Survey of Modern Decorative Finishes : Wrinkle, Crystalline, Metallic, Polychromatic, and Other Aluminium Finishes.—I. H. J. Testro (*Brit. Indust. Finishing*, 1950, 2, (21), 573–574, 576, 579–580, 582).—A short survey of some types of organic finish, including those obtained by incorporating Au, Cu, and Al powders in a varnish or lacquer vehicle.—H. A. H.

Watch Your Surface Finish.—I.–II. —— (*Steel*, 1950, 126, (25), 88–90, 115–116 ; (26), 78, 79, 90).—Various types of surface finish, its effect on the life of a component, and methods of measuring surface finish are described.—E. J.

Uniform High-Quality Finish Obtained with Diamond-Powder Abrasives. S. G. Kelley, Jr. (*Materials and Methods*, 1951, 33, (2), 65–67).—Methods of using graded diamond powder for finishing very hard materials are described. The importance of high accuracy in grading is emphasized, and a table of the recently modified U.S. Bureau of Standards standard grades, with their most important uses, is included. —J. W. C.

Finishing Armament Equipment : Some World War II Problems. W. F. Coxon (*Indust. Finishing* (*Lond.*), 1951, 3, (35), 802, 805–806, 809).—H. A. H.

21 — JOINING

Metal-Bonding : A Review of Development Work on Metal Adhesive Processes by the Bristol Aeroplane Co., Ltd. F. H. Parker (*Aircraft Prodn.*, 1951, 13, (150), 107–114).—P. discusses the processes and manufacturing technique developed by the Bristol Aeroplane Co., Ltd., in the use of adhesive in metal airframe construction, with particular ref. to the Redux process.—W. A. M. P.

The Bonding of Metals. C. J. Moss (*Metallurgia*, 1951, 43, (260), 267–272).—The use of synthetic-resin adhesives in bonding metals is discussed. Fields of appn., limitations, and the design and properties of bonded joints are considered. —F. M. L.

*Strength of [Adhesive-Bonded] Aluminium Lap Joints at Elevated Temperatures. Edward W. Kuenzi (*Forest Products Lab. Rep.*, 1949, (1808), 6 pp. ; *Appl. Mechanics Rev.*, 1951, 4, 157).—Tests were carried out at 6 temp. from 75° to 600° F. (25°–315° C.) of lap-joint specimens with an overlap of $\frac{1}{2}$ in., made of 0·064-in.-thick clad 24S-T3 Al alloy suitably etched and bonded with 9 different adhesives, 6 of them commercially available and 3 laboratory formulated. The strength was invariably found to fall with increasing temp. ; highest strengths at higher temp. were obtained with a hot-setting phenol-resin–synthetic-rubber adhesive (type M) ; the highest strengths (4500 lb./in.^{2}) at room temp. were obtained with hot-setting modified phenolpolyvinyl butyral adhesive. Failure was assessed by the percentage of glued area which failed in adhesion (i.e. where the adhesive separated from the metal surface), as distinct from area which failed in cohesion of the adhesive. Similarity in decrease of strength with rise of temp. in lap joints and sandwich constructions of the aircraft type (*ibid.*, 1804) leads to the conclusion that the former may be used to predict the latter.

*Evaluation of Several Adhesives and Processes for Bonding Sandwich Constructions of Aluminium Facings on Paper Honeycomb Core. H. W. Eickner ([*U.S.*] *Nat. Advis. Cttee. Aeronautics, Tech. Note*, 1950, (2106), 23 pp.).—In preliminary tests, 6 of 14 bonding processes evaluated by tension tests on sandwich specimens of 1-in. Al cubes bonded to a resin-impregnated paper honeycomb core gave good-quality bonds which had average strengths >350 lb./in.^{2} or >75% failure in core when tested at both 80° and 200° F. (27° and 95° C.). The four bonding processes that had given the best results in preliminary tests were used to determine the effect of the amount of adhesive spread on the tensile strength of joints made between Al and resin-impregnated paper honeycomb core, and found capable of producing good joints between Al and paper honeycomb when moderately heavy spreads were applied to both core and facings. The use of lighter spreads or appn. of adhesive to only the core or facing usually resulted in lower-quality bonds, but with some of the adhesives the strength of joints was sufficiently high for lighter spreads to be considered satisfactory.—AUTHOR.

Rubber-to-Metal Bonding : Ingenious Adaptation of the Redux-Bonding Process. —— (*Aircraft Prodn.*, 1951, 13, (151), 131–132).—A short note describing a modification of the Redux process for bonding rubber to metal at pressures as low as 10 lb./in.^{2}.—W. A. M. P.

Metal-Stitching. C. A. H. Pollitt (*Aircraft Prodn.*, 1951, 13, (148), 62–64).—A metal-stitching machine is described which forms stitches from a continuous length of steel wire, the wire punching holes as the stitches are driven. The wire used is 0·048-in.-dia., 320,000-lb./in.^{2} C steel wire, covered with 0·0015-in. hot-dip Zn coating, which does not scrape off when piercing Al. The possibilities of the process, strength characteristics, and appn. are discussed, and metal-stitching is compared with spot-welding. —W. A. M. P.

Stitch It Instead. A. G. Denne (*Iron Age*, 1951, 168, (3), 97–100 ; (4), 78–82).—D. describes the equipment, process, and advantages of stitching metal to metal and to non-metals for Al, steel, brass, and Cu. Stitch penetration is limited by the shear resistance of the work.—J. H. W.

*Tests in the Production and Driving of Large Aluminium Alloy Rivets. J. C. Bailey and A. W. Brace (*Aluminium Development Assoc. Research Rep.*, 1950, (8), 60 pp.).—Data have been obtained on the manufacture of large Al alloy rivets on normal cold-heading machines, and on the pressures required to close them with various shapes of driven point. Methods of closing them with hand-held pneumatic hammers, using existing equipment under practical conditions, have also been explored. Experiments were confined to rivets of $\frac{5}{8}$ and $\frac{7}{8}$ in. dia. made in alloys conforming to the British Standard Specifications (N4, N5, N6, H10, H13, H14). The investigations show that existing rivet-making machines are suitable for making satisfactory Al alloy rivets of the sizes mentioned, and also establish the influence of head shape and alloy compn. on behaviour during driving. Tables show the pressure required to close rivets with standard points and the max. sizes that can be closed with rivet hammers. The possibilities of hot driving are also briefly discussed in the light of preliminary experiments. Hole clearances should preferably be smaller than for steel. Two alternative modifications in the shape of the driven point are suggested as means of enabling rivets up to $\frac{7}{8}$ in. dia. to be readily closed with existing pneumatic hammers : (i) a pan shape smaller than usual, and (ii) a "recessed " rivet of simple design, capable of considerable adaptation in accordance with the strength required. Appendices give the method of determining pin shear values and procedures for compression tests.—AUTHORS.

*Structural Light-Alloy Riveting with the High-Tensile Alloys. J. G. Whitman (*Engineer*, 1951, 192, (4987), 228–231).—W. discusses the appn. of light-alloy riveting to structural members, and presents experimental data on the strengths of joints in which rivets are in tension and in shear. Cold-driven rivets are employed, owing to the critical conditions demanded by hot-driving techniques for light alloys. The method ensures complete filling of the rivet hole, but relatively poor clamping of the joint members between the rivet heads. W. states that this permits redesign of the normal snap and conical heads commonly used, with advantage, and further suggests that the high frictional forces between rivet shank and hole makes "headless " rivets feasible. Alloys with a low closing-load : shear-stress ratio

are most suitable for riveting, e.g. AW10B and AW15B alloys. In general, hard heat-treated rivets are preferable to soft ductile ones. W. concludes with an account of experiments on head design and rivet-closing techniques.
—D. K. W.

Characteristics and Applications of Standard Aluminium Rivet Alloys and Tempers. —— (*Materials and Methods*, 1951, 33, (6), 105).—Materials Engineering File Facts No. 209. Driving conditions and comments and assembly alloys for which it is suitable are given for each rivet alloy.
—J. W. C.

*Results of Shear Fatigue Tests of Joints with 3/16-in.-Diameter 24S-T31 Rivets in 0·064-in.-Thick Alclad Sheet. (Holt). See col. 221.

New Copper Paste Permits Brazing Economies. Kenneth Rose (*Materials and Methods*, 1951, 33, (2), 79–80).—A Cu paste, trade name " Cubond," has been developed for the furnace-brazing of steel parts and has considerable advantages over the use of metallic Cu foil or wire.—J. W. C.

Copper Furnace-Brazing's Place in Forming Aircraft Components. Howard Eubank and George F. Brown (*Western Metals*, 1951, 9, (6), 42–44).—T. G.

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[Contents : M. Pourbaix, " Equilibres électrochimiques "; P. Delahay, M. Pourbaix, and P. Van Rysselberghe, " Diagramme potentiel pH du plomb : Comportement électrochimique et corrosion du plomb : Accumulateurs à plomb "; P. Delahay, M. Pourbaix, and P. Van Rysselberghe, " Diagramme potentiel pH de l'argent : Comportement électrochimique et corrosion de l'argent "; P. Delahay, M. Pourbaix, and P. Van Rysselberghe, " Diagramme potentiel pH du zinc : Comportement électrochimique et corrosion du zinc "; P. Delahay, M. Pourbaix, and P. Van Rysselberghe, " Comportement électrochimique de l'oxygène, de l'eau oxygénée et des radicaux OH et HO₂ "; G. Valensi, " Contribution au diagramme potentiel pH du soufre "; A. Juliard, C. Rorive-Boute, and D. Bermane, " Etude du comportement électrochimique d'un élément à partir de courbes d'électrolyse "; R. Piontelli, " Considérations sur la cinétique électrochimique "; H. Fischer, " Einfluss von Inhibitoren auf elektrochemische Prozesse an metallischen Grenzflächen "; L. Gierst and A. Juliard, " Etude oscillo-graphique des régimes transitoires de diffusion "; M. Pourbaix, " Mécanisme de l'oxydation électrochimique "; R. Piontelli and G. Poli, " Vue d'ensemble sur les phénomènes de polarisation des métaux "; R. Piontelli and C. Guerci, " Influence du matériau cathodique sur l'électrodeposition des métaux "; P. Van Rysselberghe and G. A. Murdock, " A Two-Step Polarographic Reduction of Hydrogen Peroxide Caused by the Presence of Magnesium or Manganous Ion "; P. Delahay, " A Polarographic Method for the Indirect Determination of Polarization Curves for Oxygen Reduction on Various Metals "; R. Piontelli, " Propriétés ioniques et réticulaires et comportement électrochimique des métaux "; R. Piontelli, " Influence de l'anion sur le comportement électrochimique des métaux "; F. E. W. Wetmore, " The Associate Committee on Corrosion Research and Prevention of the National

Research Council of Canada "; A. Rius and A. S. Terol, " Potentiel et passivité "; L. Cavallaro and A. Iandelli, " Méthodes d'étude des inhibiteurs et stimulateurs de corrosion "; M. Pourbaix and P. Van Rysselberghe, " Remarques sur la corrosion et sur la passivation des métaux "; F. Tödt, " Die kathodische Stromlieferung von Metall-oberflächen als Mass für die Oxydbedeckung, die Wasserstoffbeladung und die Konzentration an gelöstem Sauerstoff "; M. Pourbaix and C. Rorive-Boute, " Procédés électrochimiques anti-incrustants "; L. Meunier, " Sur le fonctionnement dynamique d'une cellule de polissage électrolytique du cuivre "; G. Valensi, " Application à l'étude de la chimie générale et de la chimie analytique "; G. Charlot, " Définition des divers types de potentiels d'oxydo-réduction utiles en chimie analytique : Potentiels apparents "; R. Gauguin, " Essais d'utilisation des courbes de polarisation en analyse quantitative : Potentiométrie à courant non nul et ampérométrie à potentiel d'électrolyse constant "; R. Gauguin, " Potentiel irréversible : Courbe de polarisation et réactions chimiques : Cas de l'ion thyocyanique "; A. Juliard and L. Gierst, " Applications analytiques de la mesure des variations de potentiel à l'aide de l'oscillographe à déflection radiale "; J. J. MacDonald and F. E. W. Wetmore, " A Test of the Diffusion Term in the Ilkovic equation "; P. Van Rysselberghe, " Polarographic Detection of Per-carbonic acid and Percarbonates "; P. Delahay, " Recent Advances in Oscillographic Polarography "; A. Rius, " Polarisation lors du dépôt du fer sur une veine de mercure "; L. Gierst and A. Juliard, " Perfectionnements à la technique polarographique classique "; G. Sartori, A. Liberti, and G. Galzolari, " Le comportement polarographique du groupement SH "; H. J. Antweiler, " Polarographische Maxima und die Grenzflächenströmungen am Quecksilbertropfen "; P. Van Rysselberghe, " Nomenclature et définitions électrochimiques : Bases fondamentales "; E. Lange, " Uber elektrochemische Grundbegriffe "; R. Piontelli, " Définition et signification de quelques grandeurs électro-chimiques "; R. H. Brown and R. D. W. Williams, " Relationship of the Measured Potential of an Electrode with the Free Enthalpy Changes of the Electrode Reaction "; R. Pourbaix, " Deux expériences de cours en thermodynamique électrochimique "; R. Piontelli and G. Bianchi, " Méthodes expérimentales particulières en électrochimie."]

*Field, Samuel, and A. Dudley Weill. *Electroplating : A Survey of Modern Practice, Including the Analysis of Solutions.* Sixth edition. 7½ × 5 in. Pp. xii + 546, with 96 illustrations. 1951. London : Sir Isaac Pitman and Sons, Ltd. (20s.)

*Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métallurgiques. [*13me*] *Congrès les 21, 22, et 23 Juin 1950.* 24 × 15 cm. Pp. 196, with numerous illustrations. [1951]. Paris (VIIᵉ) : Laboratoire Centrale de l'Armement, 1 Place Saint-Thomas d'Aquin.

[Contents : — Jean, " Sur différentes méthodes d'analyse spectro-graphique des aciers "; J. M. Lopez de Azcona et A. Camunas Puig, " Etude de l'effet de l'état physique dans le système binaire Sb–Sn " (see *Met. Abs.*, this vol., col. 309); W. Van Tongeren, " Sur quelques difficultés expérimentales de l'évaluation directe des intensités des raies spectrales "; — Parisot, " Charbons pour spectrographie "; G. Brucelle, " Etablissement d'un mode opératoire pour le dosage du carbone dans les aciers par voie spectrographique "; R. Breckpot et — de Clippeleir, " Le multispectromètre électronique. Application au dosage du carbone dans l'acier "; J. Orsag, " Un équipement spectro-graphique à lecture directe "; A. Hans, " Dosage des laitons par analyse spectrale directe " (see *Met. Abs.*, this vol., col. 310); — Cojan, " Les spectrographes à grande luminosité "; (Mlle) Marie-Louise Delwaulle, " Application de l'effet Raman à l'étude du germano-chloroforme, du germanobromoforme et des germanochlorobromo-formes "; Félix François, " Mise en évidence par l'effet Raman de l'influence du pouvoir ionisant de solvants sur certaines dissociations "; J. M. Lopez de Azcona, A. Santos Ruig, et Dean Gelbenzu, " Recherche et dosage des oligoéléments métalliques dans les matières et les tissus organiques "; — Libessart, " Spectrophotométrie des hautes tem-pératures de flammes de combustion vive en régime évolutif "; A. C.

Menzies, "Nouvelle lampe et nouveau spectrographe pour l'effet Raman"; P. Michel, "Contribution à l'étude comparative du générateur Feussner, construction Durr et du générateur Multisource pour l'analyse des aciers et des fontes"; P. Michel, "Recherche d'une émulsion sensible sur film pour l'analyse spectrographique rapide"; R. Schmidt et F. Woutman, "La suppression des bandes de cyanogène dans le spectre du graphite"; Henri Triché, "Etude des spectres du fer obtenus à partir du fer pur et du sulfure de fer"; E. Lœuille, "Mesure de l'intensité absolute d'une raie".]

*Groupement pour l'Avancement des Méthodes Spectro-graphiques des Produits Métallurgiques. *Colloque International de Spectrographie de Strasbourg, les 12, 13, et 14 Octobre 1950.* 24·5 × 16 cm. Pp. 208, illustrated. [1951]. Paris (VIIe): Laboratoire Centrale de l'Armement, 1 Place Saint-Thomas d'Aquin.

[Contents: M. Jean, "Echantillons-Etalons pour métaux"; J. Gillis et J. Eeckhout, "Etude comparative des résultats de l'analyse spectrale et de l'analyse chimique. Importance des étalons"; — Parisot et — Badoz, "Charbons pour spectrographie"; M. F. Hasler, "Méthode de calcul convenant à l'analyse spectrale d'un système à composants multiples"; M. F. Hasler, "Analyse d'aciers inoxydables couvrant une gamme étendue de teneurs à l'aide du Quantomètre A.R.L."; F.-C. Mathieu, "Le 'Spectro-lecteur' automatique pour l'analyse spectrale à lecture directe"; Ernest H. S. van Someren, "The Search for the Cause of Calibration-Line Drift in the Spectrographic Analysis of Steel"; F. W. J. Garton et R. J. Webb, "Display Microphotometry"; W. Rollwagen, "Folgeerscheinungen der Temperaturanregung im Kohlebogen"; Heinrich Kaiser, "Zusammenhang des spektralen Charakters von Funkenentladungen mit den elektrischen Daten"; B. Rosen et I. Ottelet, "Utilisation de la cathode creuse et d'un four à vide en spectroanalyses"; H. Schüler, "Uber die Möglichkeit der Anwendung von Hohlkathodenentladungen für spektralanalytische Untersuchungen"; A. Gatterer, "Die Bestimmung der Halogene sowie von Schwefel und Selen auf spektrochemischem Wege"; Heinrich Kaiser, "Systematische Theorie der spektrochemischen Auswertung"; Heinrich Kaiser, "Uber ein neues Rechengerät für spektrochemische Analysen"; Henri Triché, "Analyse d'une phase d'un alliage et manifestations de l'état physicochimique d'un élément sur son spectre d'émission"; D. M. Smith, "L'organisation des Groupements spectroscopiques en Grande-Bretagne"; Wolfgang Seith, "Uber den Stand und die Organisation der Spektrochemie in Deutschland"; O. Masi, "Le Centre de Recherches spectrochimiques italien"; F. Datwyler, "Le Groupement de spectrographie suisse".]

*Heraeus (W. C.) G.m.b.H. *100 Jahre Heraeus Hanau. Eine wissenschaftlich-technische Festschrift aus Anlass des 100jährigen Jubiläums der Firma W.C. Heraeus G.m.b.H., Platinschmelze Hanau.* Zusammengestellt von K. Ruthardt. 23 × 17 cm. Pp. 406, with many illustrations. [1951]. Hanau: The Company.

[Contents: Walther Gerlach, "Der W. C. Heraeus G.m.b.H.-Platinschmelze zum hundertsten Geburtstag"; G. Grube, A. Schneider, und U. Esch, "Das System Gold-Platin"; E. Grafried, K. Protzmann, K. Ruthardt und H. Speidel, "Aus dem Arbeitsgebiet 'Spinndüsen'"; G. Grube und R. Jauch, "Das System Palladium-Aluminium"; W. Köster und F. Braumann, "Uber den Einfluss eines Palladium-Zusatzes auf die Konstitution der Silber-Zink-Legierungen"; W. Köster und U. Zwicker, "Das System Palladium-Zink"; H. L. Grube, "Chemische Untersuchungen an silberreichen Zahnlegierungen"; H. Fischer, "Kontaktwerkstoffe für die Schwachstromtechnik"; W. Köster und E. Horn, "Zum Aufbau der Systeme des Osmiums mit Kobalt und Nickel"; E. Raub und Paul Walter, "Die Legierungen der Platinbeimetalle mit Wolfram"; Gretelore Raunecker, "Aus unserem Arbeitsgebiet 'Federspitzen'"; K. Ruthardt und A. Schott, "Aus dem Arbeitsgebiet 'Netze'"; E. A. Bommer, "Uber eine neue Elektrodenform zur elektrolytischen Kupfer-Blei-Bestimmung in Bleibronzen"; Hans Oberländer, "Uber die quantitative Spektralanalyse der Platinmetalle"; H. L. Grube, "Uber ein neues galvanisches Hartglanzversilberungsverfahren und die Eigenschaften der damit erzielten Niederschläge"; R. Kieffer und E. Nachtigall, "Zum Oberflächenschutz hochschmelzender Metalle"; R. Ackermann, "Die Entwicklung der elektrisch beheizten Laboratoriumsgeräte"; W. Seith und H. Schmeken, "Sinterwerkstoffe aus metallischen und nichtmetallischen Pulvern"; F. Lieneweg, "Die Bestimmung von Temperaturmessfehlern mittels Thermometer-Kennzahlen"; E. Killius, "Die Entwicklungsgeschichte der elektrischen Widerstandsthermometer"; Walter von Sauer, "Normalwiderstände höchster Genauigkeit und Konstanz"; W. Brügel, "Das Emissionsvermögen des heissen Rotosils"; H. Mohn und A. Lorenz, "Der Heraeus-Infrarotstrahler und seine Strahlungseigenschaften"; H. Mohn, "Geschmolzener Quarz als Baustoff für elektrisch beheizte Geräte für Laboratorien und für die chemische Industrie"; H. Mohn, "Die optischen Eigenschaften des optischen Quarzglases"; L. Hiesinger, "Die Entwicklung der Hochvakuum-Aufdampftechnik"; L. Hiesinger und H. König, "Optik und Struktur kathodenzerstäubter Schichten unter besonderer Berücksichtigung der Reaktion mit dem Restgas"; W. Reichelt, "Zur Bandbreite der Reflexionsauslöschung von Zweifachschichten".]

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*Institution of Metallurgists. *Principles of Extraction and Refining of Metals.* Lectures delivered at the Institution's Refresher Course 1950. 8½ × 5½ in. Pp. 102, illustrated. 1951. London: The Institution, 4 Grosvenor Gardens, S.W.1. (12s. 6d.; 6s. 6d. to members of the Institution.)

[Contents: A. J. E. Welch, "Physical Chemistry and Its Use in Extraction Operations"; E. J. Pryor, "The Place of Mineral Dressing in Extraction Metallurgy"; C. W. Dannatt, "Principles of Ore Reduction"; L. Northcott, "Fundamentals of the Production of Metal and Alloy Ingots"; F. D. Richardson, "Principles Underlying Refining Processes".]

*International Tin Study Group. *Tin 1950–1951: A Review of the World Tin Industry.* 24 × 17 cm. Pp. 84. [1951]. The Hague: International Tin Study Group, 7 Carel van Bylandtlaan. (6s.; 3 guilders; $0.75; 45 Belgian fr.; 300 French fr., post free.)

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*Krekeler, Karl. *Zerspanbarkeit der metallischen und nicht-metallischen Werkstoffe.* 24 × 16·5 cm. Pp. xii + 358, with 148 illustrations. 1951. Berlin-Göttingen-Heidelberg: Springer-Verlag. (DM 34.50.)

*Kröger, Carl. *Grundriss der Technischen Chemie.* Teil III. *Technische Elektrochemie und Metallurgie.* 23 × 15·5 cm. Pp. 167, with 158 illustrations. 1951. Göttingen: Vandenhoeck und Ruprecht. (Brosch., DM 12.80.)

[The other parts of the series are of no metallurgical interest.]

*Lewis, D. M. *Magnetic and Electrical Methods of Non-Destructive Testing.* Report prepared for the Magnetic and Electrical Methods Sub-Committee of the British Iron and Steel Research Association. 8½ × 5½ in. Pp. 242, with 118 illustrations. 1951. London: George Allen and Unwin, Ltd. (35s.)

*Masing, Georg. *Grundlagen der Metallkunde in anschaulicher Darstellung.* 3. Auflage. 23 × 15·5 cm. Pp. viii + 148, with 140 illustrations. 1951. Berlin W35 (West-Berlin): Springer-Verlag. (Brosch., DM 12.60.)

*Mellon Institute. *Current Scientific Researches in Mellon Institute, 1950–51.* The Thirty-Eighth Annual Report of the Director, E. R. Weidlein, to the Board of Trustees of the Institute, for the Fiscal Year Ended February 28, 1951. 11 × 8½ in. Pp. iv + 52, illustrated. 1951. Pittsburgh (13), Pa.: The Institute, 4400 Fifth Avenue.

*Ministry of Supply. *Powder Metallurgy.* (Selected Government Research Reports, Vol. 9.) 9½ × 6 in. Pp. [iv] + 159, illustrated. 1951. London: H.M. Stationery Office. (18s.)

[Contents: C. J. Leadbeater, "The Surface Energies of Metals and Alloys"; L. Northcott and C. J. Leadbeater, "The Examination of Sintered Metal Components"; C. J. Leadbeater, L. Northcott, and F. Hargreaves, "Some Properties of Engineering Iron Powders"; G. E. Gardam, "Production of Iron Powder by Electrodeposition"; L. Northcott and C. J. Leadbeater, "German Sintered-Iron Bearings"; L. Northcott and B. D. Burns, "Trials of Sintered-Iron Driving Bands for 20-mm. Oerlikon Ammunition"; L. Northcott and C. J. Leadbeater, "Sintered Iron–Copper Compacts"; R. L. Bickerdike "An Aluminium Alloy Made by Powder Metallurgy" (see *Met. Abs.*, 1947–48, 15, 345); C. J. Leadbeater, "Notes on German Developments in Non-Carbide Powder Metallurgy (1939–1945)" (see *Met. Abs.*, 1947–48, 15, 346); R. L. Bickerdike and D. A. Sutcliffe, "The Tensile Strength of Titanium at Various Temperatures".]

*Mond Nickel Co., Ltd. *The Mechanical Properties of Nickel Alloy Steels.* 9¾ × 6½ in. Pp. 84, illustrated. 1951. London: The Company, Sunderland House, Curzon Street, W.1. (Gratis.)

*Parkinson, N., S. C. Britton, and R. M. Angles. *Electrodeposited Tin–Nickel Alloy Coatings.* $9\frac{3}{4} \times 7\frac{1}{4}$ in. Pp. 12, with 4 illustrations. 1951. Greenford, Middx.: Tin Research Institute, Fraser Road. (Gratis.)

[Reprinted from *Sheet Metal Ind.*, 1951, 28, (292), 756–767, 770.]

*Price, J. W. *Sampling and Analysis of Tin Ingots.* $9\frac{3}{4} \times 6$ in. Pp. 5. 1951. Greenford, Middx.: Tin Research Institute, Fraser Road. (Gratis.)

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*Thompson, P. F. *Corrosion of Metals : Metals Under Stress. I.—Aluminium.* $11 \times 8\frac{1}{2}$ in. Pp. 22, illustrated. Report ACA-49. 1950. Melbourne: Commonwealth of Australia, Department of Supply, Aeronautical Research Consultative Committee.

[For abstract, see *Met. Abs.*, this vol., col. 295.]

*Tibbenham, L. J. *Encyclopædia of Oxy-Acetylene Welding.* $8\frac{1}{2} \times 5\frac{1}{2}$ in. Pp. v + 73, illustrated. 1951. London: Sir Isaac Pitman and Sons, Ltd. (10s. 6d.)

*U.S. Department of Commerce, National Bureau of Standards. *Recommendations of the International Commission on Radiological Protection and of the International Commission on Radiological Units, 1950.* (National Bureau of Standards Handbook 47.) 8×5 in. Pp. vi + 29. 1951. Washington 25, D.C.: U.S. Government Printing Office. ($0.15.)

*Wiggin (Henry) and Co., Ltd. *The Nimonic Alloys.* 9×6 in. Pp. [v] + 40, illustrated. 1951. Birmingham (16): The Company, Wiggin Street. (Gratis.)

[Data on the properties and heat-treatment of wrought and cast Nimonic alloys are given.]

NEW JOURNALS

Cuivre, Laitons, Alliages. Published bimonthly by Riegel, 45 Avenue du Roule, Neuilly (Seine). First issue is dated March/April 1951. (1600 fr. per annum.)

Reviews of Pure and Applied Chemistry. Published quarterly by the Australian Chemical Institute, Kelvin Hall, 55 Collins Place, Melbourne, C.1, Vic. Vol. 1, No. 1 appeared in March 1951. (A£1 5s. per annum.)

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25 — BOOK REVIEWS

Interpretation of Tests and Correlation with Service. A Series of Four Educational Lectures on Interpretation of Tests and Correlation with Service. Presented to Members of the American Society for Metals during the Thirty-Second National Metal Congress and Exposition, Chicago, 23–27 October 1950. 9×6 in. Pp. [v] + 198, illustrated. 1951. Cleveland (3), O.: The Society, 7301 Euclid Avenue. ($5.00.)

Of the four educational lectures contained in this book, the first, " Correlation of Laboratory Tests and Service Performance ", by M. F. Garwood, H. H. Zurburg, and M. C. Erickson, describes fatigue tests upon six steels used by the Chrysler organization, and upon three automobile components ; a steering knuckle, a front-suspension coil spring, and a rear axle shaft. Changes in material and in processing, particularly the effects of shot peening, were investigated, with resulting strengthening or re-design to save weight. Service tests upon vehicles, which confirmed the conclusions of the laboratory tests, are referred to briefly. The stated objective of demonstrating a definite correlation between laboratory tests and service performance is, perhaps, not very strongly brought out, but the technique of component testing and the usefulness and limitation of materials data in the design of components are well described. There is an admirable account of the preparation of both notched and plain fatigue test-pieces, which includes a description of a rotational quenching device and comparison of surface finishes. A rather surprising result, which well merits further investigation, is that better fatigue properties are shown by the best longitudinal polishing finishes than as a result of shot peening, though the merits of the latter upon as-forged surfaces are emphasized. The inclusion of rather condensed accounts of stress analysis and photo-elastic experiments, while illustrating the part played by these in the investigation, somewhat interrupts the main theme, and there is regrettable inconsistency and inaccuracy in expressing the percentage improvements in properties secured by the various treatments. One or two obvious misprints were also noted. In spite of its mainly ferrous interest, this paper has much to commend it to non-ferrous metallurgists.

The second lecture, " Limitations of Mechanical Testing ", by M. Gensamer, is a somewhat academic essay which will not be informative to those interested in limitations of specific tests or test methods. The limitations of mechanical testing are logically treated, but rather from the point of view of research than that of control testing. Two omissions of importance are the absence of warning that generally accepted tests of uniformity may be valueless if conditions change (for example, room-temperature tests have no significance under high-temperature conditions) and the lack of reference to the important factor of extensometer sensitivity in the ordinary tensile test.

J. T. Burwell, the author of the third lecture, " Wear Tests and Service Performance ", has a very difficult task. He provides an excellent review of the different kinds of wear met in service, and in an Appendix he briefly mentions various types of wear-testing devices which have been reported in the literature. In spite of his best efforts, however, he is unable to provide a real link between the Appendix and the paper, of which the most interesting part describes research on the fundamentals of certain types of wear, taking into account the true contact area between rubbing surfaces.

The fourth lecture, " Corrosion Tests and Service Performance " by F. L. LaQue, is one of the most useful which has ever been printed on the subject of corrosion. It is packed with common sense, so much condensed that a précis is impossible. Three of the conclusions cannot too often be reiterated and must be quoted in full :

" When performance in service does not match the performance in test, then some important factor was not recognized or subjected to proper control in the test or taken properly into consideration in applying the results."

" Any test designed to accelerate corrosion by distorting one or more of the controlling factors or introducing a corrosion agent not encountered in actual service involved is almost sure to give misleading results."

" The principal requirement in the planning of a test or in using the results is to apply as much common sense as possible."

H. W. G. HIGNETT.

X-Ray Analysis of Crystals. By J. M. Bijvoet, N. H. Kolkmeyer, and C. H. MacGillavry. $9\frac{1}{4} \times 6$ in. Pp. xii + 304, with 226 illustrations. 1951. New York: Interscience Publishers Inc. ($6.50); London: Butterworths Scientific Publications, Ltd. (50s.; 51s., post free.)

This work forms a welcome addition to the list of elementary books on X-ray diffraction. Students of all branches of the subject will find that it provides a useful general intro-

duction, illustrated by some refreshingly original approaches and diagrams; the authors have certainly avoided many of the stock figures which make some books appear dull even before one reads them.

As an elementary text-book, then, the book can be strongly recommended. As an advanced text-book, however, it is disappointing, for it deals with a much wider range of subject than the title would suggest, and serves rather to whet the appetite than to provide a satisfying meal. Perhaps this is the authors' intention; their other published works indicate that they must have held themselves severely in check in deciding how far to extend the treatment of many of the topics with which they deal.

The breadth of approach is illustrated by a general survey of the contents. The book begins with a treatment—essentially descriptive and non-mathematical—of crystallography and X-ray diffraction, and then deals with experimental techniques and their applications; it continues with two chapters on crystal-structure analysis, and concludes with three chapters on the classification of crystals and crystal chemistry. There are several appendices, which cover nearly seventy pages.

The metallurgist will not find the book particularly informative, as metals are discussed only incidentally; X-ray powder photographs of some metals and alloys are shown, and some brief descriptions of simple structural types are given. Preferred orientation, order–disorder phenomena, and the study of cold-worked metals are discussed, but none is developed in any great detail. The last subject is particularly disappointing, as no reference is made to much of the work that has been published, and which conflicts with that which the authors have chosen to mention.

The book is beautifully printed and produced, and it is therefore to be deplored that several departures from current scientific practice occur. First, there is no name index. Secondly, the references are given as a mixture of bibliography and literature references; they are given in groups at the ends of the chapters and are thus not easy to find. Thirdly, footnotes are extensively used. Fourthly, use is made of change of type instead of parentheses in listing headings and so forth. For example, one legend reads:

Figure 48. *Back reflection diagrams of* a *cold rolled electrolytic iron* b *after recrystallization at 500° C.*

A re-reading of this makes the meaning clear, but re-reading should not be necessary. And, finally, why does only Bijvoet's name appear on the spine?

To summarize, the book forms a valuable introduction for the student, but its range is too wide to be of great value to the specialist. The metallurgical content is not important.

<div align="right">H. LIPSON.</div>

The Interpretation of X-Ray Diffraction Photographs. By N. F. M. Henry, H. Lipson, and W. A. Wooster. $10 \times 7\frac{1}{2}$ in. Pp. ix + 258, with 231 illustrations. 1951. London: Macmillan and Co., Ltd. (42s. net.)

X-ray diffraction is now applied in many diverse fields of science, and it is true to say that in none of these has it been more successful than in the study of metal structures. The book under review, being concerned with the techniques of producing X-ray-diffraction photographs and with their interpretation, is primarily of interest to practising crystallographers. But it is essentially a handbook in which is to be found all that is required to extract the maximum information from X-ray observations and, as such, will also be invaluable to the X-ray metallographer. The authors, in producing a work of such general usefulness, have rendered a fine service to students and research workers in crystallography and particularly to those who have not had the benefit of a formal course in the subject.

After three concise chapters on crystal symmetry, X-rays, and geometry of X-ray reflections, which are necessary for the full treatment of the subject, in four subsequent chapters all the main methods of recording diffraction patterns of single-crystal and polycrystalline specimens are described

with the aid of a number of effective diagrams. By presenting the subject in a geometrical way, together with charts and example patterns reproduced to scale, interpretation of these patterns is simplified and the student provided with plenty of exercises.

The remaining ten chapters are of the greatest value to the X-ray metallographer, since they range over his whole field of interest, covering, as they do, orientation of single crystals and polycrystalline aggregates; powder photographs; measurement of cell dimensions, intensities of X-ray reflections, and of grain-size; and, finally, identification of crystalline materials. Although the natural place for Chapter 10 would appear to be following Chapter 1, metallurgists concerned with the study of structural transformations in alloys will find it useful.

A few deficiencies occur, however, which are only to be expected when the authors cover so wide a field. Thus, the section in Chapter 2 on "Production of X-Rays" is far too elementary, and some mention ought to have been made of fine-focus and rotating-anode tubes. The authors are rather prejudiced in favour of the ordinary powder camera when they state (p. 74) that the greater accuracy in measurement of cell dimensions claimed for the focusing camera is not borne out by results. Reference ought also to have been made, in the section on measurement of grain-size from "spottiness" of diffraction rings, to the paper by Stephen and Barnes (*J. Inst. Metals*, 1937, 60, 285). The authors have managed in the main to steer clear of controversial issues, but they have resurrected the old controversy over interpretation of line broadening in X-ray-diffraction patterns of cold-worked metals and are biased in favour of their own views.

The inclusion of the last chapter on identification of crystalline materials by means of powder photographs is welcome, since the technique is becoming established as a useful tool in industrial laboratories. The distinction is drawn between identification of elements by the spectrographic method and of crystalline materials from their powder-diffraction patterns. The latter task is the more formidable, since the number of crystalline compounds runs into many thousands. It is for this reason that the A.S.T.M. index of diffraction patterns, which is rapidly expanding, was produced. In this connection, the authors would have done well to have acknowledged the contribution of the British Institute of Physics. It is difficult to understand the criticism (p. 220) of the indexing system adopted by the A.S.T.M. The index is for the purpose of identifying crystalline compounds by their diffraction patterns, the two main characteristics of which are line position and intensity. The choice of interplanar spacing d makes the first characteristic independent of camera constants and wavelengths used. For the identification of a pattern in the index, great accuracy is not required in measurement of either the spacing d or the line intensity, and the whole procedure may be speeded up by using a calibrated d scale, such as that mentioned in another section (p. 176), and by estimating intensities visually.

New methods are constantly being developed and improvements made in existing techniques, so that it is hoped we may look forward to a supplement in a few years' time.

The ground, however, is covered very effectively by the present volume, and the adoption of a numerical system of indexing should make the book useful both as a standard text-book and a work of reference.

<div align="right">T. LL. RICHARDS.</div>

Practical Electron Microscopy. By V. E. Cosslett. $9\frac{3}{4} \times 6$ in. Pp. xiii + 299, with 147 illustrations. 1951. London: Butterworths Scientific Publications, Ltd. (35s.; by post 36s.)

In the decade since the electron microscope emerged as a tool promising to reveal new information over a wide field of investigation, many hundreds of papers and a number of books have appeared dealing with aspects of design and

operation, with methods of specimen preparation, and with the interpretation of images. The period has been one of experiment and change both in microscope construction and in techniques for the examination of specimens; only in the last year or two have signs appeared that a degree of maturity in both has been reached. From now on it seems that developments must take place more slowly.

Many microscopes are now in service, in the hands of operators who have not all had the privilege of contributing to the development of the subject and whose interest centres primarily on the success with which what is now an established instrument can be applied. It is sufficient that they should understand and be able to perform the now orthodox methods of microscope maintenance and adjustment and of specimen preparation. For them particularly, Dr. Cosslett's book will form a standard text for some years to come. At the same time, however, it is a concise and readable account which can be commended to the reader who has only a general interest in the scope of this new branch of microscopy.

The first half of the book is occupied by a description of the essential components of electron microscopes, their functions and principles of operation, and the respects in which their performance falls short of perfection. The reader is thus enabled to understand the operator's part in minimizing the image aberrations under his control, as far as need be for his purpose.

A substantial part of the remaining chapters is occupied by an introduction to specimen techniques. Of particular interest to metallurgists is Chapter 9, which deals with replica methods. The author rightly lays stress on the need for proper preparation of the metal surface to be copied, an aspect which requires much further study, and on the care required to avoid misleading interpretations due to several recognized types of artefact. Although it is no part of his purpose to establish a case for the application of the electron microscope in metallurgy, it is a pity that none of the several examples Dr. Cosslett uses in illustration of replica methods shows features of particular or readily appreciated metallurgical significance. Such examples could have been found, and would have been valuable in counteracting any impression which may prevail that, because early success has not been so readily won in the metallurgical as in other fields, the electron microscope has no service to render to metallurgy.

G. L. J. BAILEY.

Hot-Dip Galvanizing. Report of a Visit to the U.S.A. in 1950 of a Productivity Team Representing the Hot-Dip Galvanizing Industry. $9\frac{3}{4} \times 7\frac{1}{4}$ in. Pp. xiii + 57, with 12 illustrations. 1951. London : Anglo-American Council on Productivity, 21 Tothill Street, S.W.1. (3s. 3d., post free.)

The specialist team on galvanizing which visited America in September and October 1950, under the auspices of the Anglo-American Council on Productivity, has published a report which is factual and lucid. The presentation of the information follows the logical sequence of describing, in an early chapter, the galvanizing shops and ancillary equipment which were inspected. Then follow dissertations on steel pre-treatments, dipping techniques, and the processing of residues. Handling of labour and methods of costing are briefly dealt with in the concluding sections. Particularized data, obtained from specific plants, are included in eight interesting appendices. Line drawings and a sufficient number of plates illustrate salient points in the text.

As in other industrial team reports, the operative word is " productivity " and this publication lays emphasis on the value of time-and-motion study as a potent factor in its attainment. The reader will find his attention directed to the value of meticulous concern for detail in all the mechanical operations involved in hot-dip galvanizing. It is stressed that the conservation of human effort by machines and

working appliances assures a steady throughput of work and thus contributes to high productivity.

It is apparent from this report that, scientifically or technically, British galvanizers are in no way behind the Americans. The co-operative research work done by the British Non-Ferrous Metals Research Association has ensured that the fundamentals are well appreciated by the industry. The technician will be interested to compare the findings of the team with the information given in books published by American authors.

On the other hand, the true value of the report will be realized by a study of the account it gives of labour relations, costing, and time-and-motion analyses. In this, and also in the consideration of the small tool equipment used by American galvanizers, will those engaged in the industry find most food for thought.

Although the terms of reference of the team limited the review to general or jobbing galvanizing, much of the information will be of value to those, such as wire and sheet galvanizers, who work in a more specialized field. One must hope that perusal of the report will not be confined to the managerial and supervisory grades of the industry.

W. MONTGOMERY.

Electroplating for the Metallurgist, Engineer, and Chemist. By J. B. Mohler and H. J. Sedusky. $8\frac{1}{2} \times 5\frac{1}{2}$ in. Pp. 257, with 26 illustrations. 1951. New York (10) : Chemical Publishing Co., Inc., 212 Fifth Avenue. ($5.00.)

It is difficult to imagine the purpose of this book. The information given on the chemical formulation and operation of electroplating baths is so generalized and vague that even the combined genius of the three professions named in the sub-title would need much more to achieve practical results. Perhaps the intention is that, armed with the principles enunciated, and in particular the tabulated remedies in a chapter entitled " Plating Bath Troubles ", the metallurgist, engineer, or chemist can keep his end up in discussions with those who actually have to do the electroplating operations. These will readily acknowledge the truth of most of the book, but will question the practical utility of the broad principles given. The authors give the impression that provided the original formulation of an electroplating bath is correct and that it is maintained " in balance ", all will be well. Thus great weight is attached to chemical analysis, and many pages are devoted to recommended methods, some of which are unorthodox, to say the least. No consideration is given to plant and the chapter on preliminary cleaning is hopelessly inadequate.

A long chapter on electroplated alloys indicates the authors' chief interest, but it unduly simplifies the practical difficulties. A short chapter on diffusion coatings gives no clue as to how such coatings may be formed or their connection with electroplating. The few diagrams and single half-tone do not justify their space. The tables, &c., of the appendix seem even less useful than is usual.

The number of books which have recently appeared on electroplating indicate that there is a large demand, but the reviewer considers that metallurgists need not be unduly influenced by the title of this one.

G. E. GARDAM.

Leichtmetalltechnik : Derzeitiger Stand und neuere Entwicklungsrichtung. Von W. Bleicher. 21×14.5 cm. Pp. 51, with 42 illustrations. 1950. Düsseldorf : Deutscher Ingenieur-Verlag G.m.b.H. (DM 4.50.)

This book is divided into two sections, the first covering developments in the light-alloy industry generally during the last 25 years, and the second developments in Germany since the war. The descriptions of individual developments are necessarily brief, and stress practical applications rather than theoretical implications, but they are readable, up to date, and accurate.

H. W. L. PHILLIPS.

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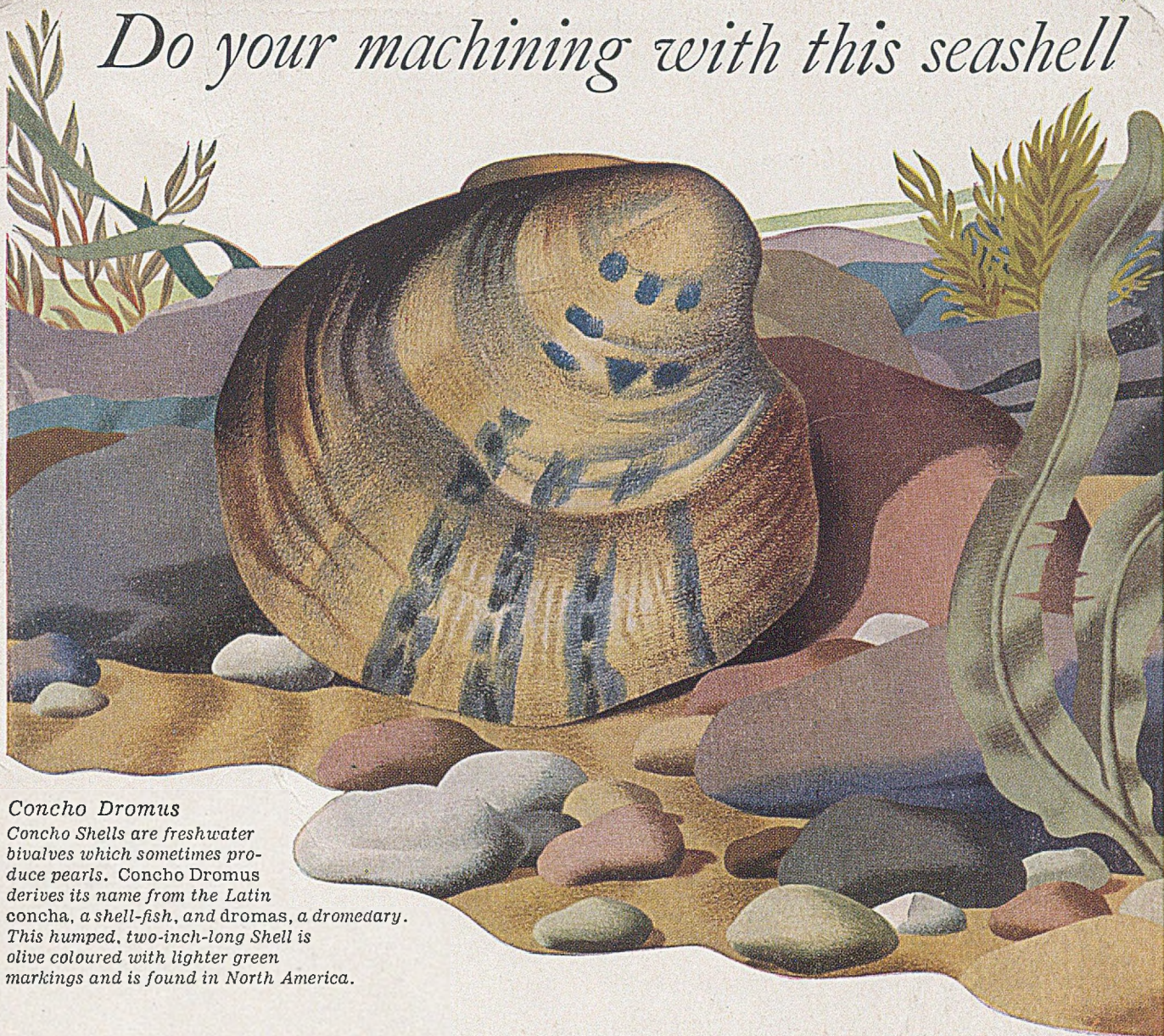
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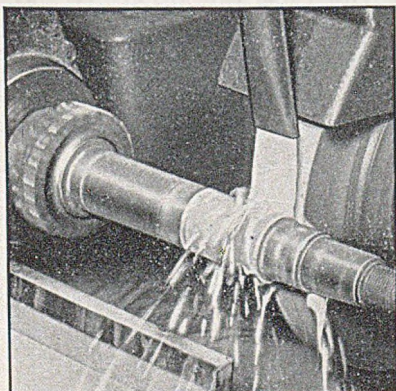
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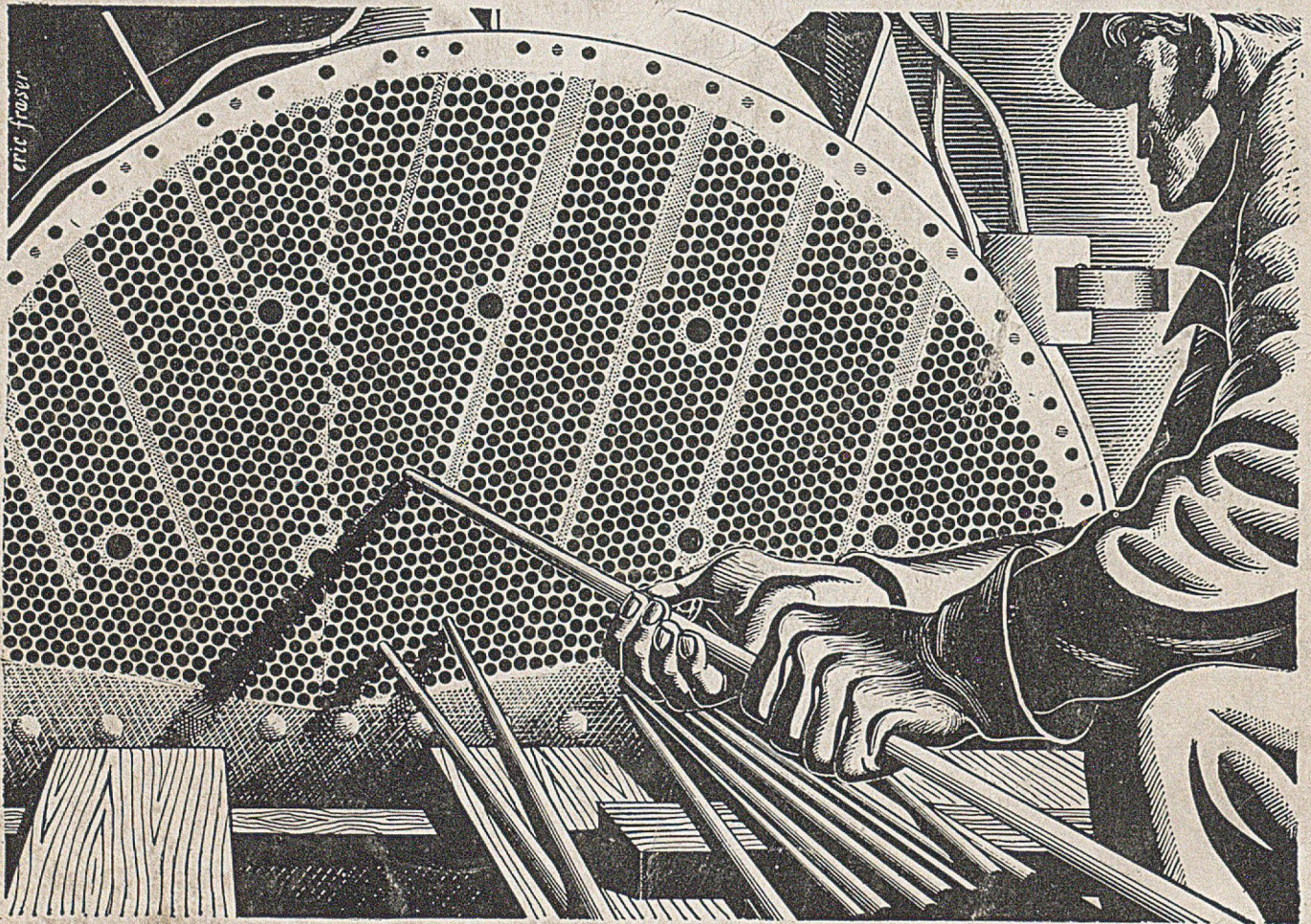
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