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P. 99/LXXIII

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S.56

The Journal of the **INSTITUTE OF METALS**

6/12/46

and

METALLURGICAL ABSTRACTS

M.



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SEPTEMBER 1946

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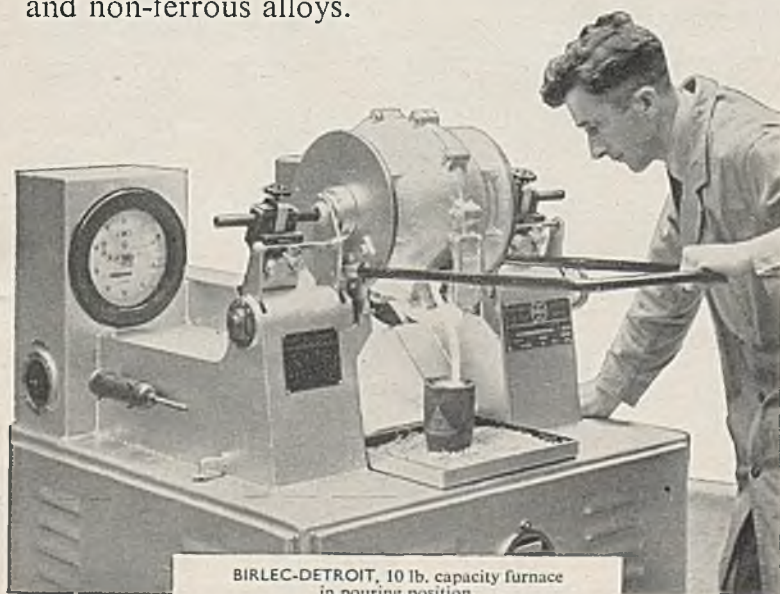
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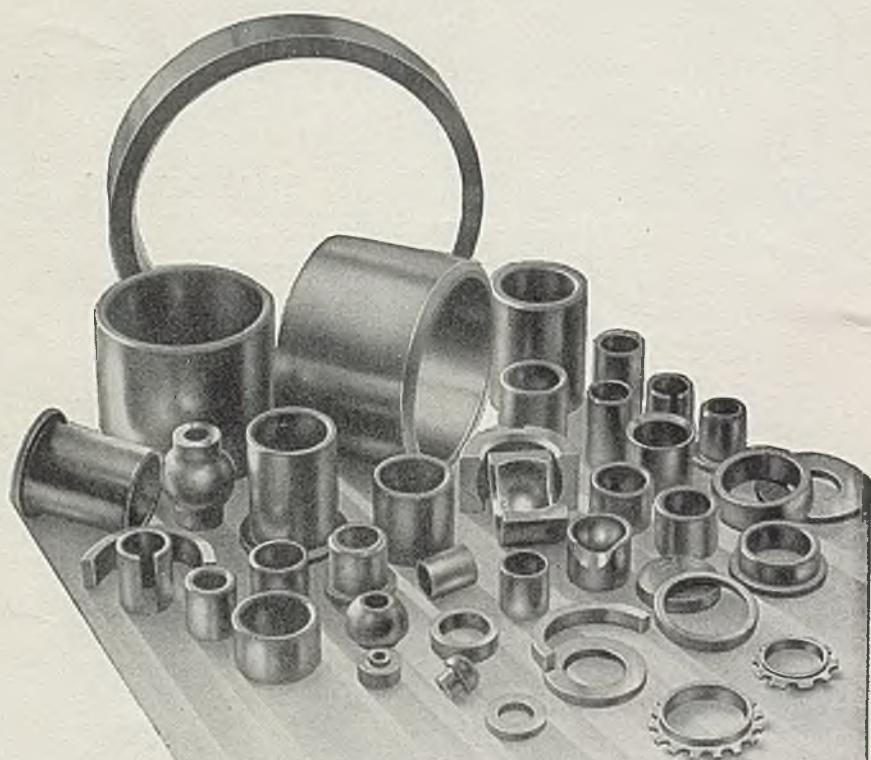
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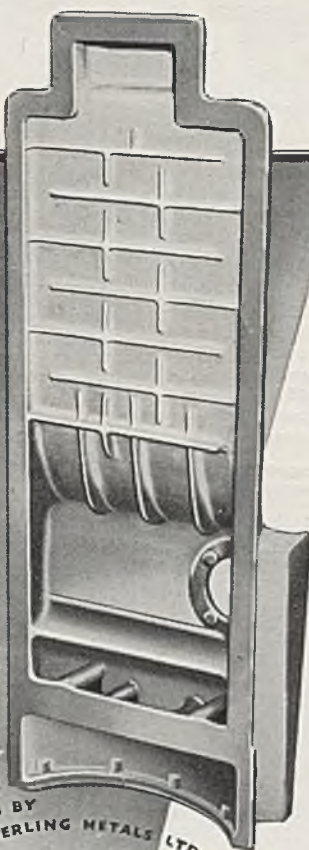
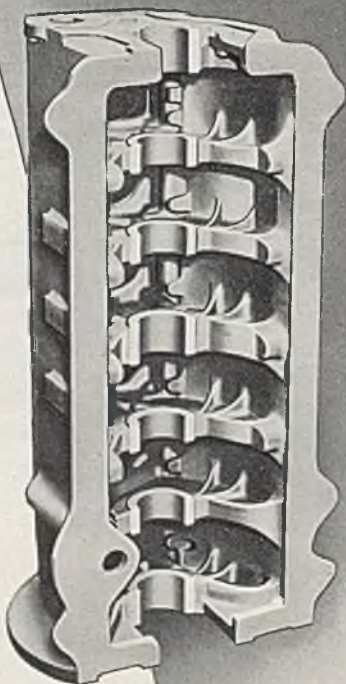
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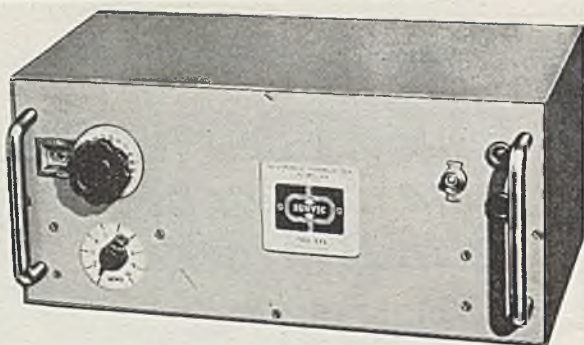
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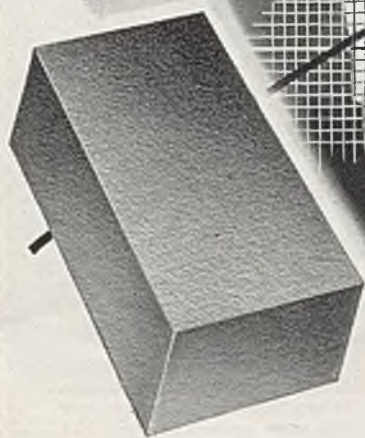
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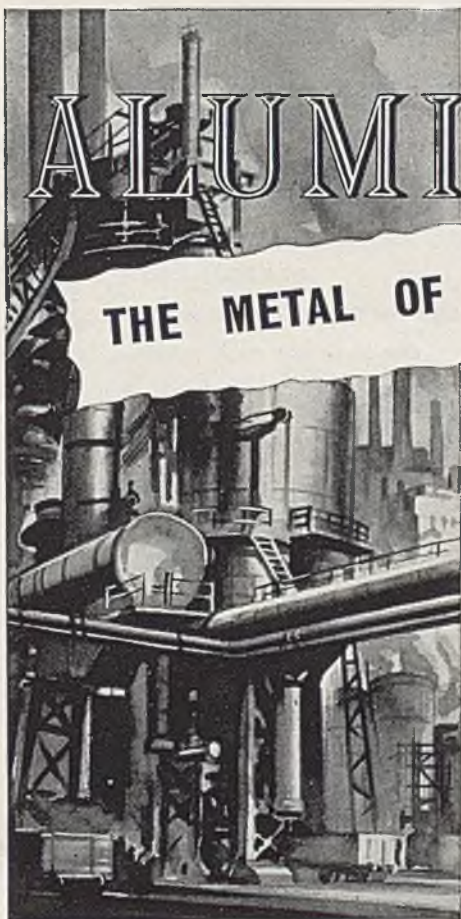
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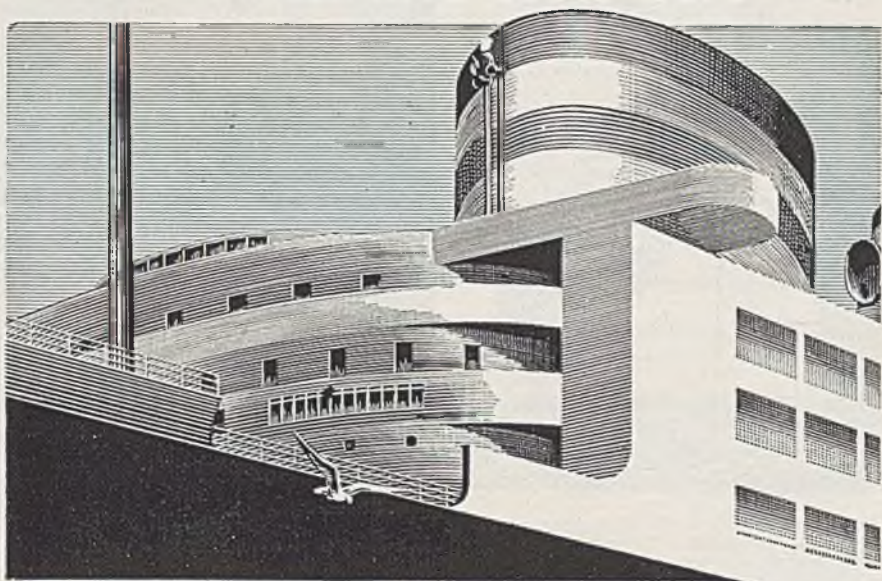
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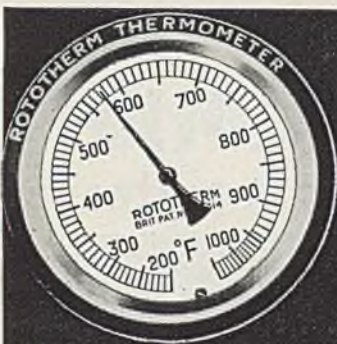
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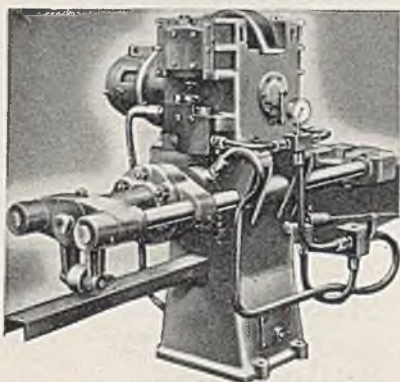
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SEPTEMBER 1946

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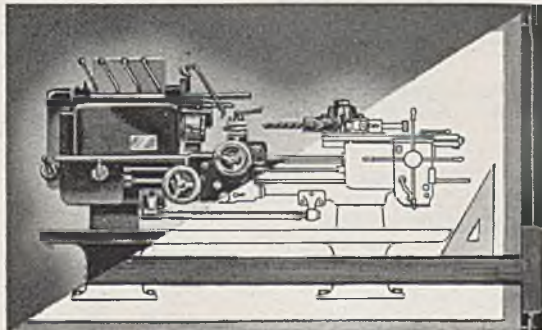
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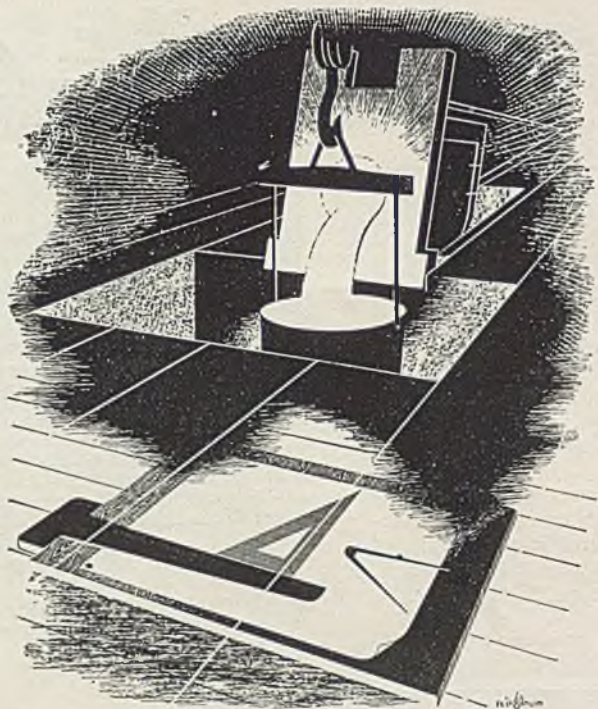
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LETTER TO THE EDITOR

Proof Stress.

DEAR SIR,

According to British Standard Specification No. 18-1938 (amended June, 1940), the proof stress of a given material is "the stress which is just sufficient to produce a permanent elongation equal to a specified percentage of the original gauge-length." This seems to be perfectly satisfactory for either ferrous or non-ferrous alloys.

The specification goes on to state "It can be determined (a) by direct measurement of the gauge length after various loads have been applied to the test-piece and removed, or (b) from the load-elongation curve by drawing a line parallel to the straight portion of the curve and distant from

it by an amount representing the required permanent elongation, thus determining the load at which the line cuts the curve. (See Fig. 1)." The specification Fig. 1 is reproduced below.

The method quoted in clause (b) of the latter paragraph is, no doubt, sufficiently near for most steels, but it is certainly considerably in error for many magnesium-base alloys, owing to the unloading branch of the stress-strain curve lacking the straightness which is normally associated with the stress-strain curve of steel and thus producing a less permanent set than corresponds to the offset straight line. As the "permanent set" limitation seems to be used quite widely as a basis of design, however (i.e. as proof stress proper), would it not be both logical and timely to modify B.S.S. 18-1938 so as to clarify this point and distin-

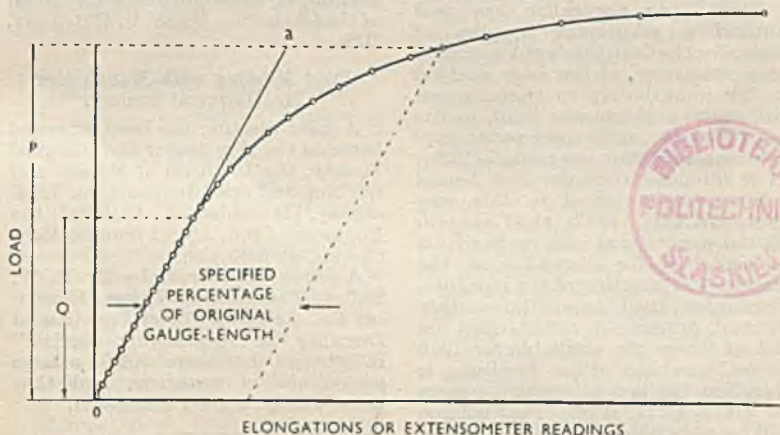


FIG. 1.—Load Extension Diagram.

If A = Original Cross-Sectional Area of the Test-Piece, Limit of Proportionality Stress = $\frac{Q}{A}$ (see Clause 1). Young's Modulus is obtained from the Slope of the Line $O-a$ (see Clause 7).

News and Announcements

guish between the true or "permanent set" proof stress and the "offset" proof stress?

Further, as the total strain incurred in producing either the permanent set proof stress or the offset proof stress is considerably greater than either the permanent set or the offset, could the aircraft manufacturers be induced to state which of the three strains is really intended to be taken into consideration in their designs?

Yours faithfully,

ARTHUR FISHER.

Deputy Chief Metallurgist,
Magnesium Elektron, Ltd.

Manchester,
August 21, 1946.

INSTITUTE NEWS AND ANNOUNCEMENTS

Journal.

It is the desire of the Council to resume the pre-war custom of providing members with bound volumes of the *Journal* and *Metallurgical Abstracts*, in addition to the monthly *Journal*. As will be remembered, the pre-war procedure had to be abandoned owing to paper shortage.

The Paper Controller has now authorized additional supplies of paper for the Institute's publications; in consequence, it has been decided to try immediately to re-commence the practice of pre-war days, in the hope that the authorized paper may be obtainable from the manufacturer. It is intended that the first bound volume to be issued in this way (Vol. LXXIII, 1947) shall contain all the papers that will be available for selection for discussion at the March, 1947, meeting of the Institute. The August, 1946, issue of the monthly *Journal*, in which were published the last of the papers available for 1946 general meetings of the Institute, is therefore the last to contain papers for Vol. LXXII, 1946. That volume will be completed by the publication of a special issue, which will contain the report of the business and discussions at the September, 1946, meeting, as well as the usual title pages, introductory matter, and index of the volume. Members will be able

to obtain in due course, post free, copies of the binding case, on application for it.

The pagination of the papers in the present issue of the monthly *Journal* (which contains the first of the papers that may normally be considered for discussion at the March, 1947, meeting), commences at page 1, while that of *Metallurgical Abstracts* will continue on the Jan.-Dec. basis, as was the case before the war.

Members should keep their monthly issues of the *Journal*, for binding as at present, until a definite announcement can be made that the paper supply is assured and that bound volumes are certain to be issued to members, in addition to the monthly *Journal*, during the present financial year.

Birmingham Local Section.

MR. E. H. BUCKNALL, M.Sc., has been appointed Honorary Secretary-Treasurer of the Section, in succession to MR. JOSIAH W. JONES, M.Sc., who has resigned on leaving the district.

London Local Section.

DR. J. H. WATSON, M.C., A.R.S.M., B.Sc., Vice-Chairman, has assumed the Acting Chairmanship of the Section, in consequence of the death of the Chairman, MR. S. V. WILLIAMS, B.Sc.

Joint Meeting with Manchester Metallurgical Society.

A joint meeting has been arranged between the Manchester Metallurgical Society, the Institute of Metals, and the Iron and Steel Institute, on Wednesday, December 4, 1946, at the Engineers' Club, Albert Square, Manchester, at 6.30 p.m.

A paper will be read by Dr. R. W. Bailey, Metropolitan-Vickers Electrical Co., Ltd., on "The Function of Ductility in Engineering Design." It is hoped that there will be a large attendance of members, and that there will be a good discussion.

PERSONAL NOTES

MR. JOHN FRY, on the occasion of his seventieth birthday, was presented by the staff and workpeople of Fry's Metal Foundries, Ltd., with

News and Announcements

his portrait painted in oils by Mr. Francis Hodge, R.O.I. In making the presentation, Mr. P. M. Parish, who has been associated with Mr. Fry almost from the formation of the firm, and Mr. F. Leech, tool-room foreman, spoke of the inspiration, encouragement, and friendship which Mr. Fry had given to all who had worked with him.

PROFESSOR N. P. GANDHI, Honorary Corresponding Member to the Council for India, was elected first President of the Bombay Metallurgical Society, which was formed in January, 1945. His term of office is for two years.

DR. J. L. HAUGHTON and DR. MARIE L. V. GAYLER (Mrs. J. L. HAUGHTON) resigned their posts at the National Physical Laboratory in April, 1946, and now reside at High View, Fishponds, Charmouth, Dorset, to which all correspondence should be addressed. Dr. Haughton has accepted the post of Consultant to Essex Aero, Ltd., of Gravesend, Kent, and has assumed the editorship of the *Journal of the British Astronomical Association*.

MR. JOSIAH W. JONES, M.Sc., has been appointed Senior Lecturer in Materials and Metallurgy at the new Aeronautical College at Cranfield.

MEMBERSHIP ADDITIONS.

These were elected on September 10, 1946:

As Members.

BATES, Frederick John, Works Director, B.S.A. Tools, Ltd., Birmingham.

BOTH, Gerardus Jan Jacobus, Delegate Member of the Board of Directors, Aluminium Rolling Mill and Extrusion Works, Ltd., Utrecht, Holland.

BROWN, William R., B.Met.Eng., Chief Development Metallurgist, A.C. Spark Division, General Motors Corp., Flint, Mich., U.S.A.

CHAPMAN, Russell Martin, B.Met., Scientific Officer, Metallurgy Division, Royal Aircraft Establishment, Farnborough, Hants.

COCHRANE, Reginald Stuart, B.Sc. (Eng.), Chemist, Royal Arsenal, Woolwich.

DAVIES, Francis Thomas Leslie, Metallurgist, British Acoustic Films, Ltd., Drybrook, Glos.

EMLEY, Edward Frederick, B.Sc., Metallurgist, Magnesium Elektron, Ltd., Clifton Junction, nr. Manchester.

HOLLEAUX, Jacques, Ingénieur en Chef, Cie Générale du Duralumin et du Cuivre, Paris, France.

KRISHNAN, Professor Sir Kariamanikkam Srinivasa, D.Sc., F.R.S., Professor of Physics, University of Allahabad, India.

KYLE, Professor Peter Edward, M.E., M.S., Professor of Applied Metallurgy, Cornell University, Ithaca, N.Y., U.S.A.

MANN, Edward Thomas, Metallurgist and Chemist, Renaldsan Bros. and Tippet Pty., Ltd., Ballarat, Victoria, Australia.

MEHTA, Ratilal K., B.Sc., Chief Engineer, Kamani Group of Industries, Bombay, India.

MINIFIE, Cyril, Plant Development Engineer, Salem Engineering Co., Ltd., Milford, nr. Derby.

MIRAVED DEL VALLE, Professor Juan José, Professor of Metallurgy, Mining School of Madrid, Madrid, Spain.

MUKERJI, J. B., M.Sc., Contro Officer, Physics Branch, Ordnance Laboratory, Cawnpore, India.

PRITCHARD, Thomas, Assoc.M.C.T., Chemist-in-Charge, I.C.I. Metals Division, Broughton Copper Works, Manchester.

PRZEGALINSKI, Stanislaw, Metallurgist, Hutniczy Instytut Badawczy (Metallurgical Research Institute), Gliwice, Poland.

ROHNER, Ferdinand, Ph.D., Research Chemist, Aluminium Industrie, A.G., Neuhausen, Switzerland.

SPURGEON, Christopher Edward, C.B.E., Chief Inspecting Engineer, Egyptian Government, London.

VACHET, Pierre, Directeur du Service des Recherches, Cie. de Produits Chimiques et Electrometallurgiques Alais, Froges et Camargue Chambery, (Savoie), France.

VEYAET, Emmanuel, (Croix de Guerre), Chef de Fonderie, Compagnie Française des Métaux, Sèrfontaine, (Oise), France.

WALKER, George Henry, Chief Analyst, Joseph Lones Laboratories, Smethwick.

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As Students.

- BRADSHAW, Rowland Patrick, Student, Christ's College, Cambridge.
- CLAREBROUGH, Leo Michael, B.Met.E., Research Student, Metallurgy School, University of Melbourne, Australia.
- DEWHIRST, David W., Student, Christ's College, Cambridge.
- DUCE, Alan Godfrey, Student of Metallurgy, St. John's College, Cambridge.
- HIPKINS, Maurice George, B.Sc., Research Metallurgist, Mond Nickel Co., Ltd., Birmingham.
- JACKSON, Brian Richard, Assistant Research Metallurgist, Dept. of Research and Technical Development, Stuarts and Lloyds, Ltd., Corby, Northants.
- KOZLOWSKI, Henry Julian, B.Sc., Metallurgist, Physical Metallurgy Research Laboratories, Bureau of Mines, Ottawa, Canada.
- LANCASTER, John F., B.Eng., Metallurgist, Aluminium Plant and Vessel Co., Ltd., London.
- LUMSDEN, Frederick, Student of Metallurgy, Battersea Polytechnic, London.
- MAXWELL, Donald, Student of Metallurgy, Emmanuel College, Cambridge.
- OGLIVIE, Graeme John, B.Met.E., Research Student, University of Melbourne, Australia.
- TAIT, Donald Blakie, B.Sc.(Eng.), Student of Metallurgy, Battersea Polytechnic, London.
- WALTERS, Roy Edwin Seaby, Student, Christ's College, Cambridge.

As Associate.

- AVERBACH, B. L., M.Met.Eng., Research Assistant, Department of Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass., U.S.A.

SYMPOSIUM ON THE FAILURE OF METALS BY FATIGUE.

A symposium, in five sessions, on the failure of metals by fatigue will be held under the auspices of the Faculty of Engineering, University of Melbourne, Australia, in a week between November 18 and December 13, 1946. There will also be excursions to the Munitions Supply Labor-

atories, Marybyrnong, and the Aeronautics Division of the Council for Scientific and Industrial Research, Fishermen's Bend. Papers to be presented for discussion are:

- J. NEILL GREENWOOD: "The Failure of Metals by Fatigue."
- H. F. MOORE: "Metallography, Fatigue of Metals, and Conventional Stress Analysis."
- W. BOAS: "Theories of the Mechanism of Fatigue Failure."
- H. SUTTON: "Fatigue Problems Associated with Aircraft Materials."
- C. W. GEORGE, S. F. GROVER, and B. CHALMERS: "The Factors Contributing to Fatigue Failure in Aircraft."
- A. G. PUGSLEY: "Repeated Loading on Structures."
- STAFF OF THE BATTELLE MEMORIAL INSTITUTE, COLUMBUS, O., U.S.A.: "Structures Liable to Fatigue Failure and Some Considerations in their Design."
- U. R. EVANS: "The Electrochemistry of Corrosion-Fatigue."
- C. W. ORR: "The Detection of Fatigue Cracks."
- F. W. HOOTON: "The Measurement of Dynamic Strain."
- W. W. JOHNSTONE: "Methods of Investigating the Fatigue Properties of Metals."
- F. S. SHAW: "Determination of Stress Concentration Factors."
- W. H. H. GIBSON: "Photo-Elasticity and Stress Concentration."
- A. L. PERCIVAL and R. WELCH: "Fatigue Tests on Four Welded H-Beams."
- G. SACHS: "Residual Stresses, Their Measurement and Their Effects on Structural Parts."
- G. G. McDONALD: "The Design of Cylindrical Shafts Subjected to Fluctuating Loading."
- J. G. RITCHIE: "Fatigue of Bolts and Studs."
- C. J. OSBORN: "The Fatigue of Welded Steel Tubing in Aircraft Structures."
- M. S. PATERSON: "Notch-Sensitivity of Metals."
- D. O. MORRIS: "Composition and Physical Properties of Steel in Relation to Fatigue."
- R. W. K. HONEYCOMBE: "Conditions Leading to Fatigue Failure in Sleeve Bearings."

News and Announcements

- A. R. EDWARDS: "Fatigue Problems in the Gas Turbine Aero Engine."
J. G. BROOKMAN and L. KIDDLE: "The Prevention of Fatigue Failures in Metal Parts by Shot Peening."
H. O'NEILL: "Failure of Railway Material by Fatigue."
G. W. C. HIRST: "The Effect of Press Fits on the Endurance Limit of Axles and Crank Pins."
E. CONNOR: "Fatigue Failure of Axles of Car and Wagon Railway Rolling Stock."
D. O'DONNELL and A. S. BUNDLE: "Some Practical Aspects of Wire Fatigue in Aerial Telephone Lines."
H. C. LEVEY and P. R. BRETT: "The Vibration of Telephone Line Wires."
S. D. CHIVERS: "Fatigue Failures of Lead Telephone Cable Sheathing."

The papers will be printed in full and will be distributed to subscribers before the meeting. Spoken and written contributions to the discussion are invited.

The complete papers with discussion will finally be printed in the form of a bound volume. The subscription rate before the Symposium is £1 5s. and includes the right to receive a full set of advance copies, attendance at the meeting, and a copy of the bound volume. Extra copies of the volume can be ordered before the meeting at the advance-copy rate of £1 1s. The charge for volumes ordered after the meeting will be £2 2s.

All enquiries and orders should be addressed to Professor J. Neill Greenwood (Chairman of the Convening Committee), Research School of Metallurgy, The University, Melbourne, Australia.

FRENCH SCIENTIFIC LIBRARY.

A small library of current scientific periodicals and recently-issued scientific books is to be established, in the near future, at the Institut Français in London. The facilities to be offered by the library are: (a) consultation of publications in the library; (b) loans by post; and (c) obtaining from France, so far as possible, books and periodicals that may be required but not available in the Library.

Members of the Institute of Metals who would like to take advantage of these facilities are requested to communicate direct with the Librarian, Institut Français, Queensberry Place, South Kensington, London, S.W.7. They will then be enrolled on the register, and will be sent any catalogue that may be issued.

LIBRARY OF CATALOGUES.

The Birmingham Exchange, Stephenson Place, Birmingham, 2, as one of its efforts to promote the recovery of trade, has formed a library of catalogues, which have been classified and indexed so that they may form a readily available source of reference for those wishing to find the names of producers of any particular articles.

While the library cannot yet claim to be complete (as paper shortage has prevented many firms from issuing catalogues), the present collection covers a great range of products, and includes the non-ferrous and ferrous materials used in the heavy industries.

The facilities are not confined only to members of the Exchange; the staff will answer enquiries from any source (Telephone: MID 1914).

LOCAL SECTION MEETINGS

LONDON LOCAL SECTION.—Dr. J. H. Watson, M.C., B.Sc., A.R.S.M.: "The Purple Alloy of Gold and Aluminium." (4 Grosvenor Gardens, London, S.W.1, Thursday, October 10, 1946, at 7 p.m.)

SCOTTISH LOCAL SECTION.—P. T. Holligan: "Modern Trends in Bearing Metals." (Institution of Engineers and Shipbuilders in Scotland, 39 Elmbank Crescent, Glasgow, C.2, Monday, October 14, 1946, at 2.30 p.m.)

News and Announcements

MEETINGS OF OTHER SOCIETIES

MANCHESTER METALLURGICAL SOCIETY.—Presidential Address. (Engineers' Club, Albert Square, Manchester, Wednesday, October 9, 1946, at 6.30 p.m.)

MANCHESTER METALLURGICAL SOCIETY.—Dr. S. J. Kennett: "The Electron Microscope as an Aid to Metallurgy." (Engineers' Club, Albert Square, Manchester, Wednesday, October 23, 1946, at 6.30 p.m.)

SOCIETY OF CHEMICAL INDUSTRY, LONDON SECTION.—Dr. W. H. J. Vernon, O.B.E.: Chairman's Address on "Chemical Research and Corrosion Control; Some Recent Contributions of a Corrosion Research Group." (Chemical Society, Burlington House, Piccadilly, London, W.1, Monday, October 7, 1946, at 6.30 p.m.)

ADVERTISEMENTS

During the war years, advertisements in this *Journal* have had to be restricted drastically because of shortage of paper. In consequence, many advertisers were unable to obtain space.

The paper situation has now improved considerably, and intending advertisers should communicate with the Institute's agents, Messrs. T. G. Scott and Son, Ltd., Talbot House, Arundel Street, London, W.C.2, for particulars of rates, &c.

APPOINTMENTS REQUIRED AND VACANT.

METALLURGICAL ENGINEER AND TECHNICAL SALES MANAGER offers his services in London, export or overseas travelling, and is fully experienced in non-ferrous alloys from raw materials to finished articles. 15 years' works experience, 12 years' sales experience home and abroad. Box No. 174, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

ANALYTICAL CHEMIST for Non-Ferrous Works Laboratory in London area. May be required to do shift work. Must have at least Inter B.Sc. Apply, giving particulars of age, experience, and salary required, to Box No. 179, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

APPLICATIONS are invited for the position of Chief Metallurgist. Applicants must have practical experience in aircraft light alloys and steels, and their heat and protective treatments. A good knowledge of works' practice is required, which should include fusion and resistance welding. A fully qualified metallurgist is preferred but is not essential. Applications by letter to Airspeed, Ltd., Christchurch, Hants.

APPLICATIONS are invited for a Metallurgist having practical experience in aircraft light alloys and steels and familiar with their heat and protective treatments, and able to carry out routine analyses and checks on works' processes, with some experience in the examination of fusion

welds. Candidates need not be fully qualified metallurgists. Applications by letter to Airspeed, Ltd., Christchurch, Hants.

EDITORIAL ASSISTANT (JUNIOR) required for publications of a metallurgical society. Knowledge of metallurgy and of German essential, but previous editorial experience is not necessary. Box No. 178, Institute of Metals, Grosvenor Gardens, London, S.W.1.

GRADUATE wanted with knowledge of metallurgy or chemistry, for executive position involving correspondence and interviewing in a Metal Spraying Company in West London. Salary £500-£700 p.a. Box No. 177, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

LEAD SHEET AND PIPE MANUFACTURERS—WORKS MANAGER required London area; knowledge of improving and refining an advantage. Write fully, stating age, experience, and salary required. Box No. 175, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

MANAGER required to take charge of a non-ferrous Wire Mill at Birmingham Works of Henry Wiggin & Co., Ltd. Age not to exceed 40 years. Excellent position for experienced man with good organizing ability. Apply, stating experience and salary required, to Managing Director, Wiggin Street, Birmingham 16.

ATOMIC THEORY for Students of Metallurgy

By

WILLIAM HUME-ROTHERY, F.R.S.

*Royal Society Warren Research Fellow ; Lecturer
in Metallurgical Chemistry, University of Oxford*

294 pages, with 124 illustrations, $5\frac{1}{2} \times 8\frac{1}{2}$

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THIS book, which is No. 3 of the Monograph and Report Series of the Institute of Metals, provides in admirable detail almost all that is necessary to allow the student to appreciate the fundamental basis of metal theory, and to acquire a sufficient knowledge of the modern approach to follow subsequent developments in the subject.

EVERY Member, Student Member and Associate of the Institute is entitled to one free copy of this Monograph, on application to the Editor.

CONTENTS :

General Background.
Structure of the Free
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Assemblies of Atoms.
Free Electron Theory
of Metals.

Brillouin-Zone Theory of
Metals.

Electrons, Atoms, Metals
and Alloys.

INSTITUTE OF METALS

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NOTICE TO AUTHORS OF PAPERS

1. Papers will be considered for publication from non-members as well as from members of the Institute. They are accepted for publication in the *Journal*, and not necessarily for presentation at any meeting of the Institute, and should be addressed to The Editor of Publications, The Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.
2. Papers suitable for publication may be classified as :
 - (a) Papers recording the results of original research ;
 - (b) First-class reviews of, or accounts of, progress in a particular field ;
 - (c) Papers descriptive of works methods, or recent developments in metallurgical plant and practice.
3. Manuscripts and illustrations must be submitted in duplicate. MSS. must be typewritten (double-line spacing) on one side of the paper only, and authors are required to sign a declaration that neither the paper nor a substantial part thereof has been published elsewhere. MSS. not accepted will be returned within 6 months of receipt.
4. Synopsis. Every paper must have a synopsis (not exceeding 250 words in length), which, in the case of results of research, should state its objects, the ground covered, and the nature of the results. The synopsis will appear at the beginning of the paper and will also be printed in the *Journal* as soon as possible after the final decision of the Publication Committee to accept the paper.
5. References must be collected at the end of the paper, and must be numbered. Initials of authors must be given, and the Institute's official abbreviations for periodical titles (as used in *Met. Abs.*) should be used where known. References must be set out in the style :
 1. W. Hofmann and W. Jäniche, *Z. Metallkunde*, 1936, 28, 1 (i.e. year, volume, page).
6. Illustrations. Each illustration must have a number and description ; only one set of numbers must be used in one paper. The set of *line figures* sent for reproduction must be drawn in Indian ink on smooth white Bristol board, tracing cloth, or good-quality drawing paper. Co-ordinate paper is not desirable, but if used must be blue-lined with the co-ordinates to be reproduced *finely* drawn in Indian ink. All lettering and numerals, &c., must be in *pencil*. Figures should be drawn approximately twice the size intended for reproduction. *Photographs* must be restricted in number, owing to the expense of reproduction, and must be trimmed to the smallest possible of the following sizes, consistent with adequate representation of the subject : 3 in. deep by 4 in. wide (two photomicrographs to a plate) ; 3 in. deep by 2½ in. wide (four to a plate) ; 2 in. deep by 2½ in. wide (six to a plate). Magnifications of photomicrographs must be given in each case. Photographs for reproduction should be loose, not pasted down, and the figure number should be written on the back of each. Lengthy descriptions to photomicrographs should be avoided where possible, owing to the very limited space available on the plates.
7. Tables or Diagrams. Results of experiments, &c., may be given in the form of tables or figures, but (unless there are exceptional reasons) not both.
8. Overseas Authors. Authors resident in countries distant from Great Britain are requested to name, if possible, agents in Britain to whom may be referred matters concerning their papers, including proofs for correction. Translations from foreign languages should preferably be accompanied by a copy of the MS. in the language of the author.
9. Reprints. Individual authors are presented with 50, two authors with 70, and three with 90, reprints (in cover) from the *Journal*. Additional reprints can be supplied at rates to be obtained from the Editor.

OBSERVATIONS ON THE RATE AND MECHANISM OF RECRYSTALLIZATION IN COPPER.*

1034

By MAURICE COOK,† D.Sc., Ph.D., MEMBER, and T. LL. RICHARDS,‡
B.Sc., Ph.D., MEMBER.

SYNOPSIS.

The rates of recrystallization in copper strip on annealing at 27°, 50°, 72°, 100°, and 125° C. have been studied using the same material as that employed in a previous investigation on the self-annealing of copper, namely, strip with initial grain-sizes of 0.015–0.02, 0.025, and 0.06 mm., cold rolled with reductions of 80, 90, 95, and 97.5%. The results indicate that the softening of heavily cold-rolled strip takes place by a two-stage activated process: recovery followed by recrystallization. The process occurs both in copper strip with twin textures which anneals to a single-texture structure, and in copper in which the twin textures recrystallize independently, the activation energy of the former being approximately two-thirds that of the latter. It appears that when recrystallization takes place by a merging of the twin textures the energy of lattice distortion due to the twin relationship contributes one-third of the activation energy required. At the temperatures considered, *i.e.* 27°–125° C., the size of the recrystallization nucleus varies from about 3.5 to 5.5×10^{-6} cm. When strip is cold rolled with reductions which decrease the initial mean diameter of the crystals to thicknesses less than these values the material recrystallizes by a merging of the cold-rolled twin textures, the critical thickness to which it is necessary to reduce the crystals for this to happen being directly proportional to the absolute annealing temperature.

INTRODUCTION.

THE conditions under which cold-rolled copper strip recrystallizes and softens at ordinary temperature, *i.e.* self-anneals, were established in a previous investigation,¹ in the course of which observations were made on the changes in structure and properties which occurred over prolonged periods. Recrystallization and softening at room temperature occur only in cold-rolled strips which recrystallize to a single-texture structure when annealed in the normal way. In the cold-rolled condition the metal possesses a twin-fibre texture in which the crystals have been elongated into long, flat strips orientated with the (110) plane parallel to the strip surface and either the [112] or [1 $\bar{1}$ 2] axis parallel to the rolling direction. The double-fibre textures are not true crystallographic twins but bear a twinned relationship, the common twinning plane being the octahedral or (111) plane normal to the strip surface

* Manuscript received January 29, 1946.

† Director, Metals Division, I.C.I., Ltd., Witton, Birmingham 6.

‡ Research Physicist, Metals Division, I.C.I., Ltd., Witton, Birmingham 6.

and parallel to the rolling direction. During self-annealing the structure is completely reoriented to form the single-texture structure in which a $\{100\}$ plane is parallel to the strip surface and a $\langle 100 \rangle$ axis in the rolling direction, the mechanism of this recrystallization at room temperature being similar in character to that which takes place at high temperature and differing only in the rate at which it proceeds. The formation of the single-texture structure on annealing appears to be the result of interaction between the two twin-fibre textures present in the rolled strip, the process being governed not only by the annealing temperature but also by the average thickness of the crystals in the strip.^{1,2} If the average grain thickness is greater than a certain critical amount then recrystallization takes place within each grain independently and a double-texture structure is formed. The slow rate of the self-annealing process at room temperature enables the changes in structure to be examined in detail; the variations in mechanical properties of the strip closely follow the changes in structure. The present investigation was undertaken to establish the precise quantitative relationship between the rate of the recrystallization process and such factors as the initial grain-size of the strip, the degree of cold rolling, and the annealing temperature.

With this end in view, isothermal annealing curves have been determined by diamond pyramid hardness tests on H.C. copper strips with three different grain-sizes, cold rolled with reductions ranging from 80–98.75%, and maintained for prolonged periods at controlled temperatures ranging from 27°–125° C.

EXPERIMENTAL.

The copper used was identical with that used in the previous investigation,¹ its composition being copper 99.95, oxygen 0.044, silver 0.003, and iron 0.001%. Three lots of strip were prepared according to the rolling scheme detailed in Table I of the previous paper,¹ with a thickness of 0.67 in. and grain-sizes of *A*, 0.015–0.02, *B*, 0.025, and *C*, 0.06 mm., respectively. From this stock material strips were cold rolled as required, with total reductions of (1) 80, (2) 90, (3) 95, (4) 97.5, and (5) 98.75%, respectively, of the initial thickness of 0.67 in. The samples were numbered with a letter and number indicating the initial grain-size and the final rolling reduction, respectively. Thus, sample *B*3 had an initial grain-size of 0.025 mm. at a thickness of 0.67 in. and was cold rolled with a reduction in thickness of 95% to a final thickness of 0.033 in.

The rate of recrystallization was followed by diamond pyramid

TABLE I.—Estimated Degree of Recrystallization and Corresponding Hardness Values.

Diamond Pyramid Hardness Numbers (10 kg. load).	Fraction Recrystallized.
122	0.0
121	0.02
116	0.15
94	0.45
73	0.67
64	0.82
55	0.92
49	0.99

hardness tests and the relation between hardness and degree of recrystallization was determined from the results given in the previous

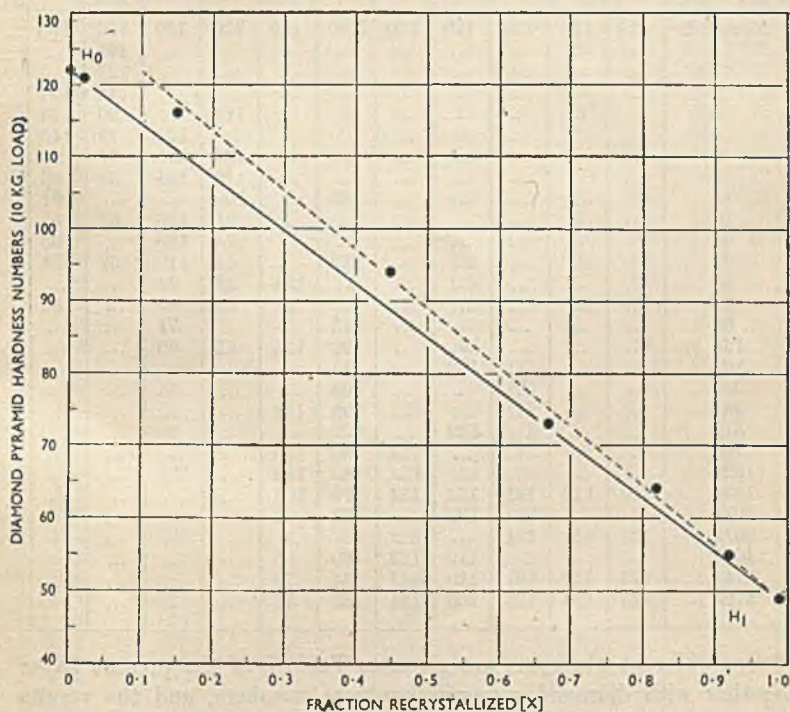


FIG. 1.—Relationship between Hardness and Degree of Recrystallization.

$$\text{— } x = \frac{H_0 - H_x}{H_0 - H_1}$$

$$\text{--- } x = 0.1 + 0.9 \frac{H_0 - H_x}{H_0 - H_1}$$

paper.¹ The fraction of the strip A4 recrystallized at a particular stage in the process of self-annealing was determined by visual estimation of the proportion of the total area of the corresponding micrograph occupied by the white recrystallized regions (see Figs. 7-21 of previous work¹ showing typical microstructures of samples A4 at various stages of self-annealing). The mean values of the estimates

TABLE II.—*Diamond Pyramid Hardness Values of Strips A1, B1, and C1.*

Time, hr.	Annealing Temp., ° C.										
	27°			50°		72°		100°		125°	
	A.	B.	C.	A.	B.	A.	B.	A.	B.	A.	B.
As rolled.	120	120	120	120	120	120	120	120	120	120	120
1	122	...
1.5	122	...
2	118	124
3	122	...	90	123
4	124	79	115
6	122	...	57	94
8	119	124	54	82
12	122	...	100	61
16	91	125	52	57
24	74	120	...	55
32	122	...	64	110	52	54
48	124	53	94
64	52	89	...	54
96	122	71
128	120	124	52	60
192	122	...	111
256	108	55
384	100	124
512	122	54
768	124	90	115
1024	82	112
1536	120	119	121	122	124	76	103
2048	120	...	73	98
3072	121	121	121
4096	119	123	60	88
6144	122	122	121	116	...	54	79
8192	123	123	123	109	121	53	74

of five different observers are given in Table I of the present paper together with diamond pyramid hardness numbers, and the results are plotted in Fig. 1.

The specimens were held at thermostatically controlled temperatures of 27°, 50°, 72°, 100°, and 125° C. and quenched in cold water after appropriate periods. Annealing at temperatures of 27°, 50°, and 72° C. was carried out in controlled-air-atmosphere ovens and annealing at 100° and 125° C. in controlled-temperature water- and oil-baths,

respectively. For all test specimens, therefore, the heating and cooling times were short in comparison with the time at temperature and the annealing period was known precisely. The results of diamond pyramid hardness tests are recorded in Tables II-VI and one typical

TABLE III.—*Diamond Pyramid Hardness Values of Strips A2, B2, and C2.*

Time, hr.	Annealing Temp., ° C.										
	27°			50°		72°		100°		125°	
	A.	B.	C.	A.	B.	A.	B.	A.	B.	A.	B.
As rolled.	122	123	124	122	123	122	123	122	123	122	123
0.125	122	123
0.25	122	123
0.375	122	...	102	123
0.5	85	123
0.75	61	116
1	122	...	58	92
1.5	120	...	56	64
2	118	123	55	61
3	90	...	54	57
4	122	...	71	123	...	56
6	57	114	54	...
8	123	56	89	...	56
12	122	...	54	64
16	120	57
24	105	123	54	56
32	122	...	90
48	68	56
64	61	123
96	122	...	58	120
128	120	123	56	110
192	106	...	54	76
256	97	62
384	122	83	123	54	58
512	74	57
768	59	56
1024	122	55	123
1536	54	121	...	56
2048	115
3072	122	54	101
4096	120	91
6144	97	75
8192	85	124	125	...	66

set of results, namely those from Table V referring to samples A4, B4, and C4, is plotted in Fig. 2.

Isothermal annealing treatments on the A and B strips were continued up to the point of complete softening excepting when the annealing period was inordinately long. The C strips showed no sign

of softening over periods in which softening in the *A* and *B* strips was complete (Fig. 2). The isothermal annealing treatments of the *C* strips in the temperature range 27°–125° C. were, therefore, discontinued.

TABLE IV.—*Diamond Pyramid Hardness Values of Strips A3, B3, and C3.*

Time, hr.	Annealing Temp., ° C.										
	27°			50°		72°		100°		125°	
	A.	B.	C.	A.	B.	A.	B.	A.	B.	A.	B.
As rolled.	124	125	130	124	125	124	125	124	125	124	125
0-0312
0-0468
0-0625	124	...
0-0937	104	...
0-125	124	...	90	...
0-1875	57	125
0-25	124	...	51	114
0-375	123	...	50	76
0-5	117	125	...	60
0-75	124	...	100	...	50	53
1	88	125	...	52
1-5	124	...	124	125	68	120
2	58	110	...	52
3	124	...	124	...	51	94
4	124	125	50	84
6	118	70
8	124	...	104	125	50	61
12	89	124	...	53
16	72	122	...	52
24	124	124	125	57	108
32	123	...	54	98	...	52
48	116	125	...	76
64	124	104	...	50	67
96	86	125	...	55
128	77	123	50	53
192	124	125	...	61	110	...	52
256	57	100
384	52	82	...	52
512	124	125	...	51	72
768	122	50	56
1024	110	53
1536	80	125	...	50	52
2048	64	125
3072	53	122	52
4096	51	116
6144	50	94
8192	50	83	131

The initial and final hardnesses of the *C* strips are, accordingly, given in only one column of Tables II–VI.

Analysis of the results of these tests raised a number of points which

could only be clarified by further subsidiary experiments. These latter included isothermal annealing tests at various temperatures up to 250° C. on strips of the C series carried out over the necessary periods to

TABLE V.—*Diamond Pyramid Hardness Values of Strips A4, B4, and C4.*

Time, hr.	Annealing Temp., ° C.										
	27°			50°		72°		100°		125°	
	A.	B.	C.	A.	B.	A.	B.	A.	B.	A.	B.
As rolled.	124	126	128	124	126	124	126	124	126	124	126
0-0312	124	...
0-0468	122	126
0-0625	124	126	112	122
0-0937	105
0-125	124	126	70	96
0-1875	48	80
0-25	124	126	47	70
0-375	106	123	46	53
0-5	88	110	...	48
0-75	124	126	66	88	46	...
1	50	74	...	48
1-5	46	52
2	124	126	...	48
3	122	...	46
4	108	126	...	48
6	124	126	80	123
8	66	116
12	48	86
16	124	126	46	69
24	124	119	53
32	107	126	46	50
48	80	124	...	48
64	124	126	...	64	120
96	49	97	...	48
128	47	82
192	124	126	...	46	52
256	123	48
384	112	46	48
512	102	126
768	81	126	48
1024	70	121
1536	50	103
2048	48	94
3072	46	76
4096	...	66
6144	...	52
8192	46	50	128

give complete softening, structural observations on samples annealed normally at temperatures up to 900° C., and finally, a brief examination of the effect of rapid heating. The subsidiary experiments are described

as they arise in the discussion of results. The results of the main experimental work which are set out in Tables I-VI and Figs. 1 and 2

TABLE VI.—*Diamond Pyramid Hardness Values of Strips A5, B5, and C5.*

Time, hr.	Annealing Temp., ° C.										
	27°			50°		72°		100°		125°	
	A.	B.	C.	A.	B.	A.	B.	A.	B.	A.	B.
As rolled.	122	124	125	122	124	122	124	122	124	122	124
0-0156
0-0234	122	...
0-0312	121	124
0-0468	108	123
0-0625	94	121
0-0937	51	90
0-125	122	...	46	80
0-1875	121	...	42	49
0-25	122	...	116	124	40	46
0-375	92	116	...	44
0-5	80	106	40	42
0-75	122	124	59	85
1	47	76	...	42
1-5	41	55
2	122	124	40	47
3	106	123	...	43
4	96	120	40	42
6	122	124	68	104
8	54	95	...	42
12	42	73
16	122	124	40	60
24	106	45
32	87	124	40	42
48	122	50	118
64	43	100	...	42
96	...	124	...	41	62
128	40	50
192	122	43
256	115	124	...	40	42
384	86	119
512	61	107	42
768	43	85
1024	41	73
1536	40	55
2048	...	44
3072	40	42
4096
6144
8192	...	42	125

indicate that a simple linear relationship exists between diamond pyramid hardness and the fraction of metal recrystallized.

The isothermal annealing curves, of which Fig. 2 is typical, are a family of curves, all of the same form when hardness on a linear scale is plotted against the annealing time on a logarithmic or G.P. scale. It

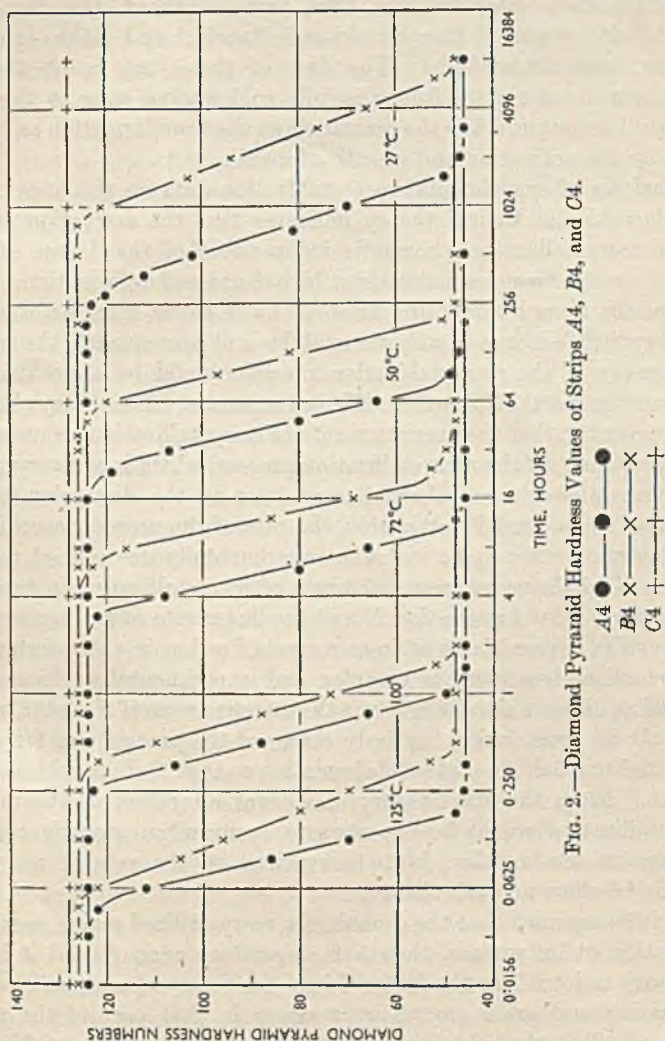


FIG. 2.—Diamond Pyramid Hardness Values of Strips A4, B4, and C4.

should, therefore, be possible to develop a general theory correlating the rate of recrystallization and such factors as initial grain-size, degree of cold rolling, and annealing temperature. The basis of such a theory is discussed in the next section.

THEORETICAL.

Since Alterthum³ in 1922 suggested a thermodynamical theory of recrystallization other theories have been advanced, the two most recent being those of Krupkowski and Balicki,⁴ and Mehl, Johnson, Stanley, and Anderson.⁵⁻⁸ The first of these two latest theories postulates direct change from the fully cold-worked state to the fully recrystallized state, while the second views the transformation as taking place by the activation and growth of nuclei.

Analysis of experimental recrystallization data on the lines of the Krupkowski and Balicki theory indicates that the activation energy of the recrystallization process is independent of the degree of cold rolling, except for some materials at high degrees of deformation. This part of the theory would appear to be quite sound since deformation has very little effect on the elastic modulus and consequently the activation energy of the recrystallization process should be approximately the same for metal deformed by different amounts. The theory also predicts, however, that the maximum rate of recrystallization occurs at the commencement of the recrystallization process, which is contrary to the experimental evidence. As indicated later in the discussion of the results of the present investigation, the rate of the process according to the theory does not agree with the experimentally determined results.

The Mehl theory expresses the rate of recrystallization in terms of two factors, rate of nucleation N and the linear rate of grain growth G . Although it is possible to express recrystallization in these terms, the mathematical treatment is complex and experimental verification is somewhat difficult since it involves the measurement of N and G , which can only be done during the early stages of the process and for metal subjected to small amounts of deformation so that there is critical grain growth. From the Mehl theory an abrupt alteration in the rate of recrystallization would be expected to occur when growing crystals impinge on one another, but observations on the rate of recrystallization⁶⁻⁸ show no such change.

If it be assumed that the growth of a recrystallized region occurs by nucleation at its surface, then G is dependent upon N and it is not necessary to introduce the factor G into the theory of recrystallization. Nucleation and grain growth must occur in that part of the metal which is still in the cold-worked state and the rate of recrystallization at any particular stage will be dependent on (i) the fractional amount of metal still in the cold-worked state and (ii) the rate of nucleation of the sample at that stage. The rate of nucleation may be different at points in the cold-worked regions well removed from the

surface of recrystallized regions and at points near to this surface. It would then be necessary to express the rate of recrystallization by two terms, first, the product of the rate of nucleation at the interface between recrystallized and cold-worked areas and the fraction of the metal still in the cold-worked state influenced by such interfaces and, second, the product of the rate of nucleation well inside cold-worked regions and the fraction of metal in these regions not influenced by the interface conditions. Such an approach may be essential in order to deal with the rate of recrystallization of specimens subjected to small amounts of deformation when the two rates of nucleation are probably very different.

For specimens subjected to considerable amounts of deformation it is likely that the two rates approach the same value and the rate of recrystallization can then be expressed by one term only, namely the product of a single rate of nucleation and the fraction of metal still in the cold-worked state. While this argument appears reasonable for any heavily worked metal, it should be particularly applicable to recrystallization in heavily worked copper specimens which self-anneal at room temperature. In these specimens recrystallization appears to take place by a merging of neighbouring cold-rolled crystals which have a twinned relationship to form a single-texture structure, and the rate of the process is determined by this relationship rather than by the particular conditions existing at the recrystallized surfaces. Even in this special instance, however, there is probably some difference in the rate of nucleation, during the early stages of the transformation, at the surface of recrystallized regions and at points well within the cold-worked regions since the first-formed recrystallized regions grow more rapidly than fresh regions appear (see Figs. 9 and 11 of the previous paper¹), but at later stages in the process the difference in rates is imperceptible (see Figs. 13, 15, and 17 of the previous paper¹).

Another effect which any comprehensive theory of recrystallization must take into account is the slowness of the process in the initial stages since nearly all observations of changes in properties associated with recrystallization indicate the presence of an incubation period, during which no observable change in properties occurs. It is well known⁹ that a recovery process takes place before recrystallization but it has hitherto been assumed that this recovery is an independent process which has no effect upon recrystallization. Consideration of the results of the present work indicates that recovery is a necessary preliminary to recrystallization.

It is considered that the lattice of copper in the severely cold-worked condition is highly distorted and has a high distribution-density

of atomic dislocations. Recovery or reduction of lattice distortion will occur by a reduction in the density of distribution of atomic dislocations. This process will take place by the activation of a dislocation, so that there is some local atomic rearrangement in which the dislocation is displaced by a simple or complex slip mechanism until it is halted by a neighbouring dislocation of the same sign, or merges with a neighbouring dislocation of opposite sign so that both vanish. Recovery then may be looked upon as a process of atomic rearrangement of very restricted range.

When recovery is well advanced there will be a much sparser distribution of atomic dislocations, and if one of the remaining dislocations is now activated a second process of atomic rearrangement will be initiated which will spread through the cold-worked metal so that a recrystallized nucleus is formed of a size restricted more by the damping of the process in its passage through the matrix than by the halting effect of dispersed dislocations. The two processes are similar atomic rearrangements, each involving the initial activation of a small group of atoms but differing in the range over which they occur. The rate of the second process is governed by the extent to which the first process has progressed. This view is supported by the work of Stanley⁸ on silicon ferrite in which it was shown that prior recovery of deformed material accelerates the recrystallization process. In order to give a mathematical expression for the rate of recrystallization it may be assumed that the rates of the two processes of recovery and recrystallization follow the natural exponential law of change. If w is the fraction of the cold-worked metal remaining in the unrecovered state at a time t and β is the recovery rate, then :

$$\frac{dw}{dt} = -\beta w \quad . \quad . \quad . \quad . \quad . \quad . \quad (1)$$

and : $w = \exp. (-\beta t) \quad . \quad . \quad . \quad . \quad . \quad . \quad (2)$

and the fraction of recovered material or degree of recovery :

$$1 - w = 1 - \exp.(-\beta t) \quad . \quad . \quad . \quad . \quad . \quad . \quad (3)$$

or, when βt is small :

$$1 - w = \beta t \quad . \quad . \quad . \quad . \quad . \quad . \quad (4).$$

Since the process is activated, β is dependent on temperature and equal to $b \cdot \exp. \left(-\frac{P}{RT} \right)$, where b is a constant the value of which varies with the initial grain-size and extent to which the copper has been cold worked, P is the activation energy of the recovery process, R is the gas constant, and T is the absolute temperature. Then :

$$1 - w = bt \exp. \left(-\frac{P}{RT} \right) \quad . \quad . \quad . \quad . \quad . \quad . \quad (5).$$

The recovery process might cause a slight change in hardness, which, in certain instances, may be a hardening, or, in others, a softening. In fully work-hardened material it is possible that no change in hardness occurs.

If it be assumed that the second stage of the recrystallization process is similar in character to the first stage and that the rate $\left(\frac{dx}{dt}\right)$ of recrystallization is proportional to the fractional amount $(1 - x)$ of the copper remaining to be recrystallized and to the degree of recovery $(1 - w)$ of the unrecrystallized copper, then :

$$\frac{dx}{dt} = a(1 - w)(1 - x) \exp.\left(-\frac{Q}{RT}\right) \quad \dots \quad (6)$$

where Q is the activation energy of the second stage of the process and a is another constant similar to b of equation (5). Equation (6) can only truly represent the rate of recrystallization when the rate of nucleation at points well inside the regions still in the cold-worked state is equal to that at the surface of recrystallized regions. If, however, it be assumed that the initial recovery occurs in patches and that recrystallization occurs in fully recovered patches, then the recovery term $(1 - w)$ in equation (6) represents the fraction of the cold-worked metal which is fully recovered and the observation that in the early stages of recrystallization preferential grain growth occurs in already recrystallized areas is explained. On this basis the adoption of equation (6) would appear to be justified. From equations (5) and (6) :

$$\frac{dx}{dt} = a \cdot b \cdot t(1 - x) \exp.\left(-\frac{P}{RT}\right) \exp.\left(-\frac{Q}{RT}\right)$$

or, substituting $c = a \cdot b$:

$$\frac{dx}{dt} = c \cdot t \cdot (1 - x) \exp.\left(-\frac{P + Q}{RT}\right) \quad \dots \quad (7).$$

The equation given by Krupkowski and Balicki ⁴ for a single-stage process is :

$$\frac{dx}{dt} = c(1 - x) \exp.\left(-\frac{Q}{RT}\right) \quad \dots \quad (8).$$

The form of equation (7) differs from that of equation (8) only by the inclusion of the term t . Equation (7) can be rewritten :

$$t dt = \frac{1}{c} \cdot \frac{dx}{1 - x} \exp.\left(\frac{P + Q}{RT}\right)$$

Integrating :

$$t^2 = \frac{2}{c} \cdot \log \frac{1}{1 - x} \exp.\left(\frac{P + Q}{RT}\right)$$

and:

$$\log t = \frac{1}{2} \log \cdot \log \frac{1}{1-x} + \frac{1}{2} \log \frac{2}{c} + \frac{1}{2} \left(\frac{P+Q}{RT} \right) \quad (9).$$

For a single-stage process, the rate of which is defined by equation (8):

$$\log t = \log \cdot \log \frac{1}{1-x} + \log \frac{1}{c} + \frac{Q}{RT} \quad (10).$$

Equations (9) and (10) show that $\log t$ in the isothermal annealing of any particular cold-worked sample varies with $\log \cdot \log \frac{1}{1-x}$, which means that isothermal curves should have the same shape and thus coincide when displaced along the $\log t$ scale. Wever,¹⁰ Tammann,¹¹ and Upton¹² have already drawn attention to this singular feature of isothermal annealing curves.

$\log t$ in both equations (9) and (10) is a linear function of $\frac{1}{T}$ for fixed values of x and c , although the same experimental observations would yield a value for the activation energy $P+Q$ from equation (9) twice that of Q from equation (10). Linearity of $\log t$ when plotted against $\frac{1}{T}$ is then no criterion for determining which equation is valid and whether, therefore, recrystallization is in fact a single-stage or a two-stage process. A better method is to examine isothermal annealing curves to see if $\log t$ is given by $\log \cdot \log \frac{1}{1-x}$ or by $\frac{1}{2} \log \cdot \log \frac{1}{1-x}$.

DISCUSSION OF RESULTS.

A.—*Relation between Diamond Pyramid Hardness and Fraction of Metal Recrystallized.*

The observations recorded in Table I and plotted in Fig. 1 indicate that diamond pyramid hardness varies in an approximately linear manner with the proportion of metal recrystallized. The points in Fig. 1 approximate reasonably closely to the full straight line connecting the hardness value of non-recrystallized to that of fully recrystallized metal but they are in somewhat better agreement with the dotted straight line; one possible explanation of this is that initial recovery causes a slight increase in the diamond pyramid hardness value of the cold-worked state to about 130.

B.—*Incubation.*

The results of isothermal tests are given in Tables II–VI, those from Table V being plotted in Fig. 2. These data indicate that there

is an incubation period during which no apparent recrystallization occurs. It should be remembered, however, that the time is plotted on a G.P. scale and in all cases the initial period of no observable change in hardness is only about one-quarter the annealing period required to attain complete recrystallization. In addition, it would seem from the analysis in the preceding section that a certain small amount of recrystallization must occur before there is an observable change in hardness. Microscopic observations in a previous investigation¹ revealed the onset of recrystallization in *A4* at room temperature in one week after rolling, while a measurable change in hardness occurred only after about eight weeks.

As indicated above, the question whether recrystallization is a single-stage or a two-stage process can be answered by determining whether $\log t$ is given by $\log \cdot \log \frac{1}{1-x}$ or by $\frac{1}{2} \log \cdot \log \frac{1}{1-x}$, or more conveniently by examining how the theoretical curves hardness- $\log \cdot \log \frac{1}{1-x}$ or hardness- $\frac{1}{2} \log \cdot \log \frac{1}{1-x}$ compare with the experimental curves hardness- $\log t$.

It has already been established in Section A that the hardness H varies in an approximately linear manner with the proportion of copper recrystallized. If H_0 and H_1 are the initial and final hardness values and H_x the hardness when a proportion x of the material is recrystallized, then x is given by either :

$$x = \frac{H_0 - H_x}{H_0 - H_1} \quad . \quad . \quad . \quad . \quad . \quad (11)$$

corresponding to the full line in Fig. 1,
or :

$$x = 0.1 + 0.9 \frac{H_0 - H_x}{H_0 - H_1} \quad . \quad . \quad . \quad . \quad . \quad (12)$$

corresponding to the dotted line of Fig. 1.

The values of x for various hardnesses have been calculated from both expressions (11) and (12) and these are given in Table VII. The corresponding $\log \cdot \log \frac{1}{1-x}$ and $\frac{1}{2} \log \cdot \log \frac{1}{1-x}$ values are also recorded, these being converted to the logarithm to base 4 scale for direct comparison with the experimental observations. The theoretical curves (a) hardness- $\log_4 \cdot \log \frac{1}{1-x}$, (b) hardness- $\frac{1}{2} \log_4 \cdot \log \frac{1}{1-x}$ [x determined from expression (11)], and (c) hardness-

$\frac{1}{2} \log_4 \cdot \log \frac{1}{1-x}$ [x determined from expression (12)] are plotted in Fig. 3, together with the experimental curves (d) and (e) for sample A4 annealed isothermally at 50° and 27° C., respectively.

The form of slope of curves (b) and (c) is seen to be in reasonable agreement with the experimental curves (d) and (e) while the slope of curve (a) is far removed from that of the experimentally determined curves (d) and (e). It follows, therefore, that $\log t$ can best be expressed as:

$$\log t = \frac{1}{2} \log \cdot \log \frac{1}{1-x} + \text{constant},$$

TABLE VII.—*Calculated Values of x for Various Hardnesses.*

Theoretical.						Experimental.			
Diamond Pyramid Hardness Numbers.	x for curves (a) & (b).	x for curve (c).	$\log_4 \frac{1}{1-x}$	$\frac{1}{2} \log_4 \frac{1}{1-x}$	$\frac{1}{4} \log_4 \frac{1}{1-x}$	Diamond Pyramid Hardness Numbers.		Anneal- ing Time, hr.	$\log_4 t, ^\circ$
			$\log \frac{1}{1-x}$	$\log \frac{1}{1-x}$	$\log \frac{1}{1-x}$	Curve (d). A4 Annealed 50° C.	Curve (e). A4 Annealed 27° C.		
124	0	0.100	0	124	...	16	0
123	0.0128	0.112	0	0	0.0416	119	...	24	0.293
122	0.0256	0.123	0.505	0.253	0.0795	107	...	32	0.500
120	0.0513	0.146	1.014	0.507	0.146	80	...	48	0.792
110	0.180	0.262	1.970	0.985	0.381	64	...	64	1.000
100	0.308	0.377	2.472	1.236	0.541	49	...	96	1.292
90	0.436	0.492	2.735	1.368	0.672	47	...	128	1.500
80	0.564	0.608	3.004	1.502	0.788	46	124	192	1.793
70	0.692	0.723	3.257	1.629	0.902	...	123	256	2.000
60	0.821	0.839	3.529	1.765	1.028	...	112	384	2.293
50	0.949	0.954	3.922	1.961	1.217	...	102	512	2.500
48	0.974	0.977	4.074	2.037	1.289	...	81	768	2.793
47	0.987	0.988	4.199	2.100	1.351	...	70	1024	3.000
46	1	1	50	1536	3.293
							48	2048	3.500
							46	3072	3.792

* The first recorded value is placed at zero of the \log_4 scale.

which means that recrystallization is a two-stage process, the rate of which is given by equation (7). The observed incubation can be regarded as associated with an initial recovery or removal of lattice distortion before the process of activation of recrystallization nuclei can occur. The initial recovery may be accompanied by a slight hardening of the metal in the cold-worked state, which would accentuate the incubation effect.

C.—*The Activation Energy $P + Q$.*

From the analysis for equation (9) in the theoretical section and from the above discussion of the experimental data it is clear that for any

arbitrary constant value of x and for a particular cold-worked state, $\log t$ is a linear function of the reciprocal of the absolute annealing temperature T .

The time taken for each sample to reach a hardness halfway between

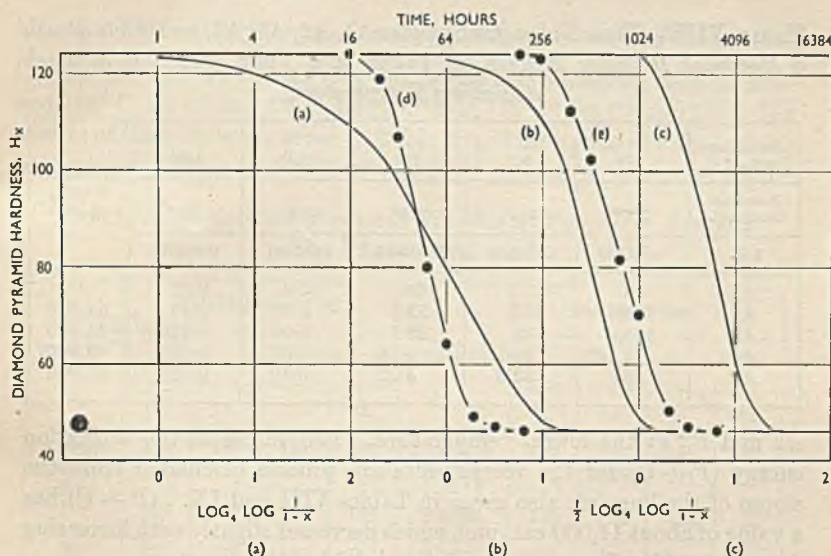


FIG. 3.—Theoretical and Experimental Curves.

Theoretical Curves.

$$(a) H_x = \log_4 \cdot \log \frac{1}{1-x}$$

$$(b) H_x = \frac{1}{2} \log_4 \cdot \log \frac{1}{1-x}$$

x from equation (11)

$$(c) H_x = \frac{1}{2} \log_4 \cdot \log \frac{1}{1-x}$$

x from equation (12)

Experimental Curves.

(d) A4 annealed 50° C.

(e) A4 annealed 27° C.

the initial and final hardness at each annealing temperature has been interpolated from curves similar to those in Fig. 2. These times, in hours, are recorded in Tables VIII and IX and are plotted in Fig. 4 against the reciprocal absolute annealing temperature.

It will be seen that, in agreement with the theoretical analysis, a linear relationship exists between $\log t$ and $\frac{1}{T}$ for samples A2, A3, A4, A5, and B2, B4, and B5 over the range of temperatures considered, except for a slight falling away of the points from the line for samples

TABLE VIII.—*Time Taken for Samples A1, A2, A3, A4, and A5 to Reach a Hardness Halfway Between the Initial and Final Hardness at Each Annealing Temperature.*

Annealing Temp., ° C.	27°	50°	72°	100°	125°	Activation Energy, cal./mol.
Absolute Temp., T.	300°	323°	345°	373°	398°	
1/T.	0.003333	0.003096	0.002899	0.002681	0.002513	
A1	840	17.8	3.52	...
A2	7780	333	33.3	3.14	0.46	44,400
A3	1415	98	12.1	1.00	0.125	44,600
A4	737	46.1	5.76	0.56	0.096	43,800
A5	387	33.3	4.92	0.49	0.070	43,000

A2 and B2 at the lowest temperature. The values of the activation energy ($P + Q$) for the recrystallization process calculated from the slopes of the lines are also given in Tables VIII and IX. ($P + Q$) has a value of about 44,000 cal./mol. which decreases slightly with increasing degree of cold rolling and decrease in initial grain-size.

TABLE IX.—*Time Taken for Samples B1, B2, B3, B4, and B5 to Reach a Hardness Halfway Between the Initial and Final Hardness at each Annealing Temperature.*

Annealing Temp., ° C.	27°	50°	72°	100°	125°	Activation Energy, cal./mol.
Absolute Temp., T.	300°	323°	345°	373°	398°	
1/T.	0.003333	0.003096	0.002899	0.002681	0.002513	
B1	3540	57.3	6.8	...
B2	...	4096	161	8.32	1.04	48,000
B3	7210	338	36.8	3.6	0.33	46,600
B4	2415	112	12.4	0.79	0.158	46,800
B5	837	74.2	9.68	0.82	0.110	44,400

This determination is based on observations in the isothermal annealing tests over the temperature range 27°–125° C. recorded in Tables II–VI. The C strips did not soften under these conditions and supplementary isothermal annealing tests at higher temperatures were

necessary to determine the value of the activation energy for these samples. Use was also made of the data obtainable from normal annealing curves already published in Fig. 31 of the previous paper.¹ The half-softening temperatures determined from the half-hour-annealing curves already established provided one set of points for the $\log t - \frac{1}{T}$ curves and further points were determined for each of the samples C1, C2, C3, and C4 by carrying out isothermal annealing tests at 250° and 185° C. for C1 and C2 and at 250° and 175° C. for C3 and C4. The results of these tests are given in Table X and interpolated values of the half-softening times and half-softening temperatures, determined from

TABLE X.—Diamond Pyramid Hardness Values of Samples C1, C2, C3, and C4.

Time, hr.	Annealing Temp., ° C.				Time, hr.	Annealing Temp., ° C.			
	185°		175°			250°			
	C1.	C2.	C3.	C4.		C1.	C2.	C3.	C4.
As rolled.	127	129	131	131	As rolled.	127	129	131	131
0.5	127	128	130	129	0.0083	...	128	131	123
0.75	0.0125
1	125	123	128	123	0.0162	126	124	112	87
1.5	123	114	0.025	83	64
2	122	113	120	104	0.033	124	98	71	60
3	120	101	105	90	0.050	118	77	63	57
4	116	87	94	73	0.067	100	68	61	55
5	81	67	0.100	74	63	58	54
6	98	70	...	63	0.133	62	61	57	54
8	87	64	70	58	0.200	59	58
12	70	...	63	56	0.267	58	57	55	53
13	...	59	0.400
16	64	58	61	55	0.533	56	56	54	...
24	0.800
30	58	58	57	...	1.067	56

these observations and the normal annealing curves respectively, are recorded in Table XI, and the $\log t - \frac{1}{T}$ relationship plotted in Fig. 4.

The three points for each of the four samples fall on a straight line, indicating that the recrystallization process for these is also an activated process. In addition, the slope of the isothermal annealing curves for the C samples was in excellent agreement with the theoretical curves (b) and (c) of Fig. 3 but not with curve (a), which indicates that the recrystallization of the C strips was a two-stage process similar to that occurring in the A and B strips. The theoretical analysis given earlier,

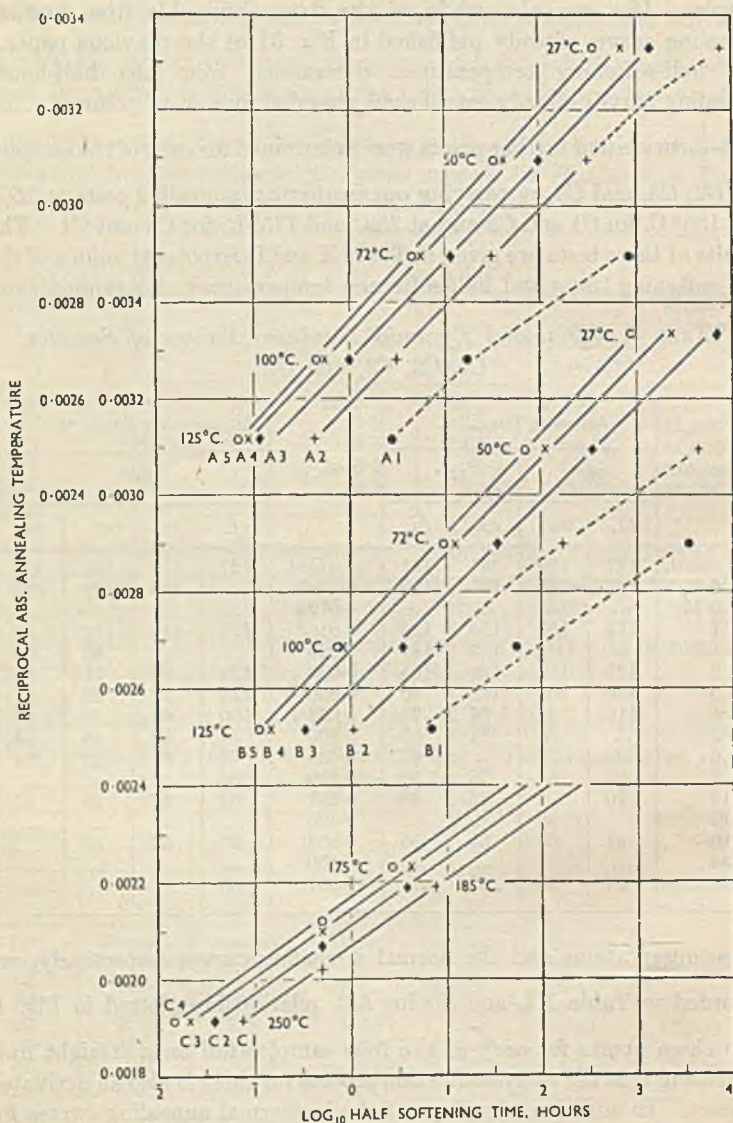


FIG. 4.—Relationship between Half-Softening Times and Reciprocal Absolute Annealing Temperature.

therefore, applies equally to both (i) heavily rolled copper strip which self-anneals in a special way and (ii) to that which does not self-anneal. The activation energy of the recrystallization process of the *C* strips was calculated from the slope of the $\log t - \frac{1}{T}$ line in Fig. 4; the values, which are recorded in Table XI, lie between 65,200 and 67,200 cal./mol. and show a slight decrease with increasing degree of cold rolling.

It is interesting at this stage, however, to note that the value for the activation energy of the recrystallization process for the *A* and *B* strips is approximately two-thirds the value for the *C* strips. The values are of the same order of magnitude as those determined for aluminium by

TABLE XI.—*Interpolated Values of Half-Softening Times and Half-Softening Temperatures.*

Annealing Temp., ° C.	175°	185°	250°	Temperature of Half-Softening from Half-Hour-Annealing Curves.		Activation Energy, cal./mol.
Absolute Temp., <i>T</i>	448°	458°	523°			
1/ <i>T</i> .	0.002232	0.002183	0.001912	Absolute Temp., <i>T</i> .	1/ <i>T</i> .	
<i>C</i> 1	...	7.3	0.078	496°	0.00202	66,600
<i>C</i> 2	...	3.7	0.039	483°	0.00207	67,200
<i>C</i> 3	4.12	...	0.0222	476°	0.00210	65,400
<i>C</i> 4	2.54	...	0.0150	473°	0.00212	65,200

Anderson and Mehl⁷ which is all that can be expected in view of the difference in the materials examined and in the degree of cold working involved. The value for the *C* strips is about twice the value determined for copper by Krupkowski and Balicki,⁴ as predicted in the theoretical analysis. The present experimental observations upon the *C* strips are, therefore, in close agreement with those of Krupkowski and Balicki⁴ on the recrystallization of copper wire cold drawn with various reductions in area.

The value of 44,000 cal./mol. has been obtained with materials which are capable of self-annealing and which recrystallize in a very special way, namely by a merging of the twin textures of the cold-rolled strip to form a complete or nearly complete single-texture structure.* Photomicrographs (see previous paper¹) of sample *A*4 at various stages in the process of recrystallization at room temperature indicate also that the recrystallized areas grow as laminae parallel to the strip surface.

* Since the publication of the previous paper¹ the first signs of self-annealing have been observed in sample *B*2, 150 weeks after cold rolling.

The larger value of 66,000 cal./mol. for the activation energy has been obtained with samples in which the twin textures of the cold-rolled strip recrystallized independently on annealing at low temperatures to form a structure composed wholly or mainly of crystals with a preferred orientation corresponding to that of one or other of the initial twin textures. *A1* and *B1* are also samples which possess a structure with a high proportion of crystals of the double-texture type when annealed at low temperatures and the slopes of the $\log t - \frac{1}{T}$ curves for these samples (see Fig. 4) approximate to those for the *C* samples, indicating that the value of the activation energy of the recrystallization process for samples *A1* and *B1* is also about 66,000 cal./mol. The falling away of the curves for samples *A2* and *B2* in the lower-temperature range is also probably due to the change in mechanism of the recrystallization process from the merging of the twin textures present in the rolled strip to the independent recrystallization of the twin textures.

The energy of activation of a process is regarded as the free energy per mol. required to form a stable nucleus. In fact, it is the experimentally determined value of the constant $(P + Q)$ of the probability factor $\exp. \left(- \frac{P + Q}{RT} \right)$, $(P + Q)$ being the product of the energy $3E$ per g. mol. per degree of freedom required for recrystallization, and the number of degrees of freedom, n , of the initially activated nucleus.

Since recrystallization and deformation are both processes involving atomic rearrangement there is probably some connection between the activation energy of recrystallization and the energy of the severest dislocation that can exist in the metal. Recrystallization can be considered as the creation of a fresh lattice to a new arrangement. The formation of a three-dimensional lattice will require three independent dislocations, each of energy E , a two-dimensional lattice two independent dislocations, and a single-dimensional lattice, or a row of atoms, only one dislocation.

The observed low value for the activation energy of recrystallization for material which self-anneals can probably be explained in either of two ways: (i) that since the recrystallization nuclei in such material appear to grow as laminæ, only two independent dislocations and consequently an activation energy of $2nE$ are necessary for the process, or (ii) which is a more likely explanation, that an activation energy $3nE$ is necessary for the initial activation of all nuclei but that in the special case of material which self-anneals by a merging of two textures which bear a twinned relationship, the energy of the lattice distortion due to the twinned relationship provides one-third of the activation energy required.

D.—The Frequency Factor c .

All the factors in equation (7), which defines the rate of the recrystallization process, have now been considered with the exception of c which may be regarded as a frequency factor. The value of c depends on the condition of the copper, namely its initial grain-size and the magnitude of the cold-rolling reduction. Equation (9) may be written :

$$\log t = \frac{1}{2} \log \frac{2}{c} + K_1 \quad . \quad . \quad . \quad . \quad (13)$$

where :

$$K_1 = \frac{1}{2} \log \cdot \log \frac{1}{1-x} + \frac{1}{2} \cdot \frac{(P+Q)}{RT}$$

and is a constant for particular values of T , x , and $(P+Q)$, the latter factor, as shown in Section C, varying slightly with the initial grain-size and amount of cold rolling. Thus $\log t$ is a linear function of $\log \frac{2}{c}$.

Without making any assumption other than that a certain definite fraction of the copper is recrystallized when the partially recrystallized strip has a hardness halfway between the initial and final hardness, then the relationship between c and the rolling reduction can be partly elucidated by plotting $\log t$ values for the half-softening times against the percentage rolling reduction.

The values of t for the A and B strips are already given in Tables VIII and IX and the values of $\log t$ are plotted against the percentage rolling reduction in Figs. 5 and 6 respectively. With both series of samples a linear relationship is observed between $\log t$ and the percentage reduction, although the slopes of the lines for the separate series are not directly comparable. This difference probably arises from the difference in initial grain-size of the A and B samples.

In order to correlate observations on the A and B series the most precise determination of their initial grain-sizes is necessary. The grain-size values quoted initially were determined by comparison with A.S.T.M. standards and are accurate only to the nearest 0.005 mm. A more exact determination of the initial grain-size was carried out and the values as estimated by five different observers were 0.018 and 0.025 mm. for the A and B samples, respectively, as compared with the previously recorded but less-accurate values of 0.015–0.02 mm. and 0.025 mm.

In the cold rolling of annealed copper strip the initial annealed crystals are deformed in the same proportion as the strip itself. Since a linear relationship exists between $\log t$ and the percentage rolling reduction for both A and B strips at all temperatures of annealing, then

$\log t$ is also a linear function of the mean grain thickness d after cold rolling and can be expressed as:

$$\log t = K_2 d + K_3 (14)$$

TABLE XII.—*Values of $2K_2$ for the A and B Samples for each Annealing Temperature.*

Annealing Temp., °C.	Strip A.			Strip B.		
	$2K_2$, cm. ⁻¹	Size of Nucleus, 10^{-3} cm.		$2K_2$, cm. ⁻¹	Size of Nucleus, 10^{-3} cm.	
		$d = \frac{1}{2K_2} \cdot \log_e^2$	$d = \frac{1}{2K_2} \cdot \log_e^2$		$d = \frac{1}{2K_2} \cdot \log_e^2$	$d = \frac{1}{2K_2} \cdot \log_e^2$
27°	34,200	2.0	3.45	30,400	1.9	3.3
50°	29,000	2.4	4.15	35,400	1.95	3.35
72°	26,200	2.65	4.55	28,000	2.45	4.2
100°	25,400	2.75	4.75	22,600	3.05	5.35
125°	23,800	2.9	5.0	21,400	3.15	5.6

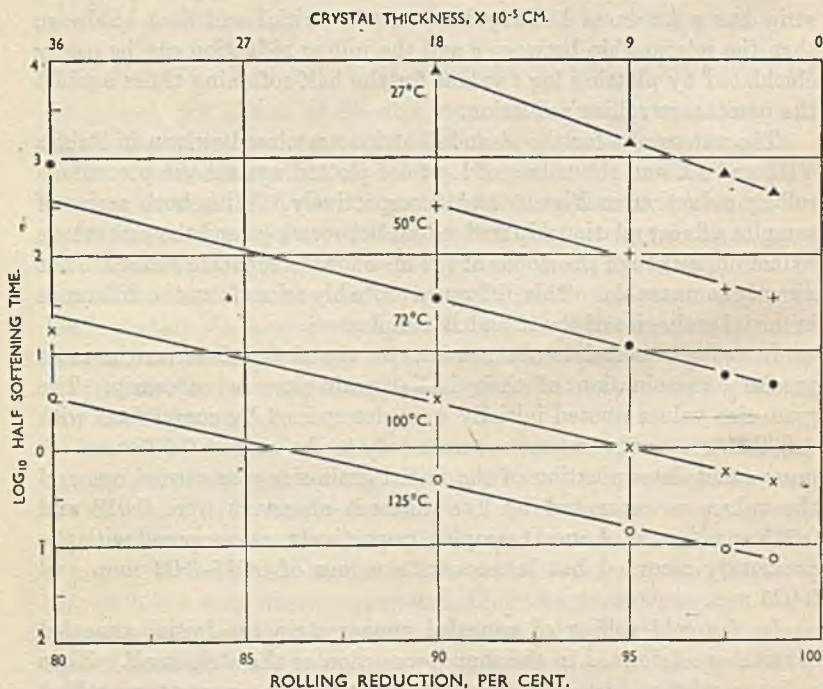


FIG. 5.—Relationship between Half-Softening Times and Percentage Rolling Reduction for the A Strips.

where K_2 and K_3 are constants for each particular annealing temperature. Combining equations (13) and (14):

$$\frac{1}{2} \log \frac{2}{c} + K_1 = K_2 d + K_3$$

and therefore:

$$c = 2 \exp. (-2K_2 d) \cdot \exp. 2(K_1 - K_3) \quad (15).$$

The corresponding values of $2K_2$ for the *A* and *B* samples which have

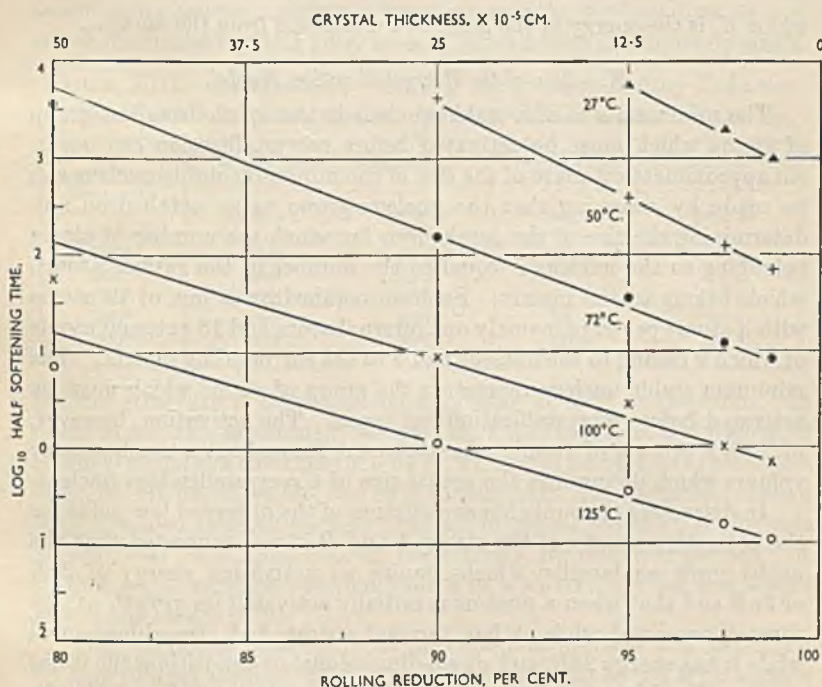


FIG. 6.—Relationship between Half-Softening Times and Percentage Rolling Reduction for the *B* Strips.

been determined for each of the annealing temperatures from the slopes of the lines in Figs. 5 and 6, and are recorded in Table XII, are in reasonable agreement with each other. $2K_2$ is, therefore, a fundamental constant for copper which is independent of the initial grain-size and dependent only on the annealing temperature.

As c is independent of temperature, the temperature dependence of K_2 arises from the variation of the activation energy ($P + Q$) with initial grain-size and degree of cold working which, in turn, influences the values of K_1 , equation (13). The factor $\exp. (-2K_2 d)$, which diminishes

exponentially with the mean grain thickness d , can be regarded as the probability that the recrystallization takes place throughout the complete thickness of the cold-rolled crystal, or $2K_2$ can be identified as an absorption factor for the thermodynamic pulses in copper. Thus when a pulse of energy E_0 develops at a nucleus, recrystallization takes place and the energy of the pulse diminishes, as it travels forward, according to the equation:

$$E_d = E_0 \exp. (-2K_2 d) \quad . \quad . \quad . \quad . \quad . \quad (16)$$

where E_d is the energy of the pulse at a distance d from the nucleus.

E.—Size of the Recrystallization Nuclei.

The minimum size of a stable nucleus is that of the smallest group of atoms which must be activated before recrystallization can occur. An approximate estimate of the size of the minimum stable nucleus can be made by assuming that the nucleus grows as an octahedron and determining the size of the octahedron for which the number of atoms belonging to the nucleus is equal to the number of the surface atoms, which belong to the matrix. Such an octahedron is one of 19 atoms with 3 atoms per edge, namely one internal atom and 18 external atoms of which 9 belong to the nucleus and 9 to the surrounding matrix. The minimum stable nucleus represents the group of atoms which must be activated before recrystallization can occur. The activation, however, initiates a process of atomic rearrangement throughout a much greater volume which determines the actual size of a recrystallization nucleus.

In discussing the probable explanation of the observed low value for the activation energy of the strips *A* and *B* it was suggested that the nuclei grow as lamellæ which require an activation energy of $2nE$ or $3nE$ and that when a nucleus is initially activated its growth will be three-dimensional while it has thermal energy $3nE$, two-dimensional while it has energy $2nE$, and single-dimensional or dendritic while it has energy nE . If the nucleus is activated by a pulse of $2nE$ or $3nE$, the pulse will still be effective in promoting growth while it has an energy nE , and the final size of recrystallization nuclei can be estimated by inserting the value $2nE$ or $3nE$ for E_0 and nE for E_d in equation (16) thus:

$$d = \frac{1}{2K_2} \log 2, \text{ or } \frac{1}{2K_2} \log 3, \text{ respectively,}$$

which places the value of d for recrystallization nuclei at 27° C. between 2 and 3.5×10^{-5} cm. and at 125° C. between 3 and 5.5×10^{-5} cm. The size of the recrystallization nuclei (see Table XII) appears to increase with increasing annealing temperature and their size at 27° C. is of the same order of magnitude as the limiting dimension, given in the previous paper¹ as less than 8×10^{-5} cm., to which the initial crystals

must be cold rolled before a completely single-texture structure is developed on self-annealing.

The close agreement between the limiting values obtaining in the two separate cases supports the argument that the merging of the twin textures in cold-rolled strip to form a single-texture structure on annealing occurs when recrystallization within one crystal has an influence extending over into the neighbouring crystals. If this is so, then the magnitude of recrystallization nuclei at any particular temperature can be determined by measuring the limiting dimension to which the initial crystals must be cold rolled before a completely single-

TABLE XIII.—*Extrapolated Values of the Critical Rolling Reduction for the A and B Strips and the Corresponding Critical Crystal Dimension.*

Final Annealing Temp., ° C.	Critical Rolling Reduction, %.		Critical Crystal Dimension d , 10^{-4} cm.	
	Strip A.	Strip B.	Strip A.	Strip B.
100°	94.5	96.5	10.0	8.8
300°	93.0	95.2	12.6	12.0
500°	90.5	93.2	17.1	16.5
700°	88.3	92.0	21.1	20.0
900°	86.1	89.7	25.0	25.8

texture structure is developed on annealing at that temperature. In order to test this assumption samples were cold rolled from the stock of *A* and *B* strips at a thickness of 0.67 in. with total reductions in thickness of 80, 85, 90, 92, 94, 96, 98, and 99% respectively, and specimens of each were annealed as small packets wrapped in copper foil in an air furnace at 300°, 500°, 700°, and 900° C. respectively, for one-hour periods. A further set of samples was maintained in a water bath for two weeks at a temperature of 100° C.

The specimens were mounted into composite blocks and examined on longitudinal sections, that is on sections normal to the strip surface and parallel to the rolling direction. They were etched electrolytically, a method which exposes the cubic planes of the crystal lattice. The presence of crystals with an orientation of the single-texture type could be detected since they appeared bright under vertical illumination and the proportion of crystals so orientated was estimated by direct visual observation. The proportions for both the *A* and *B* strips after annealing at each of the specified temperatures are plotted in Fig. 7 against the percentage rolling reduction.

The curves for the *B* series of strips comprise a family of complete summation curves, while only the top portion of the summation curves are obtained with the *A* series of strips. The complete summation curve shows first a gradual increase in percentage single-texture structure

with increasing cold-rolling reduction, then a rapid and almost linear increase over a wide range, and finally a gradual approach to the completely single-texture structure or 100% value. The initial and final parts of the curves represent reduction ranges where crystals or sections of crystals considerably below the average or considerably above the average size, respectively, are cold rolled to the critical dimension for which recrystallization at the common boundary of two cold-rolled

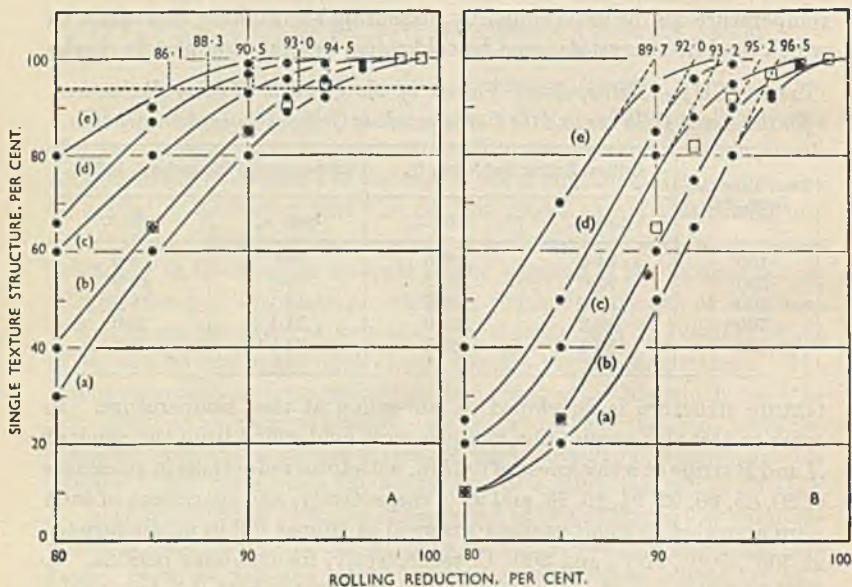


FIG. 7.—Relationship between Percentage Single-Texture Structure and Percentage Rolling Reduction for *A* and *B* Strips.

Final Annealing Temperatures.

- (a) 100°, (b) 300°, (c) 500°,
(d) 700°, (e) 900° C.

KEY.

- Air-furnace annealing.
□ Salt-bath annealing at 500° C.

crystals has an influence extending over the thickness of the crystals. The reduction at which the mean crystal diameter is cold rolled to the limiting dimension is taken as that at which the extrapolated linear portion of the summation curve meets the 100% single-texture structure line. The extrapolated values of the critical rolling reduction for the *B* strips for each annealing temperature are given in Table XIII and

the corresponding limiting crystal dimension calculated from the mean grain-size value.

It can be seen from Fig. 7 that the critical reductions of the *B* strips for each annealing temperature correspond to an actual percentage of single-texture structure of 94%. The curves for the *A* strips are not complete and the critical reductions cannot, therefore, be determined by

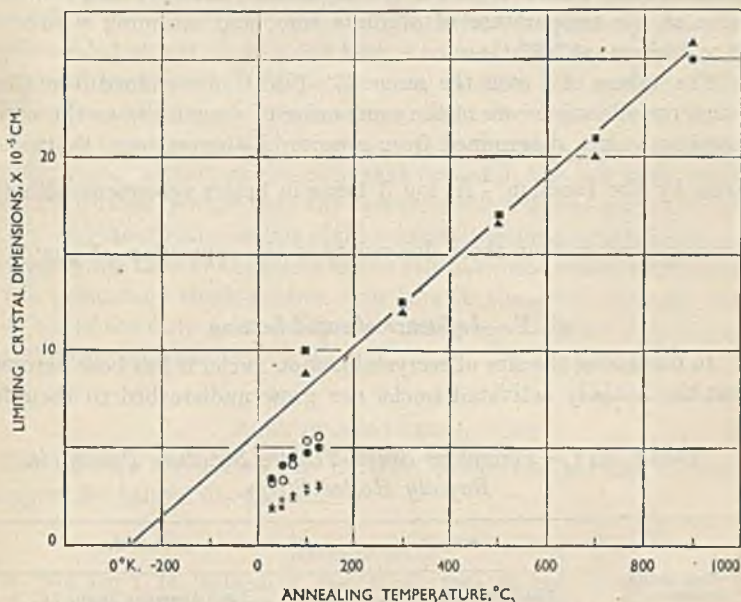


Fig. 8.—Limiting Crystal Dimensions for the *A* and *B* Strips for Various Annealing Temperatures.

Limiting Crystal Dimension, *d*.

- | | | |
|---|---------------------|--|
| ■ | For <i>A</i> strips | } From structural observations. |
| ▲ | For <i>B</i> strips | |
| + | For <i>A</i> strips | } Theoretical value $d = \left(\frac{1}{2K}\right) \log_2 2$. |
| × | For <i>B</i> strips | |
| ○ | For <i>A</i> strips | } Theoretical value $d = \left(\frac{1}{2K}\right) \log_3 3$. |
| ● | For <i>B</i> strips | |

extrapolation but have been taken as the reductions which, on subsequent annealing, yield a structure consisting of 94% single-texture structure. The critical reductions and corresponding limiting dimensions for the *A* strips are also given in Table XIII.

The limiting crystal dimensions for the *A* and *B* strips for various annealing temperatures are plotted in Fig. 8, together with the values

determined over the range 27°–125° C. from the thermodynamical theory of recrystallization, which are given in Table XII. Considering the difficulty in making precise determinations of the grain-sizes of the initial strips and also of the critical reductions, the agreement between the corresponding limiting crystal dimensions for the *A* and *B* strips is good. Within the limits of experimental error the limiting dimension *d* varies linearly with the annealing temperature, extrapolating to a zero value at the temperature of absolute zero, and attaining a value of 25×10^{-5} cm. at 900° C.

The values of *d* over the range 27°–125° C. determined from theoretical considerations are of the same order of magnitude as the corresponding values determined from structural observations, the values given by the function $\frac{1}{2} K_2 \log 3$ being in better agreement with the structural values than those derived from the expression $\frac{1}{2} K_2 \log 2$.

F.—Influence of rapid heating.

In discussing the size of recrystallization nuclei it has been assumed that the initially activated nuclei can grow undisturbed to their full

TABLE XIV.—Percentage Single-Texture Structure Present in Rapidly Heated Strips.

Final Rolling Reduction, %.	Strip A.		Strip B.	
	Final Annealing Temp., ° C.		Final Annealing Temp., ° C.	
	500° (5 min.).	500° (1 hr.).	500° (5 min.).	500° (1 hr.).
80	40	40	10	10
85	65	65	25	25
90	85	85	65	65
92	90	90	80	82
94	96	96	88	92
96	98	99	96	97
98	100	100	98	99
99	100	100	100	100

potential size as defined at the beginning of section E. When the rate of heating is slow, or at low temperatures, this may be true, since recrystallization will take place first in the crystals of thinnest section and nucleation will be well distributed in time and space. If, however, the rate of heating is rapid, then there may be a rapid onset of nucleation and the effective radius will, therefore, be diminished. In order to study

the effect of rapid heating, materials similar to those used in the previous experiment were annealed both for five-minute and for one-hour periods in a salt bath at 500° C., uncovered samples being immersed directly into the molten salt. The percentage single-texture structure present in these annealed strips was estimated as before. The results are recorded in Table XIV and the values for one-hour periods are plotted in Fig. 7 for direct comparison with observations upon samples annealed in air. The percentage single-texture structure present in strip annealed at 500° C. in a salt bath is considerably less than that in corresponding strip annealed in air and, in fact, the curve for the samples annealed in a salt bath is more or less coincident with the curve for samples annealed at 300° C. in air. It would seem, therefore, that with the rapid heating of the strip that occurs in the salt bath, fresh nuclei are formed quicker than the first-formed nuclei can grow to their full size and the effective radius of the recrystallization nuclei is reduced. Prolonging the time of annealing in the salt bath was without influence on the percentage single-texture structure in the strip although the grain-size of the strip annealed for one hour was somewhat larger than that of strip annealed for five minutes.

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PROPERTIES OF THE ALUMINIUM-SILICON 1035 ALLOYS AT TEMPERATURES IN THE REGION OF THE SOLIDUS.*

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SYNOPSIS.

The tensile properties of the aluminium-silicon alloys were determined at temperatures in the region of the solidus, to obtain information that would assist in explaining the mechanism of hot-shortness.

The tensile strengths of aluminium-silicon alloys containing up to 12% of silicon show a gradual decrease as the temperature increases above 400° C. to finite values at the solidus temperatures. The steepness of the curves increases with increasing silicon content, but up to the solidus temperature the elongation and reduction of area of all the alloys remain high. At the solidus temperature, elongation and reduction of area decrease to zero and the strength rapidly decreases to a very low value. The alloys retain this small strength, or coherence, up to a temperature intermediate between that of the solidus and liquidus. The temperature range above the solidus over which the alloys are coherent is dependent on their chemical composition, and in conditions approaching equilibrium increases to a maximum at approximately 1.8% of silicon and then decreases to a small value at higher percentages of silicon.

By micrographic investigation, it has been possible to link the tensile properties at temperatures in the region of the solidus with the aluminium-silicon equilibrium diagram.

It appears that the extent of the temperature range above the solidus over which the alloys possess a finite strength yet have no appreciable ductility may be one of the principal factors determining their hot-shortness characteristics.

I.—INTRODUCTION.

ONE of the most important properties concerned in the casting and welding of aluminium alloys, and one which is particularly troublesome in a variety of ways, is their hot-shortness. In the case of castings it frequently happens that aluminium alloys fracture while cooling in, or being extracted from, the mould. Failure occurs while the castings are still at a high temperature and the phenomenon is often known as "hot-tearing." Similarly, welds frequently crack immediately after being formed, giving the same type of brittle, intercrystalline fracture as the castings. This property is especially important in processes involving a change from the liquid to the solid state, for quite often, as in the instances mentioned above, conditions are such that movement is restrained, and in consequence contractional stresses are set up which bring about failure of the material by cracking.

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Although the mechanical and physical properties of alloys at temperatures near the solidus must ultimately be related to failure by hot-shortness, and although a better knowledge of these properties would be invaluable in solving many of the numerous problems of casting and welding technology, little success has so far been achieved in efforts to bring this about. The purpose of the present investigation was to study the mechanical properties at high temperatures of one of the simpler alloy systems—aluminium-silicon—to obtain more information on the causes of failure by hot-shortness. To obviate undue complexity and lengthiness the investigation has been limited to a study of mechanical and other directly related properties. The correlation of the work with practical casting and welding experiments is the subject of a separate investigation.

II.—PREVIOUS WORK.

The tensile strengths of the aluminium alloys are, in general, found to decrease rapidly to very low values between 300° and 400° C. Most investigators have been interested in the mechanical properties of the alloys in relation to their application to industry and there is a wealth of literature on the subject. Because of the very low strengths encountered, however, the field above the recrystallization temperature has not been so extensively explored; in particular, very few determinations of mechanical properties have been made in the immediate neighbourhood of the melting point.

The first detailed investigation in the region of the solidus was made by Archbutt, Grogan, and Jenkin,¹ who were endeavouring to find a clue to the hot-shortness of a number of aluminium alloys by determining their impact values at temperatures near the melting point. It was found that most of the alloys examined lost their resistance to fracture practically completely at the temperature of incipient melting. As a consequence, it was assumed that differences in hot-shortness determined by casting experiments could only be correlated with the differences in resistance to fracture at temperatures lower than that of incipient melting. It was thought that a rapid increase in resistance with decreasing temperature indicated a low degree of hot-shortness and the reverse indicated a high degree of hot-shortness. However, below the point of incipient melting the alloys assumed an increasing ductility, which property was essentially different from that displayed by alloys in the hot-short range, and, in fact, no wholly satisfactory correlation of hot-shortness with mechanical properties was achieved in the investigation.

The most extensive study of the subject was made by Verö ² who determined the strength of a number of aluminium alloys, including a series of aluminium-silicon alloys, at temperatures both below and above the solidus. Verö used both tensile tests and bend tests, but, being unable to obtain reproducible results with the tensile test, discarded it in favour of the bend test. His results on the aluminium-silicon alloys are shown in Fig. 1, from which it will be seen that the bending strength of all the alloys decreased rapidly with increasing temperature to a point near that of incipient fusion, and then in some alloys decreased more gradually to zero at a temperature below the liquidus for the alloy in question.

Verö found that he could distinguish no essential difference between the bending strength at high temperatures of an alloy containing 1.6% silicon (which in some casting experiments he had shown to have a high degree of hot-shortness) and that of one containing 1.88% silicon, having a low degree of hot-shortness. In view of this apparent contradiction, the attempt to correlate mechanical properties with hot-shortness was abandoned.

Tammann and Rocha ³ showed that the tensile strength of zinc, cadmium, iron, and silver containing traces of tin, bismuth, FeS, and copper, respectively, showed a sharp decrease at the temperature of the solidus and then a much smaller decrease as the temperature was increased. Their investigation, however, was carried out primarily to establish a method of detecting minute quantities of eutectic not readily detectable by other means and was, therefore, not directly related to the hot-shortness properties of the alloys.

Some experiments on cast-steel bars were carried out by Hall ⁴ in which the tensile strengths and elongations were determined during and immediately after solidification of the castings to throw some light on their propensity to hot-tearing. The results were made difficult to

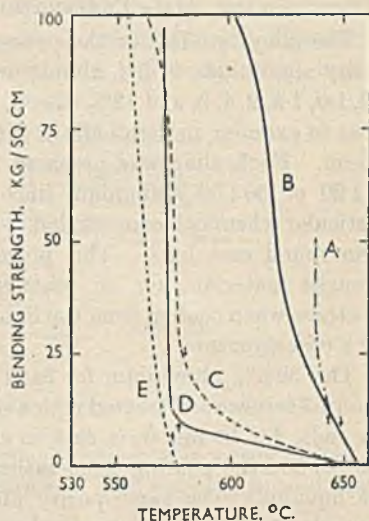


FIG. 1.—Bending Strength of Aluminium Alloys in the Melting Range. (Verö.)

- A = High-purity aluminium.
- B = 99.5% aluminium.
- C = 1.60% silicon.
- D = 1.88% silicon.
- E = 12% silicon.

interpret by the steep thermal gradient present in the castings at the time of testing. Hall found that the tensile strength increased steadily as the temperature decreased and concluded that it was necessary to go some distance below the solidus temperature before an appreciable elongation was acquired and hot-tearing was no longer possible.

III.—PREPARATION OF MATERIALS.

The alloys studied in the present investigation consisted of super-purity aluminium, 99.5% aluminium, and eight alloys containing 0.25, 0.6, 1.0, 1.5, 2, 4, 8, and 12% silicon, respectively, the series being chosen so as to examine in detail alloys within Verö's critical range of 0.5–2% silicon. Each alloy was prepared from 99.5% aluminium and either 80:20 or 50:50 aluminium-silicon temper alloy, depending on its particular chemical composition. All the test-pieces were machined from round cast bars. This procedure was preferred to the use of wrought materials, for, in practice, the alloys usually exhibit hot-shortness when cooling from the liquid state and therefore when possessing a cast structure.

The 99.5% aluminium for each alloy was melted and alloyed in a gas-fired furnace and treated with a sodium fluoride/sodium chloride flux. Six rods, $\frac{3}{4} \times 10$ in., were cast in chill cast-iron moulds maintained at 100° C. and the pouring temperature was standardized at 70° C. above the liquidus. The super-purity aluminium was melted and cast in precisely the same way, except that the salamander pot was lined with silicon-free Alundum cement to reduce contamination.

It was noticed that, in the case of the alloys containing 0.6 and 1% of silicon, deep cracks appeared at the junction of the feeder head and the parallel portion of the rod, where the parts under contractional stress were last to solidify.

TABLE I.—*Analysis of the Aluminium-Silicon Alloys.*

Nominal Composition.	Chemical Analysis.	
	Silicon, %.	Iron, %.
Super-purity aluminium	<0.01	<0.01
99.5% aluminium	0.12	0.27
Al + 0.25% Si	0.21	0.29
Al + 0.6% Si	0.61	0.21
Al + 1.0% Si	1.08	0.27
Al + 1.5% Si	1.62	0.25
Al + 2% Si	2.08	0.28
Al + 4% Si	3.87	0.27
Al + 8% Si	7.65	0.27
Al + 12% Si	11.83	0.23

In addition, a small ingot was cast for purposes of analysis. Silicon and iron were determined chemically, with the results detailed in Table I.

IV.—APPARATUS.

A Hounsfield tensometer, fitted with a motor drive, was used for the tensile tests. A resistance tube furnace, 8 in. long, was constructed so as to slide on the horizontal supporting columns of the tensometer and to enable the test-pieces to be kept at a high temperature during

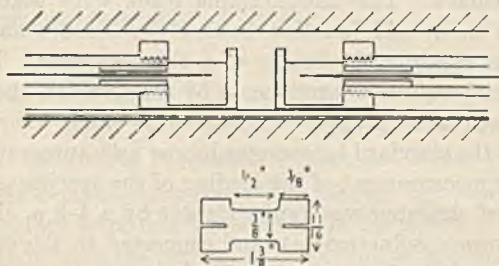


FIG. 3.—Apparatus for Testing at Temperatures below the Solidus.

testing. A general view of the furnace assembly is given in Fig. 2 (Plate I), and details of the two forms of test-piece grips and test-pieces are shown in Figs. 3 and 4.

Apparatus for Testing at Temperatures below the Solidus.

At temperatures below the solidus, test-pieces were used of the dimensions shown in Fig. 3 so as to enable measurements of elongation

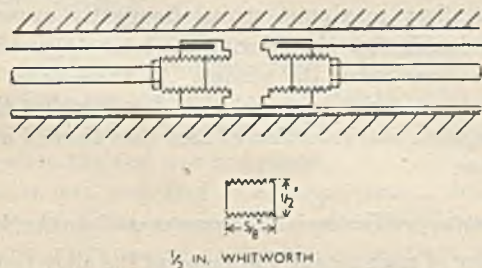


FIG. 4.—Apparatus for Testing at Temperatures above the Solidus.

and reduction of area to be made, and at the same time to ensure a minimum variation of temperature over the length of the sample and small opportunity for any appreciable amount of sagging.

To minimize the changes in structure which would probably occur if the test-pieces were kept at the testing temperature for long periods of time, the apparatus was constructed so that they could quickly be dropped into position in the preheated grips and tested in as short a time as was practicable. When a test-piece was in the correct position for preheating, the two Chromel-Alumel thermocouples were advanced axially along the tubular links into the holes drilled in the ends of the test-piece to receive them. Temperature measurements were thus taken at each end of the parallel portion by actual contact of the couple with the specimen. The thermocouple leads were taken along the tubular links to a cold junction behind the furnace and the e.m.f. generated was measured by means of a potentiometer. To avoid upsetting the temperature measurements by air currents, the ends of the tubular furnace were plugged with asbestos wool. Throughout the investigation the standard tensometer beams and autographic recorder were used for measurement of the loading of the specimens.

The rate of straining was kept constant by a $\frac{1}{3}$ -h.p. electric motor fitted with a worm reduction gear and connected to the tensometer by means of $\frac{1}{2}$ -in.-pitch chain wheels. The greater part of the work was carried out at a rate of straining of 0.25 in./min., but, in view of the important effect of speed on mechanical properties, two other speeds of testing, 0.05 and 1.0 in./min., were used on 99.5% aluminium and on the alloys containing 1.5 and 4% silicon at temperatures both below and above the solidus.

The normal type of test-piece gave reproducible results throughout the work, even though it represented the centre portion and only one-quarter of the full section of the cast rod. Little porosity was observed in the rods, although very small defects or segregations in the castings (insufficient to affect maximum stress) frequently caused premature fracture of the ductile test-piece during the later stages of its deformation at temperatures below the solidus. In these cases it was quite apparent that the material possessed high ductility, but the numerical values for elongation and reduction of area then showed a considerable degree of scatter.

Apparatus for Testing at Temperatures above the Solidus.

The difficulty of maintaining the shape of the alloy test-piece in the partially liquid condition was overcome by using a short sample totally enclosed by the grips. The test-piece was essentially a $\frac{5}{8}$ -in. length of cast bar, on which was machined a $\frac{1}{2}$ -in. Whitworth screw thread. This type of test-piece was chosen because it is easy to produce and fit and its short length in relation to diameter results in very small tempera-

ture differences within the test-piece. A diagram of the screw sample and the grips is given in Fig. 4.

At the outset, it was anticipated that one of the difficulties that would arise would be contamination of the partially liquid alloys by iron from the grips. In practice it was found that results obtained with chromium-plated, 18 : 8 stainless steel and mild-steel grips were identical, the reason apparently being that the oxide skin on the screw specimen thickened on heating and held the partially liquid metal in an almost impervious sac. Careful chemical analysis of the outside layer of some of the samples heated in the mild-steel grips confirmed that no such contamination had occurred.

V.—PRACTICAL WORK.

Testing at Temperatures below the Solidus.

In tests at temperatures below the solidus, the procedure was to preheat the empty grips to a temperature in the region of that required for testing, slide back the furnace to a position to the left (so exposing the grips on a tongue projecting from the furnace tube), drop the cold aluminium alloy test-piece into place, and slide the furnace back to the original position. The whole operation took only a few seconds, so the decrease in temperature was quite small. By this means it was found possible to attain a temperature 10° C. less than that of testing within 3 min. and a steady temperature within 15 min. All the tests with this type of test-piece were carried out after 15 min. preheating. The experimental values of maximum stress, elongation, and reduction of area are given in Tables II–XI and a diagram of the experimental curves is shown in Fig. 5.

Testing at Temperatures above the Solidus.

The screw-type test-pieces were machined to fit loosely in the grips and a slot was cut at each end so that they could be removed by a screwdriver when the test was completed.

The furnace was preheated to a temperature a little greater than that necessary for the test and it was then slid over the grips, test-piece, and thermocouple assembly. The time of heating was standardized at 30 min., the assembly requiring 15 min. to attain the approximate testing temperature. At the end of 30 min. a steady state had been reached and differences in temperature between the two halves of the shackle were less than 1° C. When carrying out the mechanical tests above the solidus temperature, the maximum of the

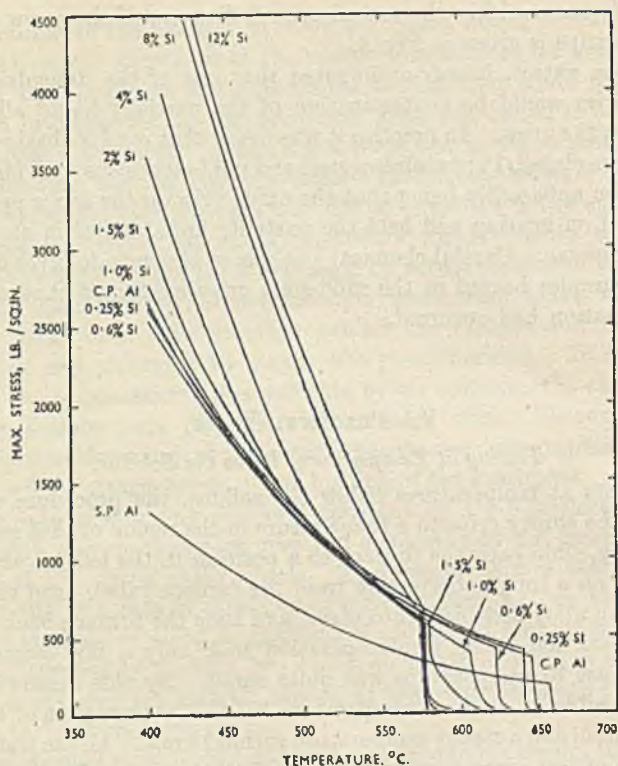


FIG. 5.—Strength of the Aluminium-Silicon Alloys at Temperatures in the Region of the Solidus.

TABLE II.—*Mechanical Properties of Super-Purity Aluminium.*

Temperature, ° C.	Maximum stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
430	1070	100	133
498	660	100	161
571	340	100	160
610	270	100	142
652	185	100	138
656	155	0	0
657	81	0	0
657	9	0	0
658	0	0	0
659	<5	0	0
662	0	0	0

TABLE III.—*Mechanical Properties of 99.5% Aluminium.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
402	2620	64	43
481	1370	60	43
538	880	84	44
578	620	64	40
597	510	100	100
602	560	68	36
621	465	60	26
629	420	82	...
645	340	0	...
647	165	0	0
650	30	0	0

TABLE IV.—*Mechanical Properties of 0.25% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
392	2750	50	35
429	2180	100	74
455	1720	20	13
490	1300	96	74
500	1250	75	49
554	750	100	96
567	710	92	74
576	720	100	97
597	550	100	110
599	560	50	44
610	550	100	120
625	496	90	60
631	440	100	110
636	430	20	...
640	70	0	0
641	350	0	0
641	84	0	0

TABLE V.—*Mechanical Properties of 0.6% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
423	2210	100	140
437	1820	58	47
478	1410	40	26
504	1300	96	100
541	900	68	50
571	630	70	62
596	540	76	58
608	550	98	150
613	460	79	100
622	315	0	0
624	108	0	0
627	63	0	0

TABLE VII.—*Mechanical Properties of 1% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
422	2310	86	79
482	1400	40	33
520	1010	62	55
549	775	58	53
569	650	100	110
587	530	64	69
597	425	12	22
601	425	0	0
605	395	0	0
607	295	0	0
610	59	0	0
612	63	0	0
613	113	0	0
616	68	0	0
617	32	0	0

TABLE VI.—*Mechanical Properties of 1.5% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
392	3300	34	26
419	2880	56	...
436	2160	20	18
479	1530	42	36
505	1120	40	23
528	1040	62	...
545	855	98	120
561	675	...	140
567	700	93	94
578	410	0	0

mercury index was read directly, as the ductility of the materials was too small to permit the plotting of a stress-strain curve.

The tensile strength of the alloys above the solidus temperature are shown in Fig. 6. In each case the crosses represent temperatures at which the strength is zero. For the sake of clarity, each curve in Fig. 6 has been displaced vertically from its neighbours by an amount of 50 lb./sq.in. stress so that the temperature at which the strength of the material decreases to zero value can be clearly indicated.

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TABLE VIII.—*Mechanical Properties of 2% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
419	3920	72	66
419	2620	30	55
442	2350	35	40
461	1910	45	...
478	1480	30	23
502	1200	92	96
515	1080	78	68
534	900	95	120
557	740	78	47
564	620	80	98
569	625	55	52
572	630	72	71
575	600	15	...
578	204	0	0
578	200	0	0
578	130	0	0

TABLE X.—*Mechanical Properties of 8% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
426	4410	55	...
449	3510	58	46
467	3085	62	53
505	2000	71	...
520	1350	82	83
542	1190	85	83
561	815	84	110
565	860	85	110
567	730	60	63
574	560	11	6
575	99	0	0
576	135	0	0

TABLE IX.—*Mechanical Properties of 4% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
435	3160	78	54
448	2620	30	29
455	2500	86	47
475	2080
493	1460	64	64
500	1450	95	70
521	1190	90	110
525	1020	100	110
536	900	84	56
538	890	80	57
554	760	96	...
555	820	92	110

TABLE XI.—*Mechanical Properties of 12% Silicon Alloy.*

Temperature, ° C.	Maximum Stress, lb./sq.in.	Reduction of Area, %.	Elongation, %.
399	6110	41	23
430	4480	36	30
463	3220	38	31
490	2520	53	43
521	1800	53	57
535	1500	64	76
548	1190	70	76
561	945	68	83
571	755	76	130
574	700	36	35
576	410	0	0
577	230	0	0
578	5	0	0
578	<5	0	0
584	0	0	0

Microscopical Observations.

To show the type of structure persisting at the temperature of testing, samples of the original castings were given precisely the same thermal treatments as those outlined above, but, instead of being pulled on the tensometer, they were quickly withdrawn from the furnace and quenched in cold water. These samples were sectioned, polished, and examined under the microscope together with polished sections from the untreated original casting.

VI.—EXPERIMENTAL RESULTS.

Super-Purity Aluminium.

Figs. 5 and 6 show the relationship between maximum stress and temperature for super-purity aluminium, while in Table II the experi-

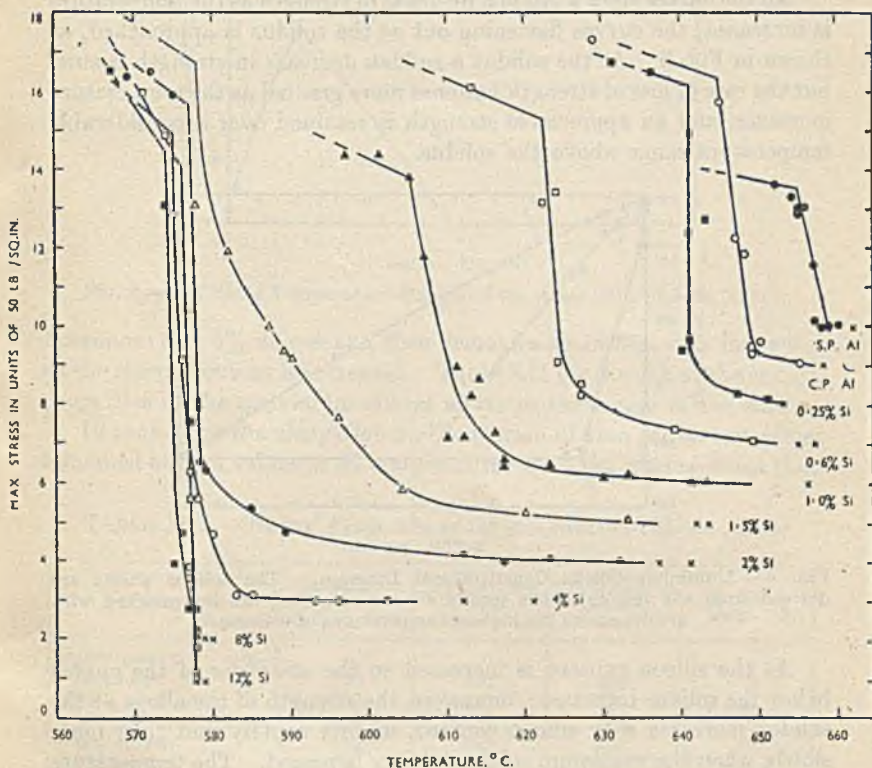


FIG. 6.—Strength of the Aluminium-Silicon Alloys at Temperatures above the Solidus.

mental values are given for strength, elongation, and reduction of area.

It will be seen that up to 655° C. the strength decreases slowly with increasing temperature, forming a very shallow curve. At 655° C., however, there is a sudden decrease in strength from 180 lb./sq.in. to zero at 659° C. This behaviour was confirmed in a remarkable way by a study of the deformed test-pieces. All those giving results lying on the horizontal portion of the curve showed a very high elongation and a 100% reduction of area. From 655°–659° C. the samples had

bright, intercrystalline fractures and showed no elongation. Fig. 7 (Plate I), which shows four of the fractured test-pieces, illustrates the remarkable transformation very clearly.

99.5% Aluminium and Alloys Containing 0.25–12% of Silicon.

All the curves show a gradual decrease in strength as the temperature is increased, the curves flattening out as the solidus is approached, as shown in Fig. 5. At the solidus a sudden decrease in strength occurs, but the rate of loss of strength becomes more gradual as the temperature increases, and an appreciable strength is retained over a considerable temperature range above the solidus.

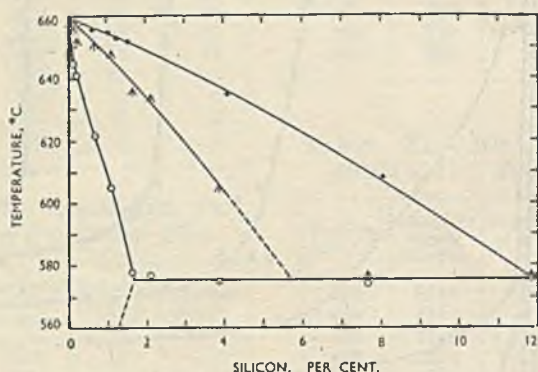


FIG. 8.—Aluminium-Silicon Constitutional Diagram. The solidus points are derived from the results of the tensile experiments, and the line marked with arrows shows the highest temperatures of coherence.

As the silicon content is increased so the steepness of the curves below the solidus increases; moreover, the strength of the alloys at the solidus increases with silicon content, at first rapidly and then more slowly when the maximum solid solubility is passed. The temperature of the solidus, indicated by the sudden increase in strength, decreases with increasing silicon content until the eutectic temperature is reached at 2% silicon; thereafter the temperature of the solidus remains practically constant.

Fig. 8 incorporates a number of the more important experimental results of the investigation. In this diagram the solidus points have been taken from Fig. 6 and represent the temperatures at which the sudden decrease in strength occurs, and the small arrows represent the highest temperatures at which the alloys retain coherence. The liquidus points were obtained experimentally by taking cooling curves on some of the alloys used in the investigation. Fig. 9 is derived from

Fig. 8 by plotting the temperature difference between the solidus and the line marked with arrows—the hot-short temperature range—against chemical composition. This curve, therefore, shows the important feature that the hot-short temperature range increases to a

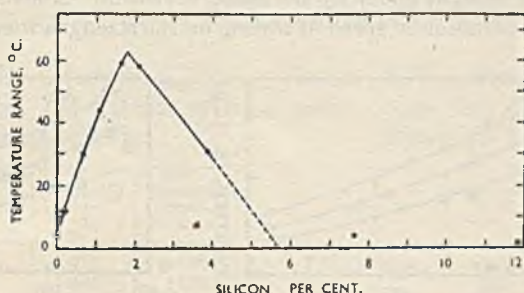


FIG. 9.—Hot-Short Temperature Ranges of the Aluminium-Silicon Alloys.

maximum at 1.8% silicon and then decreases again to a very low value as the silicon content is increased. Table XII gives some of the critical properties of the aluminium-silicon alloys in the region of the solidus.

In each alloy the elongation and reduction of area of the test-pieces remained at high values at all temperatures up to the solidus, when they

TABLE XII.—Critical Properties of the Aluminium-Silicon Alloys.

Alloy.	Strength at Critical Point, lb./sq.in.	Temperature of Decrease in Strength, °C.	Temperature of Loss of all Strength, °C.	Hot-Short Range, °C.
Super-purity aluminium .	180	655	659	4
99.5% aluminium .	370	645	657	12
0.25% silicon .	410	641	653	12
0.6% silicon .	420	622	652	30
1.0% silicon .	400	605	649	44
1.5% silicon .	590	578	637	59
2% silicon .	590	577	635	58
4% silicon .	600	575	606	31
8% silicon .	630	574	578	4
12% silicon .	660	576	578	2

decreased to zero. The maximum elongation and reduction of area were, in general, shown by samples tested at temperatures a few degrees below the solidus; they were greatest with the alloys low in silicon, and least with the eutectic alloy. Fig. 10 (Plate I) illustrates this sudden decrease in ductility in the case of a typical alloy containing 0.6% silicon.

Variation in the Rate of Straining.

The effect of varying the rate of straining from 0.05 to 1.0 in./min. in three different alloys is shown in Figs. 11 (a), (b), and (c). It will be observed that below the solidus an alteration in maximum stress of less than 30% is brought about by the speed variation. Above the solidus the effect of variation in speed of testing on the strength was less clearly

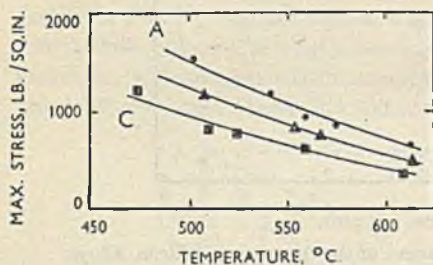


FIG. 11 (a).

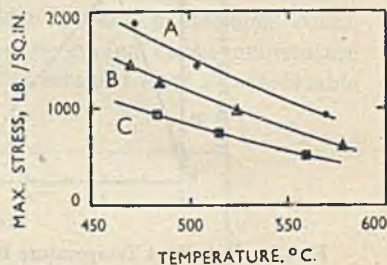


FIG. 11 (b).

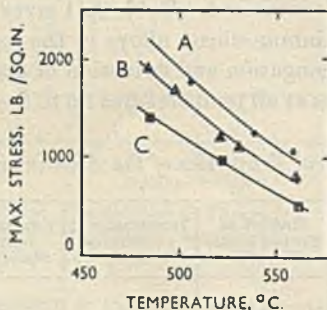


FIG. 11 (c).

FIG. 11.—Effect of Variation of the Rate of Straining on the Mechanical Properties of (a) 99.5% Aluminium, (b) 1.5% Silicon Alloy, and (c) 4% Silicon Alloy.

- 1 in./min.
- ▲— 0.25 in./min.
- 0.05 in./min.

defined, because of the lower order of stress. The effect of speed of testing within the above-mentioned limits, however, did not appear to cause any substantial alteration to the point at which all strength was lost.

The Effect of Prolonged Heating.

Several samples were heated at temperatures both above and below the solidus for periods up to 2 hr. before final testing in an effort to

determine the effect on the mechanical properties of a closer approach to equilibrium conditions and a possible change in geometric structure of the test-piece. It was found that below the solidus negligible differences in strength occurred after prolonged heating, but that above the solidus strengths were somewhat reduced. The values of strength determined on the 1.5, 2, and 4% silicon alloys are given in Tables XIII and XIV.

TABLE XIII.—*Mechanical Properties below the Solidus after Heating for 1 Hr.*

Alloy.	Temperature, ° C.	Maximum Stress, lb./sq.in.
1.5% silicon.	471	1490
	501	1170
	509	1080
	568	630

TABLE XIV.—*Mechanical Properties above the Solidus after Heating for 2 Hr.*

Alloy.	Temperature, ° C.	Maximum Stress, lb./sq.in.
1.5% silicon.	602	28
	607	24
	624	4
	628	4
2% silicon.	599	14
	619	4
4% silicon.	583	4
	588	4

Microscopical Examination.

Sections of the original castings showed that all the alloys possessed a fine-grained, equi-axed structure, the grain-size decreasing with increasing silicon content. The super-purity aluminium and 99.5% aluminium castings showed the type of columnar structure depicted in Fig. 12 (Plate II).

The structure of some of the samples quenched from temperatures just below the solidus differed somewhat from the cast structure. The eutectic silicon originally present in the 1.0 and 1.5% silicon alloys was completely absorbed into solid solution, and the quantity present in the 2% silicon alloy showed a marked reduction. The 4, 8, and 12% silicon alloys, however, appeared to be virtually unchanged in structure.

Above the solidus temperature it was found that each crystal contained considerable quantities of an originally liquid silicon-rich phase in globular form within its boundaries, and each crystal was separated from its neighbour by a thin film of similar material. As the quenching temperature was increased, so the liquid films at the grain boundaries widened and the globules within the crystal became more pronounced. It was apparent that only a fraction of the liquid in the samples at temperatures above the solidus was effective in reducing

strength, for the greater part of it was in globular or land-locked form, well away from the grain boundaries where separation of the crystals was likely to occur. Fig. 13 (Plate II) illustrates this type of structure. It will be observed that the greater part of the chilled silicon-rich liquid is contained within the grains and consequently is ineffective in reducing strength.

Sections of the fractured test-pieces were etched to reveal the grain structure. It was found that during the period at the testing temperature, both above and below the solidus, no marked grain growth had taken place. Cracks which had formed near the point of fracture of those test-pieces tested above the solidus were in every case found to be of an intergranular type.

VII.—DISCUSSION.

The field covered by this investigation was so wide that it was found impracticable to give any but the scantiest attention to the many problems arising during the course of the work. In particular, a great deal of further study is necessary to determine the precise effect of grain-size and shape, and of the mode of distribution of liquid within the grain, on the mechanical properties of the alloys at temperatures above the solidus. The time factor is another problem of which only the fringe has been touched, for it is of importance that we should know in what way the speed of the thermal treatment affects the structure and tensile properties.

Measurement of Stress and Temperature.

It was soon apparent that the absolute values of strength at temperatures in the region of the solidus were of less importance than the characteristic shapes of the strength curves, for differences in test-piece shape and structure and particular conditions of testing each had a slight effect on individual measurements.

At temperatures below the solidus the accuracy of measurement was comparatively high and reproducibility good, but above the solidus the accuracy decreased markedly. Loads down to 1 lb. (stress 8 lb./sq. in.) were estimated numerically, while those less than this were recorded as 0.5 lb. (stress 4 lb./sq. in.), less than 0.5 lb. (stress less than 4 lb./sq. in.) where there was a perceptible but non-measurable load, and zero where no deflection of the beam was observed. The accuracy in the lower reaches was, of course, very low, but these measurements enabled a fairly close estimation of the temperature range above the solidus over which the alloys were coherent. Calculation was made of the effect of the inertia of the connecting links and tensometer beam in giving rise to

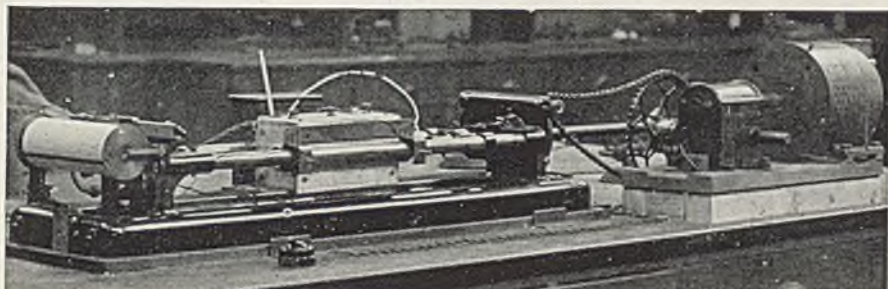


FIG. 2.—General View of Apparatus.



FIG. 7.—Super-Purity Aluminium Samples Tested at 498° , 652° , 656° , and 659° C., respectively. Natural size.

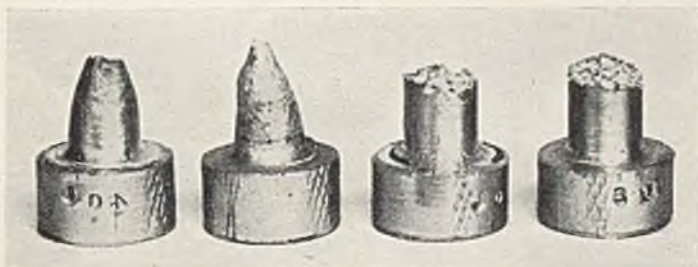


FIG. 10.—0.6% Silicon Alloy Samples Tested at 541° , 613° , 622° , and 627° C., respectively. Natural size.



FIG. 12.—Microstructures of Chill-Cast Rods of 99.5% Aluminium (Columnar) and of 1% and 8% Silicon Alloys (Equi-Axed).



FIG. 13.—4%-Silicon Alloy Quenched from 580° C., Showing the Distribution of the Chilled Silicon-Rich Liquid. $\times 75$.

a stress in the sample greater than that measured by the mercury index; it was found to be wholly inappreciable, even at the maximum speed of testing of 1 in./min. The frictional effect of the grips on the guides at the bottom of the horizontal furnace was studied experimentally and it was found that no perceptible deflection of the beam occurred when the furnace was moved in relation to the grips.

The temperature gradients within the normal type of test-piece were investigated by drilling a hole axially through the whole length of the test-piece, advancing a search couple to various positions along the gauge length, and comparing with a second couple placed in the normal position. No differences in temperature greater than 1° C. were observed throughout the sample after heating for 15 min. The Chromel-Alumel thermocouples were checked regularly against pure metals to guard against errors due to contamination from the aluminium samples. It is estimated that amongst themselves the temperature measurements of the samples were accurate to $\pm 1^{\circ}$ C., with an absolute accuracy of $\pm 2^{\circ}$ C.

In changing over from the normal type of test-piece to the screw type it was not thought that above the temperature of the solidus the strength values would be seriously altered by the notch effect of the screw threads. It seemed probable that, when dealing with an alloy in the region above the solidus, discontinuities in the form of liquid films and pools within the material would always be present to initiate crack formation, and a notch of the order of size of a screw thread would be unlikely to cause a great decrease in strength. In practice, a measure of correlation was obtained by extending the use of the normal-type test-piece to temperatures slightly above the solidus, so as to overlap the field of the enclosed type.

Measurements of stress on the vertical portions of the curves were difficult to make, but, as can be seen from the values given in Tables III–XI for normal test-pieces and in Fig. 6 for screw test-pieces, on the whole the agreement was quite good.

General Characteristics of the Strength Curves and Theoretical Considerations Relating to Them.

The chief characteristic of the curves is their gradual approach to a certain strength at the solidus, followed by a steep, and practically vertical, descent to a very low value of strength, which slowly diminishes to zero at a point intermediate between the solidus and liquidus temperatures.

This is the type of curve that would be expected from theoretical considerations. If, for the moment, the system be considered to be in

equilibrium and it be supposed that an alloy of composition "C," Fig. 14, be under examination, then, as the temperature increases near the solidus at "H," the strength will decrease gradually until it reaches a certain positive value at a temperature immediately below "H." This positive strength at the melting point was well shown by Tammann and Rocha³ in their study of zinc and silver containing small quantities of tin and copper, respectively. Immediately above the solidus the solid will be surrounded by a finite, small quantity of liquid eutectic, causing the strength of the mass to decrease sharply to a low value. Provided that this quantity of liquid does not exceed a certain amount, a definite strength will be shown because of the interlocking effect of the remaining solid crystals and the viscosity of the liquid eutectic in the narrow channels between them. As the amount of liquid increases with increase in temperature, the strength will decrease and will reach zero value at a definite temperature between the solidus and liquidus, as shown by "C," Fig. 15, which depicts the strength-temperature relationship for an alloy of composition "C."

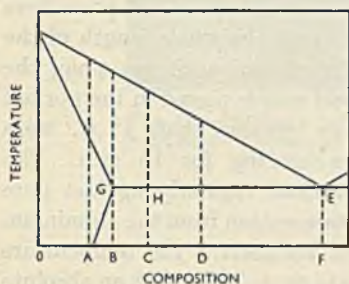


FIG. 14.—Theoretical Diagram of Typical Eutectiferous System.

As the composition of the alloy varies, so the shape of the curves in Fig. 15 will alter both above and below the solidus. At compositions to the left of "B," the amount of liquid in equilibrium with the solid at the solidus temperature would be zero, but would increase as the temperature increased, so that, although there would be a discontinuity in the curve, there would be no vertical descent and the strength would

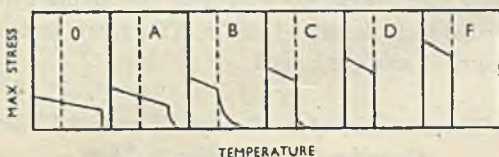


FIG. 15.—Theoretical Strength-Temperature Relationships for the Alloys of Compositions 0-F, derived from Fig. 14.

not decrease to such a marked degree as with alloys having compositions to the right of "B." If it be supposed that when less than a certain amount of liquid is present the partially liquid mass has a positive strength, then the temperature range over which the alloy possesses appreciable strength will be greatest at "B" and will decrease to zero

at composition "O" on the one side and at a point intermediate between "B" and "F" on the other, say at composition "D." The theoretical types of curves to be expected with six different alloys are shown in Fig. 15, which is intended to represent only the types of change taking place and is not drawn to scale.

There is quite good agreement between these theoretical curves and those obtained by experiment in the aluminium-silicon series. The fact that the experimental observations conform quite closely to theoretical expectations appears to indicate that, at these relatively high temperatures, an approach to equilibrium as regards chemical composition has been made even though the time at the testing temperature has only been of the order of 10-15 min.

Hot-Short Temperature Range.

During the tensile tests, it was observed that the temperature at which an alloy first lost all indication of strength, as determined on the tensometer, appeared to be approximately that at which the alloy began to collapse under its own weight and ceased to be coherent or give the characteristic brittle, hot-short fracture. The indication of this was that after testing the samples had a glazed and rounded fracture instead of the sharp, intercrystalline one characteristic of lower temperatures.

Thus, it appears that in order to produce a brittle fracture (at least, at normal rates of straining) the material must be extended within the temperature range above the solidus over which the alloy has an appreciable strength. Below the solidus the ductility is high, and above the temperature at which all strength is lost so much liquid is present that the material deforms, and then, provided that the quantity of liquid is not too great, breaks with a typical rounded type of fracture.

The temperature range beyond the solidus is, therefore, probably of great importance in determining the general hot-shortness characteristics of an alloy. If, for sake of simplicity, the systems be considered to be in equilibrium, then in castings and welds it would be expected that the greater this temperature range the greater would be the thermal contraction in portions at a more advanced stage of solidification during the time required for the temperature of the hottest part to decrease below the solidus, and, consequently, the greater would be the likelihood of cracking.

The rate of straining of 0.25 in./min. was chosen with the object of applying the experimental results to the practical problem of hot-shortness. From a number of observations on the casting and welding of aluminium alloys, the rate at which intergranular separation or crack formation took place was almost invariably found to be between 0.05

and 1.0 in./min., and 0.25 in./min. was taken as an average rate of separation.

A further investigation, undertaken as a continuation of the present one, in which the hot-shortness of the aluminium-silicon alloys has been determined by practical casting and welding experiments, has justified the importance attributed to the temperature range above the solidus. In this research, it has been shown that the tensile properties of the aluminium-silicon alloys in the region of the solidus can be correlated with their casting and welding behaviour.

From Fig. 9, in which the extent of the temperature range above the solidus (the hot-short temperature range) over which the alloys are coherent has been plotted against chemical composition, it will be seen that a maximum occurs at 1.8% silicon, the approximate composition

TABLE XV.—*Temperature Range Calculated from Data obtained by Vero.*

Alloy.	Temperature Range, °C.
Super-purity aluminium	10
0.22% silicon	15
1.6% silicon	75
1.88% silicon	76
12% silicon	4

of the maximum temperature range of solidification. This range applies, of course, only to systems at equilibrium or approaching equilibrium. The hot-short ranges of 4° and 2° C. shown by the 8% and 12% alloys, respectively, probably have no great significance and the curve in Fig. 11 has therefore not been drawn through these points.

Contrary to Vero's opinion, it thus appears that there is a possible foundation for the hot-shortness characteristics in the shape of the tensile curves at temperatures in the region of the solidus. In fact, Vero showed the existence of a temperature range above the point of incipient melting over which some of the aluminium-silicon alloys have a measurable strength, but, when unable to link it directly with his hot-shortness casting experiments, abandoned the attempt at correlation. A calculation of the interval between the temperature of incipient melting and that of loss of all strength, estimated from Fig. 1, shows that, with the few alloys on which experiments were carried out, Vero's results are in substantial agreement with those of the present investigation. The calculated temperature intervals are given in Table XV.

Structure of the Alloys in the Hot-Short Temperature Range.

From Fig. 8 it is apparent that the alloys lose coherence in conditions of continuous extension in the tensile test when approximately 40% of liquid is present. Alloys containing 6% or more of silicon appear to have a negligible hot-short temperature range.

The proportion of liquid in some of the samples retaining appreciable strength appeared, at first, to be unbelievably high. On subsequent microscopical examination, however, the reason for this was made apparent. In the original casting, the liquid last to solidify was contained partly in thin films around individual crystals and partly "land-locked" between the multitudinous arms of the original dendrites. During cooling, therefore, the only part of the liquid which could play an effective part in allowing separation of the crystals at their boundaries was that in the form of the thin intergranular films, the "land-locked" liquid being wholly ineffective.

On reheating to a temperature above the solidus and maintaining at this temperature for 10-15 min., as in the conditions of testing, the "land-locked" portions of liquid tended to form themselves into globules, and this was, of course, augmented by a decrease in the amount of silicon in solid solution. The grain-boundary films did not seem to show the same tendency to form globules and with silicon in excess of 1.5% no grain boundaries were observed free from the silicon-rich film.

The presence of "ineffective" liquid in globular form within individual grains is shown in Fig. 13 (Plate II) in the case of the 4% silicon alloy quenched from 580° C. The alignment of the globules indicates their origin and also the orientation of the grains. In this particular area, approximately 25% of the chilled silicon-rich liquid is present at the grain boundaries, the remainder being contained within the grain.

Super-Purity Aluminium.

The strength curve for the super-purity aluminium is worthy of special consideration. In Fig. 5 it will be seen that it is displaced considerably below that of 99.5% aluminium and this is undoubtedly due to the smaller content of both iron and silicon. The shallow, concave curve obtained for aluminium, and the order of strength at the melting point, 180 lb./sq.in. or 0.08 tons/sq.in., can be compared to the results of Lander and Howard,⁵ who obtained a similar curve for pure mercury showing a strength of 0.14 tons/sq.in. at the melting point.

The interesting feature, however, is the way in which the point of embrittlement occurs 4° C. below the true melting point. At 655° C. a bright, intercrystalline type of fracture was observed, which persisted

up to the melting point at 659°C. , as shown in Fig. 7 (Plate I). The elongation decreased to zero at 655°C. and the strength decreased rapidly to zero between 655° and 659°C. The explanation of this pre-melting phenomenon may be the presence of minute quantities of impurities segregating at the grain boundaries in sufficient concentration to bring about a localized decrease of 4°C. in the melting point.

CONCLUSIONS.

It has been shown that in the aluminium-silicon alloys there exists a range of temperature above the solidus over which some alloys have a finite strength or coherence, while at the same time having a negligible ductility. This temperature range, termed the hot-short temperature range, increases in equilibrium conditions with increasing silicon content to a maximum at approximately 1.8% silicon and then decreases to a low value.

It is suggested that the extent of the hot-short temperature range is one of the most important factors in determining the hot-shortness properties of the alloys.

ACKNOWLEDGEMENTS.

The authors express their thanks to Professor D. Hanson, D.Sc., under whose general direction the investigation was carried out, and desire to acknowledge their indebtedness to the Wrought Light Alloys Association, not only for permission to publish the work, but for the many services which it has placed at their disposal.

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METALLURGICAL ABSTRACTS

(GENERAL AND NON-FERROUS)

Volume 13

SEPTEMBER 1946

Part 9

I.—PROPERTIES OF METALS

*Negative Resistance-Temperature Coefficient of Thin Evaporated Films of Bismuth. T. J. Tulley (*Nature*, 1946, 157, (3986), 372).—Thin films of bismuth were evaporated on to microscope slides (which were washed and wiped, but not otherwise cleaned) at a pressure of 10^{-3} mm. in a chamber to which mercury vapour from a diffusion pump had access. In all cases the film showed a negative temp. coeff. of resistance, which decreased with increasing film thickness. Numerical results are given.—G. V. R.

*The Thermal Conductivity of Bismuth at Low Temperatures. S. Shalyt (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (6), 250–252).—[In Russian] See *Met. Abs.*, 1945, 12, 205.—N. A.

*On the Thermal Conductivity of Bismuth. A. I. Akhiezer and I. Ya. Pomeranchuk (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (10), 387–392).—[In Russian] See *Met. Abs.*, this vol., p. 41.—N. A.

*Magnetic Self-Recovery in Cold-Worked Copper. J. Reekie and T. S. Hutchison (*Nature*, 1946, 157, (3998), 807–808).—Rods of high-purity copper were cold worked by drawing through hardened steel dies, giving a 20–35% reduction in area. Specimens were then maintained at various constant temp. and the magnetic susceptibility measured at regular intervals. Results, which are given graphically, show that at -180° C. no measurable change in susceptibility occurs, while at room temp. a change of 5% occurs in 100 hr. At 90° C. a change of 8% occurs in 25 hr. It is concluded that the magnetic susceptibility is strain sensitive and that the results obtained are connected with the self-recovery of the metal. No evidence of ferromagnetic impurity was obtained.—G. V. R.

Calculation of the Elastic Constants of a Single Crystal of Copper. I. Sal'nikov (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (6), 301–311).—[In Russian] A simple method is given for computing that part of the elastic const. c_{11} – c_{12} and c_{44} which is dependent on the so-called energy of overlapping. Using this method the elastic const. of a single crystal of copper are calculated.

—N. A.

*Resistivity of Thin Nickel Films at Low Temperatures. A. van Itterbeek and L. De Greve (*Nature*, 1946, 158, (4003), 100–101).—Continuing work on the electrical resistance of thin nickel films (*Nature*, 1945, 156, 634; *Met. Abs.*, this vol., p. 153), the resistance of such films as a function of temp. was measured down to liquid helium temp. On cooling films thicker than $40\ \mu$, the resistance passes through a reversible minimum and the temp. coeff. changes from positive to negative. The nearer the thickness approaches to $40\ \mu$, the more the minimum in the resistance curve is displaced towards higher temp.

—G. V. R.

*Investigation of the Thomson-Nernst Thermomagnetic Effect in a Nickel Crystal. R. Aniaev (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (11), 629–642).—[In Russian] See *Met. Abs.*, this vol., p. 242.—N. A.

*The Electrolytic Growth of a Single Crystal of Silver. A. T. Vagramyan (*Zhur. Fiz. Khim.*, 1945, 19, (6), 305–313).—[In Russian] In the electro-

* Denotes a paper describing the results of original research.

† Denotes a first-class critical review.

crystallization of silver on an unpoisoned, growing portion of the cathode, the rate of the reaction is limited by the addition of material to that growing portion, while the rate of passivation is governed by the addition of passivators.

—N. A.

On the Displacement of the Critical Temperature of Supraconducting Tin by Extension. N. Alexeevsky (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (6), 244-245).—[In Russian] See also *Met. Abs.*, this vol., p. 42. The relation between temp. and the change of induction and electrical conductivity of a single crystal of tin is demonstrated. On increasing the stress producing extension of the crystal, the critical temp. is also increased.—N. A.

***Dilatometric Study of Pure Zinc.** A. A. Bochvar and Z. A. Sviderskaya (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (3), 209-211).—[In Russian] An experimental study was made of the direction taken by dilatometric curves obtained during the heating of deformed single-crystal specimens of zinc.

—N. A.

Creep of Metals. N. P. Allen (*Nature*, 1946, 157, (3989), 469-471; see also *Engineering*, 1946, 161, (4182), 233-235; (4183), 258-259; *Engineer*, 1946, 181, (4701), 148-149; (4702), 170-171; and *Chem. and Ind.*, 1946, (10), 109).—A summary report of a conference held at the Royal Society in Feb. 1946. Papers were presented by several authors and there was a general discussion. E. N. da C. Andrade reviewed the general characteristics of deformation and creep of single crystals of metals, directing attention to need for working at constant stress rather than constant load. H. J. Tapsell referred to engineering problems necessitating long-time tests and outlined an extrapolation technique which enables use to be made of data from short-time tests. E. Orowan dealt with a new theoretical approach to the problem of creep and A. McCance proposed a mathematical approach enabling predictions to be made from short-time tests. N. P. Allen reviewed the influence of metallurgical structure, and the metallurgical control of creep.—G. V. R.

Strength and Ductility. Maxwell Gensamer (*Metal Progress*, 1946, 49, (4), 731-734).—Twentieth Campbell Memorial Lecture. The measured mechanical properties of metals are reviewed in relation to their behaviour in service and their dependence on chemical composition and structure. Attention is primarily directed to the effect of alloying elements, present either in solid solution or as undissolved compounds, on the mechanical properties of steel, but the principles involved are of general interest. The main conclusions are: *Strength* is not sensitive to local variations in composition and structure. It is closely related to composition in solid solutions, but depends in aggregate structures on the average spacing between particles, the shape of the latter being important only in so far as it influences the mean spacing. Fine dispersions influence strength to a much greater extent than do dissolved elements. *Ductility* is much influenced by local conditions, but unless local ductility is exceptionally low, local deformation has a much greater effect on energy absorption. The distribution of deformation can most advantageously be improved by modifying the plastic properties of materials, except at high strengths where the intrinsic ductility is sufficiently low to be an important factor in failure.—P. R.

The "Degree of Loosening" of Solids. K. Lichtenecker (*Z. Elektrochem.*, 1942, 48, 669-671; *C. Abs.*, 1943, 37, 5897).—Cf. *Z. Elektrochem.*, 1943, 49, 174; *Met. Abs.*, 1945, 12, 242. The "degree of loosening," λ , is defined as $Q - (E/2)/E$, where Q is the integral of the heat capacity from abs. zero to the melting point and E is the sum of Q and the energy of fusion. For many solids, λ is positive and is relatively large for the softer metals. The "mobility" and coeff. of self-diffusion are related qualitatively to λ .

Analysis of the Process of Absorption of Gases by Metals. I.—Characteristics of Gas-Metal Systems. A. D'yakonov and A. Samarin (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (9), 813-820).—[In Russian] A general account.—N. A.

***On the Structure of Thin Solid Films.** Nicolas Cabrera (*Compt. rend.*, 1946, 222, (16), 950-952).—Observations with an electron microscope revealed a discontinuous structure in thin solid metal films at room temp. For thicknesses of less than about 15 m μ , the film contained holes; thinner films consisted of independent grains. These discontinuities are due to surface tension. If v_0 is the volume of material per c.c. of support and this material is distended uniformly, a continuous film of thickness $h_0 = v_0$ results. The stable structure is that for which the free energy is a minimum. The expression for the free energy for the volume v_0 contains a term proportional to v_0 and always the same, and another proportional to the surface and differing from one structure to another. The equation for the energy, E , of the film is worked out mathematically. The determination of the surface entropy is more difficult and rests on an assumption to be discussed later. The free surface energy, $F = E - TS$, where T is the temp. and S the surface. There are two critical temp.: the first, T_1 , from which the hole structure will be the more stable, the second, T_2 , from which the grain structure will always replace the hole structure. At a certain temp., T_0 , there will be critical thicknesses, h_{01} and h_{02} , for which T_0 is equal to T_1 and T_2 . Hence, for the intermediate case, $h_0 > h_{01}$, there is a continuous film; for $h_{01} > h_0 > h_{02}$, there is the hole structure; and for $h_{02} > h_0$, there is the grain structure.—J. H. W.

***Infra-Red Emissivity of Metals at High Temperatures.** D. J. Price (*Nature*, 1946, 157, (3997), 765).—By measurement of the intensity of radiation from the inside and outside of a hollow cylindrical specimen, the emissivity of various metals was measured at 1000°-1500° C. and wave-lengths of 1.0-4.5 μ . Results show that at wave-lengths less than 0.93 μ for platinum or 1.0-1.5 μ for iron the temp. coeff. of emissivity is zero. In the far infra-red region the coeff. is equal to half the temp. coeff. of resistivity. Over a range of intermediate wave-lengths the relation between wave-length and temp. coeff. of emissivity is linear.—G. V. R.

An Extension of the London Theory of Supraconductivity. M. v. Laue (*Ann. Physik*, 1942, 42, 65-83; *C. Abs.*, 1943, 37, 5633).—Two independent mechanisms of conductivity assumed in London's theory are harmonized relativistically; certain difficulties are thus removed.

On the Measurement of the Depth of Generation of Secondary Electrons in a Metal. A. Kadyshovich (*Zhur. Eksper. Teoret. Fiziki*, 1945, 15, (11), 651-654).—[In Russian] See *Mel. Abs.*, this vol., p. 244.—N. A.

Theory of Metals. P. Gombás (*Nature*, 1946, 157, (3994), 668-669).—Theoretical. An expression is derived for the lattice energy of a metal, which is considered statistically as a lattice of positive metal ions and a uniformly distributed electron-gas consisting of the valency electrons. From this expression the fundamental const. of the alkali metals are deduced and are in fair agreement with experiment.—G. V. R.

II.—PROPERTIES OF ALLOYS

***Micro-Hardness Study of the Structure of Cast Aluminium-Copper and Aluminium-Silicon Alloys.** A. A. Bochvar and O. S. Zhadaeva (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (10/11), 1089-1092).—[In Russian] It is shown that the hardness of the crystals of primary solid solution in the region of the eutectic is not uniform, but depends on composition. The variation is reduced by homogenization but is not entirely eradicated.—N. A.

***On the Mechanical Properties and Corrosion-Resistance of Cold and Warm Age-Hardened Aluminium-Zinc-Magnesium-Copper Forging Alloys.** K. L. Dreyer and H. J. Secinan (*Aluminium*, 1944, 26, (5/6), 76-82).—The mechanical properties and corrosion-resistance of six alloys of aluminium containing 0.7% manganese with: (1) zinc 3.7, magnesium 2.5, and copper 1.5, 2.0, and 2.5%; and (2) zinc 4.5, magnesium 1.5, and copper 1.5, 2.0, and 2.5%, together with the effects thereon of various solution temp., ageing temp., and times, are shown in tables and graphs. It was established that as the values for the variable factors increase, the values for 0.2% proof stress and tensile strength increase, the max. values being obtained by solution-treatment at 500° C. for 30 min., followed by water-quenching and ageing at 50°-100° C. for 10-90 days, when the 0.2% proof stress becomes 25-30 kg./sq. mm. and the tensile strength becomes 45-50 kg./sq. mm., the corresponding figures for the quenched alloys being 12-15 kg./sq. mm. and 30-35 kg./sq. mm., respectively. When unpolished and polished sheets of the alloys were exposed to a 3% salt solution for 0-120 days (a) polished sheets corroded less than unpolished sheets; (b) both series of alloys had almost equal layer-corrosion-resistance; and (c) the second series of alloys had superior stress-corrosion-resistance, although the resistance decreased with increase of copper content, solution temp., ageing temp., and time. The best alloy is that with zinc 4.5, magnesium 1.5, copper 1.5, and manganese 0.7%.—E. N.

***The Relationship Between the Working Properties and Final Rolling Reduction Before Annealing of Sheets of Duralumin-Type Alloys Containing 3% and 4% of Copper, Made from Chill- and Continuously Cast Billets.** Hans Hartmann (*Aluminium*, 1944, 26, (5/6), 84-88).—Chill-cast and continuously cast billets of Duralumin-type alloys containing 3 and 4% copper were cold rolled, annealed, subjected to further rolling reductions of 10-70%, and re-annealed, after which their grain structure and deep-drawing and stretching properties were studied. Heavy rolling reductions produced coarse grains on annealing, causing a decrease in the working properties of the alloy and this tendency was more pronounced (a) with the chill-cast metal, and (b) with the alloy containing 4% of copper. To obtain the best material the final rolling reduction should be in the region of 30%.—E. N.

***Metallurgical Requirements for the Production of Corrosion-Resisting High-Strength Sheets of Aluminium-Zinc-Magnesium Alloys.** W. Patterson (*Metallwirtschaft*, 1944, 23, (18/21), 161-173).—An extensive study was made of the influence of the degree of cold work, heat-treatment, and additions of up to 0.3% of beryllium, cerium, chromium, thorium, titanium, vanadium, and zirconium ("stabilizers") on the mechanical, recrystallization, and corrosion-resistance properties of aluminium alloys with up to 0.5% of manganese and <8% of zinc plus magnesium. The stabilizers broaden the recrystallization range and displace it to higher temp.; diffusion takes place more slowly and precipitation at the grain boundaries is hindered. These are factors which increase the resistance of the alloys to layer- and stress-corrosion, the optimum conditions being obtained with additions of 0.3% chromium or 0.1% vanadium, a final rolling reduction of 50-75%, and a solution-treatment at 470°-530° C. for 20-30 min., followed by quenching and warm ageing at 110°-120° C. for 4 hr. The additions, however, produce alloys of inferior working qualities; such alloys are sensitive to the presence of iron, are subject to cracking when continuously cast, are liable to burst when hot rolled, and have a tendency to coarse-grain formation on annealing after heavy rolling reductions. The mechanical properties of the resulting alloys are inferior to those of the straight aluminium-magnesium-zinc series; the tensile strength and 0.2% proof strength are reduced, the elongation increased, and, unless recrystallization is fully completed, the resulting heterogeneous structure has not only variable mechanical properties but is more susceptible to stress-corrosion. It is con-

cluded that in alloys of this type a combination of high mechanical and corrosion-resistance properties is not fully compatible.—E. N.

Aluminium Alloys 2S, 3S, and 52S (Wrought, Non-Heat-Treatable Types).—(*Machine Design*, 1944, 16, (7), 149–153).—A comprehensive review of the properties of three commercial aluminium alloys giving tables showing mechanical and physical properties of the material in various degrees of hardness, nominal compositions, high-temperature properties, effect of drawing on mechanical properties, commercial tolerances for sheet, plate, wire, rod, and bar, official designations, accounts of welding properties, corrodibility, and recommendations for piercing and riveting.—P. R.

***Control of Shrinkage Porosity and Mechanical Properties in Chill-Cast Phosphor-Bronzes, Leaded Bronzes, and Gun-Metals.** W. T. Pell-Walpole (*Metallurgia*, 1946, 34, (200), 85–90; (201), 155–161).—The variables in the chill-casting process which affect the extension and distribution of shrinkage porosity were examined for phosphor-bronzes, gun-metals, and leaded bronzes. The rate of pouring is the most important factor and freedom from the most harmful types of shrinkage porosity is obtained when the rate is the lowest at which the mould will fill without cold shuts. This critical rate is related to the shape and size of the ingot by the formula $R = KP$, where R is the pouring rate, P is the periphery of the ingot cross-section (or, in the case of cored sticks, the sum of the peripheries of mould and core), and K is a const. which, for constant mould conditions and pouring temp., depends on the alloy composition. Values of K were determined for all standard bronze compositions. The mechanical properties of these bronzes vary, in general, with casting conditions in the same manner as density—i.e. optimum values are obtained uniformly across the section when the pouring rate is the minimum which will fill the mould without cold shuts. Slight variations in properties are also produced by the effect of pouring and mould temp. on the distribution of the brittle compound phases.—J. W. D.

***Welded Magnesium Alloys and Their Alternating Tension-Compression Fatigue Strength.** A. Thum and R. Zoege von Manteuffel (*Metallwirtschaft*, 1944, 23, (22/26), 215–220).—A study was made of the static tensile properties and alternating tension-compression fatigue strength of unwelded and welded test-pieces of alloys of magnesium with (a) aluminium 3.18, manganese 0.40, and zinc 0.90%; (b) aluminium 6.70, manganese 0.19, and zinc 0.70%; and (c) manganese 1.93%. After welding, the static and fatigue properties of alloy (a) were slightly superior to those of (b), while both (a) and (b) had properties greatly superior to those of (c). For example, the fatigue strength of alloy (a) after 50×10^6 cycles was ± 4.2 kg./sq. mm. (unwelded ± 7.6 kg./sq. mm.), while that of (c) was ± 2.6 kg./sq. mm. (unwelded ± 4.6 kg./sq. mm.). Examination of the weld structure showed that (i) microporosity was present in all cases; (ii) the failure of alloy (c) was explained by the coarse, dendritic crystal formation which had taken place in the weld seam. It is concluded that steel is not replaceable by light alloys when alternating tension-compression fatigue stresses have to be endured.—E. N.

The Structure of Nickel-Aluminium Magnet Alloys and the Control of Brittleness. Alan Torry (*Metallurgia*, 1946, 34, (201), 147–153).—After an historical survey of the progress in explaining the cause of high coercivity in the ternary alloys of nickel, iron, and aluminium, an attempt is made to show how the factors responsible for coercivity are also associated with a weak intergranular structure. It is suggested that excessive brittleness in these alloys may be directly attributed to the differences in the rates of contraction or expansion of the phases already present or being precipitated. A study of the thermal dilatation of Alni, Alnico, and Alcomax alloys of varying compositions revealed the magnitude of the influence of such elements as aluminium, copper, and titanium in causing departure from uniform expansion or con-

traction, both after casting and in subsequent heat-treatments. By controlling the relative proportions of all the elements present, with improved knowledge of their separate effects on the magnetic characteristics, heat-treatment, and dilatation behaviour, economic production has been improved and mechanical stability obtained without sacrifice of magnetic performance.—J. W. D.

Inconel, Wrought and Cast. — (*Machine Design*, 1944, 16, (8), 113–116).—The mechanical properties of Inconel in various conditions of work and heat-treatment are tabulated, together with information on creep strength of hot-rolled rod, high-temperature and low-temperature properties, physical const., fabrication (machinability, drilling, reaming, tapping, perforating, &c.), corrosion-resistance, annealing, tolerances for cold-rolled sheet and for hot-rolled rods and flats, and designations.—P. R.

***The System Platinum-Mercury.** I. N. Plaksin and N. A. Suvorovskaya (*Izvest. Sekt. Platin*, 1945, 18, 67–76).—[In Russian] Cf. *Met. Abs.*, 1941, 8, 225, 298. Thermal, X-ray, and micrographic analyses were carried out and the solubility of mercury in platinum studied up to 200° C. The results reveal the existence of a solid solution of mercury in platinum up to 23 at.-% and three intermetallic compounds: β (73–76 at.-% platinum), γ (60–70 at.-% platinum), and δ (45–55 at.-% platinum). These compounds melt congruently at 485.1°, 236°, and 159.1° C. They correspond to Pt_3Hg , Pt_2Hg , and $PtHg$.—N. A.

***The Phenomenon of Super-Plasticity in Zinc-Aluminium Alloys.** A. A. Bocharov and Z. A. Sviderskaya (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (9), 821–824).—[In Russian] On heating above 100° C., alloys of zinc with 20% aluminium show an unusual reduction in hardness and increase in plasticity. The hardness of pure zinc and aluminium at 300° C. is approx. 10 kg./sq. mm., whereas the hardness of the alloy of zinc with 20% aluminium is only 0.6–1.0 kg./sq. mm. at that temperature.—N. A.

***Contribution to the Question of Changes in Materials Under Fatigue Stress.** [Brass, Aluminium, Duralumin, Nickel-Chromium-Iron Alloy, Steels.] A. Karius (*Metallwirtschaft*, 1944, 23, (48/52), 419–434).—Experiments were carried out on specimens of brass, aluminium, Duralumin, iron-19% nickel-25% chromium alloy, and various steels to determine the influence of fatigue stress on the elastic modulus and damping capacity. Plain and notched test-pieces were subjected to alternating-bend tests in a Schenck fatigue-testing machine and, after various numbers of cycles, the alterations in damping capacity and elastic modulus were determined in a Förster machine. It was shown that with all polycrystalline materials changes due to fatigue are of a similar nature and follow a similar course throughout the test. Initially there is a sharp increase in damping capacity and a corresponding decrease in elastic modulus due to cold working; these effects almost vanish with a slight increase in stress and duration of test but recur, as a result of the formation of macroscopic cracks, shortly before the onset of fracture. The initial changes are dependent on the volume of stressed material in the specimen and are not limited to the points at which subsequent failure takes place; changes in properties just before fracture are independent of the volume. Notched specimens exhibit only slight changes in properties; no volume effect occurs as cracks are formed in the region of the notch. Preliminary static or dynamic stressing causes a reduction in damping and modulus changes by fatigue testing, but high preliminary stressing brings about a reversal of this effect. The initial changes occurring in fatigue testing are partly eliminated by resting at room temp. and are completely removed by heat-treatment at 150° C., showing that such changes are the result of lattice displacement and not plastic deformation, the low-temp. annealing relieving the internal stresses set up. In the case of individual crystals, submicroscopic cracks are only eliminated by heating above the recrystallization point.—E. N.

***On the Temperature Curve of the Commencement of Linear Contraction in Binary Alloys.** A. A. Bochvar and V. I. Dobatkin (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (1/2), 3-6).—[In Russian] The curve denoting the commencement of linear contraction in a series of alloys divides the two-phase region between the liquidus and the solidus into two parts: (i) the upper, consisting of liquid with traces of the solid phase, and (ii) the lower, consisting of solid with isolated inclusions of the liquid phase. An apparatus is described with which this curve can be determined and with which experiments were carried out on alloys of the magnesium-aluminium and aluminium-silicon systems. The results are correlated with the microstructures.—N. A.

III.—STRUCTURE

(Metallography; Macrography; Crystal Structure.)

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

Distinguishing the Common Aluminium Alloys. Vernon H. Patterson (*Metal Progress*, 1946, 49, (2), 346-347).—Brief note. Common wrought aluminium alloys are classified into three groups, which contain (i) copper 4% or more; (ii) neither copper nor chromium; and (iii) chromium 0.25%. The groups may be distinguished by the blackening of (i) in warm NaOH solution (removable by nitric acid) and the golden colour produced in (iii) by anodizing. Methods of separation within the groups are briefly described.—P. R.

Fractography. J. E. Hurst and R. V. Riley (*Metal Progress*, 1946, 49, (1), 96-97).—It has been claimed by Zapffe and Clogg (*Trans. Amer. Soc. Metals*, 1945, 34, 108-141; *Met. Abs.*, 1945, 12, 80) that the study of cleavage facets on fractured metals ("fractography") reveals unusual structures which may throw light on molecular structure, internal stresses, &c. It is pointed out that certain alloys with a tendency to form surface films, e.g. the high-silicon-iron alloys, may show misleading structures which should not be accepted as confirming the results of "fractographic" examination. The method is considered most suitable for the examination of hard, brittle alloys.—P. R.

Estimation of Spatial Grain-Size. William A. Johnson (*Metal Progress*, 1946, 49, (1), 87-92).—A plea is made for the estimation of grain-size in terms of spatial size instead of by observations on plane surfaces. J. points out that even if the spatial grain-size is nearly constant not more than half the grains cut by such planes will come within the range covered by a single grain-size number (A.S.T.M. specification). An account is given of a method developed by E. Scheil for correlating "planar" with "spatial" grain-sizes. The relative area of each size of the grains shown on a micrograph is estimated and from this the relative (statistical) distribution of spatial grains is calculated by means of an expression based on the A.S.T.M. definition of grain-size numbers. The average size of "spatial" grains is then computed from the percentage by volume of the different sizes and the standard deviation is determined in order to give a measure of size uniformity. The procedure is described in detail.—P. R.

Rapid Estimation of Grain-Size in Magnesium Castings. J. P. Ogilvie (*Metal Progress*, 1946, 49, (2), 349).—Brief note. A simplified method of preparing magnesium castings for grain-size determination consists in polishing on emery, wax lap, and "gamal" lap, and etching with 10% tartaric acid solution. This obviates the usual need for solution heat-treatment.—P. R.

Preparation of Micro-Specimens [of Magnesium Alloys] by Strong Etchants. E. M. Savitsky (*Zavod. Lab.*, 1945, 11, (2/3), 192-194).—[In Russian] Micro-specimens of magnesium and its alloys are prepared by grinding on coarse

grades of emery paper and then immersing in two etchants, the first of which removes loose particles and the deformed surface layer while the second reveals the grain boundaries. The compositions of the etching reagents are given.

—N. A.

Electropolish-Attack : a New Metallographic Technique. L. Ya. Popilov (*Zavod. Lab.*, 1945, 11, (1), 60-71).—[In Russian] A review of the method and of P.'s experiments.—N. A.

***X-Ray Examination of Self-Recovery in Metals.** L. L. van Reijen (*Nature*, 1946, 157, (3986), 371).—The recovery of copper filings has been demonstrated by comparison of X-ray photographs obtained at an interval of some months. Only two low-angle rings were recorded on a film placed behind the specimen. The second photograph, however, showed many distinct spots on the reflection rings, indicating the occurrence of a number of relatively large crystallites. Several aspects of the X-ray study of self-recovery are discussed.—G. V. R.

Electron Microscopy. V. N. Vertsner (*Zavod. Lab.*, 1945, 11, (6), 543-554).—[In Russian] A review.—N. A.

X-Ray Methods for the Determination of the Lattice Type and Dimensions of Sub-Microscopic Crystals. G. S. Zhdanov (*Zavod. Lab.*, 1940, 9, (5/6), 566-571; (7), 732-740).—[In Russian] A review.—N. A.

Electron Microscopy in Metallography. B. A. Ostroumov (*Zavod. Lab.*, 1945, 11, (6), 554-561).—[In Russian] A review.—N. A.

Methods for the Precision Determination of the Lattice Dimensions of Polycrystalline Specimens. A. Kh. Breger (*Zavod. Lab.*, 1940, 9, (1), 55-63).—[In Russian] A review.—N. A.

V.—POWDER METALLURGY

Powder, Pressure, and Heat. — (*Fortune*, 1942, 25, (1), 44-47, 120).—A review of powder metallurgy.

***Appraisal of Aluminium Powder for Paint Pigments.** H. C. J. de Decker (*Verfkronek*, 1942, 15, 24-29, 39-42; *Chem. Zentr.*, 1942, 113, (II), 1407; *C. Abs.*, 1943, 37, 6143).—Six different samples of aluminium powder were tested for quality and dispersibility (reflecting power). For a normal spatula sample, a mixture of 80 g. of linseed oil stand oil (viscosity 60 centipoises at 20° C., $d = 0.976$, $n = 1.4912$, acid no. = 12.2) and 95 g. of benzine lacquer is placed in a Pyrex glass cylinder (1.8 × 19 cm.) so that it contains 25 c.c. of this mixture per 10 cm. fluid height. To this is added 2 g. of aluminium powder; the Pyrex cylinder is closed with a cork and shaken for 5 min. at 4 turns per second. A spatula is immersed in the colour so that it is covered up to a height of 10 cm. The time of withdrawal and shaking off of free material adhering to the spatula up to periods of 2 min. is determined with a stop watch. After 2 min. both sides of the spatula are observed, and reflection intensity and total colour intensity are read in mm. upon a line. The dispersibility or reflecting power of aluminium powder corresponds to the average reflection intensity in % of the total colour intensity. The spatula test can be used for determination of the dispersibility of aluminium bronzes.

The Leafing Properties of Aluminium Bronze Powders. H. Th. Mayer (*Farben-Chem.*, 1942, 13, (5), 83-85).

Sintered Powders for Electrical Contacts. Heinz H. Hauser (*Bull. Assoc. Suisse Elect.*, 1942, 33, 29-34; and (translation) *Engineering*, 1946, 161, (4180), 188-190).—Theoretical aspects of arcing are discussed and the properties required of materials able satisfactorily to withstand this phenomenon are considered with reference to sintered materials such as molybdenum-silver, tungsten-silver, and tungsten-copper.—R. Gr.

VI.—CORROSION AND RELATED PHENOMENA

*On the Corrosion-Resistance After Cold and Warm Age-Hardening of Sheets of Aluminium-Copper-Magnesium Alloys with Various Clad Coatings. W. Bungardt (*Metallwirtschaft*, 1944, 23, (22/26), 221-227).—A study was made of the corrosion-resistance of Duralumin after cladding with pure aluminium, aluminium-manganese, and aluminium-magnesium-silicon-manganese alloys. The specimens were solution-treated at 500° C., cold and warm (160° C.) aged for varying periods, immersed in a stirred aqueous solution of 3% sodium chloride and 0.1% hydrogen peroxide, and the coatings compared by measurements of tensile strength, elongation, and rate of diffusion of copper into the coating. Solution-treatment at 500° C. for 30 min. followed by cold ageing for 3 hr. gives materials which have the best corrosion-resistance. After artificial ageing, coatings of pure aluminium, which allow copper to diffuse rapidly into them, give the least protection. The two alloy coatings are of approx. equal value. Although the aluminium-magnesium-silicon-manganese alloy is liable to intercrystalline corrosion, this has no deleterious effect on the resistance of the clad sheet.—E. N.

*Metallurgical Requirements for the Production of Corrosion-Resisting High-Strength Sheets of Aluminium-Zinc-Magnesium Alloys. (Patterson). See p. 316.

*On the Mechanical Properties and Corrosion-Resistance of Cold and Warm Age-Hardened Aluminium-Zinc-Magnesium-Copper Forging Alloys. (Dreyer and Seeman). See p. 316.

*Attack of Various Superheated Steam Atmospheres Upon Aluminium-Bronze Alloys. A. P. C. Hallows and E. Voce (*Metallurgia*, 1946, 34, (201), 119-122).—Based on part of a report to the British Non-Ferrous Metals Research Association. The action of superheated steam on complex aluminium-brasses was studied as a result of some failures of these alloys when used industrially in certain superheated-steam atmospheres under service conditions involving stress. Pure steam was found to be harmless. Prolonged exposure to an atmosphere of steam contaminated with a small amount of sulphur dioxide produced attack in the surface layers of aluminium bronzes at elevated temp. Preferential oxidation of aluminium to alumina occurred and the oxide remained embedded in a copper-rich matrix. A volume increase of more than 9%, leading to rupture of the surface layers, accompanied the reaction. The attack was more severe at higher temp. and with higher concentrations of the contaminating agent in steam, also when the aluminium content of the alloy was increased. Attack of the same nature was obtained by using chlorine as the contaminating agent in steam. Specimens of 60 : 40 brass in the as-cast condition were not attacked after similar treatment but, in certain conditions, cast tin bronze was subjected to localized attack, after which stannic oxide was found to be embedded in a copper-coloured matrix.—J. W. D.

*Electrochemical Investigation of the Corrosion of Metals [Iron, Lead] in Acid Media in the Presence of Oxidizing Agents. I. Oknin (*Zhur. Priklad. Khim.*, 1945, 13, (9/10), 494-504).—[In Russian] An electrochemical study was made of the corrosion of iron and lead in acid media containing oxidizing agents. Factors studied were the rate of corrosion, the potential of the corroding metal, the potential of a platinum cathode as a function of cathode c.d., the electrical conductivity of the systems platinum-medium-platinum and corroding metal-medium-corroding metal, and the hydrogen-ion concentration. It is shown that the rate of corrosion of the metal is determined by the magnitude of its potential during corrosion. A theoretical survey of the problem is made.—N. A.

***Study of A.C. Sheath Currents and Their Effect on Lead Cable Sheath Corrosion.** C. M. Sherer and K. J. Granbois (*Elect. Eng.*, 1945, 64, (5), 264-268).—Two cases of A.C. electrolysis are discussed, in which the lead sheaths on single-conductor power cables were corroded through. Field observations and laboratory tests are described which show that at first the corrosive action was accelerated by rectified alternating currents which were induced in the sheaths by the load current. Several corrosion-prevention methods are given.
—E. V. W.

Further Data on Corrosion-Test Equipment. [Corrosion of Galvanized Pipe by Water.] Wm. B. Schworm (*J. Missouri Water Sewerage Conf.*, 1943, 14, (3), 26; *C. Abs.*, 1943, 37, 6780).—Water having carbonate alkalinity corrodes galvanized pipe, to give a basic zinc carbonate.

Selecting Materials for Corrosion-Resistance. L. G. van de Bogart (*Machine Design*, 1945, 17, (3), 165-168).—A chart is given comparing the resistance of iron and steel, Ni-resist, 18-8 stainless steel, Monel metal, nickel, red brass, acid-resisting bronze, and aluminium to corrosive attack by a large number of materials, the concentration and approx. degree of purity being indicated in each case. Special precautions to be taken and recommendations regarding use are appended in some instances.—P. R.

On the Stress-Corrosion of Metals. A. Beerwald (*Metallwirtschaft*, 1944, 23, (18/21), 174-187).—Methods of determining the stress-corrosion properties of materials are illustrated, and their application to the testing of steels, non-ferrous alloys of aluminium, copper, magnesium, nickel, zinc, and the precious metals are described. Existing knowledge of the nature and incidence of stress-corrosion is reviewed, together with its relationship to (1) other mechanical properties, especially hardness; and (2) the crystal state. Trans- and intercrystalline stress-corrosion are referred to and illustrated by photomicrographs.—E. N.

***Galvanic Corrosiveness of Soil Waters.** Howard S. Phelps and Frank Kahn (*Elect. Eng.*, 1945, 64, (4), 156-159).—Describes a study of the relation of pH of soil waters to galvanic action between couples of lead, copper, iron, and carbon.—E. V. W.

***Crack-Heal Mechanism of the Growth of Invisible Films on Metals.** U. R. Evans (*Nature*, 1946, 157, (3996), 732).—The invisible films formed on unheated iron exposed to air were studied by using a sodium-carbonate-bicarbonate buffer solution to reveal major discontinuities as rust spots. Specimens were exposed alternately in air and in this solution and records of the discontinuities taken between each exposure. Records were taken by placing filter paper moistened with the buffer solution in contact with the specimens and developing by conversion of the rust spots to Prussian Blue. Results obtained show that the primary effect of air-exposure is to heal existing pores and the secondary effect is to introduce new ones, owing to strains introduced on oxidation. In addition to chemical blockage, produced by oxidation at the base of an already existing pore, mechanical blockage may occur anywhere in a pore. The chance of mechanical blockage increases with film thickness, leading to a decrease in the rate of oxidation. The logarithmic film-thickening law may be explained by these considerations.—G. V. R.

Hydrogen Overvoltage as a Factor in the Corrosion of Metallic Couples. T. P. Hoar (*Nature*, 1946, 157, (3987), 408-409).—The results reported by Le Brocq and Cocks (*Nature*, 1945, 156, 536; *Met. Abs.*, this vol., p. 172) are interpreted in terms of potential/current diagrams. The usefulness of such constructions is pointed out.—G. V. R.

Apparatus for the Determination of Corrosion Losses by the Method of Hydrogen Evolution. E. I. Gurovich (*Zavod. Lab.*, 1940, 9, (4), 484).—[In Russian] Descriptive.—N. A.

VII.—PROTECTION

(Other than by Electrodeposition.)

Anodizing Aluminium (Theory and Application). Gilbert C. Close (*Indust. Finishing*, 1945, 21, (4), 42, 44, 46, 48, 52, 53).—C. briefly reviews the nature of the anodizing process and gives a short account of the coating produced. A tabulated summary is given of the usual electrolytes, concentrations, c.d., duration of treatment, solution temp., rinsing required, thickness and continuity of film, and suitable materials for the equipment required in the Alumilite and chromic acid processes. Many of these data are also considered in detail.

—P. R.

***On the Corrosion-Resistance After Cold and Warm Age-Hardening of Sheets of Aluminium-Copper-Magnesium Alloys with Various Clad Coatings.** (Bundgaldt). See p. 321.

Finishing Aluminium with Paint Coatings [—II]. Robert I. Wray (*Indust. Finishing*, 1945, 21, (3), 66, 69, 70, 72, 74, 76).—See also *Met. Abs.*, this vol., p. 173. The choice of primers depends on the purpose of the part and on its conditions of use: synthetic resin coatings are suitable for decorative work, pigmented alkyd-base primers for domestic equipment, alkali-resistant phenolic-base primers for washing machinery, and zinc chromate-synthetic resin mixtures for parts likely to be exposed to industrial atmospheres. Other pigments used in primers include strontium and calcium chromates, zinc tetroxychromate, ammonium ferrous phosphate, and a mixture of zinc dust and aluminium powder (in linseed oil). Primers are recommended for different classes of work, with special reference to aluminium and its alloys. Coatings for high corrosion-resistance should preferably be non-pigmented. Special precautions are indicated to obviate corrosion at joints between dissimilar metals.—P. R.

Chromium Plating by Diffusion—a New Process. Walter Eckardt (*Z. Metall- u. Schmuckwaren-Fabrik. Verchrom.*, 1942, 23, 152; *Chem. Zentr.*, 1942, 113, (II), 456; *C. Abs.*, 1943, 37, 6630).— CrCl_3 is precipitated on ceramic material and the articles to be plated are packed in the ceramic material and heated to 950°C . The CrCl_3 is decomposed and chromium diffuses into the metal. High carbon content of iron hinders the diffusion of chromium.

***One-Sided Tinning of Sheet Iron.** I. S. Vylegzhanin (*Vestn. Inzhn. Tekhn.*, 1941, 45–46; *Chem. Zentr.*, 1942, 113, (I), 924–925; *C. Abs.*, 1943, 37, 6631).—In order to conserve tin, tests were made of a bilateral, electrolytic method of tinning sheet iron by the use of alkaline baths and cloth-covered tin anodes. Non-porous, matte tin coats (0.18 g./100 sq. cm.) were obtained that were spotty and unsatisfactory. In one-sided hot tinning (the other side being covered with synthetic-resin lacquer), the flux adheres to the lacquered side of the sheet. Diminished roll velocities cause greater burning of the tin, so that there is no saving of tin in spite of the thinner coats. Velocities of 10–14 r.p.m. for a bilateral hot tinning gave good results (0.3–0.45 and as little as 0.2 g. of tin/100 sq. cm. of sheet iron). When two iron sheets, welded together at the edges, were hot tinned, the enclosed air expanded on heating and caused the sheets to bulge. A special welding machine is described. The product, however, is well suited for the manufacture of tin cans for food. Coatings of lacquer are also helpful. Soldering tests with tin-lead solder and alcoholic rosin solutions were unsuccessful; this alloy can be used successfully only with ZnCl_2 . Experiments made to replace HCl by AcOH likewise gave unsatisfactory results; however, AcOH can be used in alcoholic solution for soldering.

On the Tendency for Spangle Formation in Galvanizing. — (*Chem.-Zeit.*, 1944, 68, (6), 108-109).—Lead and iron are both detrimental to the working of the bath. Tin, up to about 1.5%, produces white, feathery crystals, but quantities in excess of this give large spangles; it tends to counteract the effects of lead and iron, but reduces the corrosion-resistance of the coating. 0.1-0.5% cadmium improves ductility. Antimony gives small crystals with a bluish sheen and reduces the corrosion-resistance and adherence of the coating, which can be overcome, to some extent, by the addition of <0.001% aluminium. These facts hold good for bath temp. of 450°-475° C., but as the temp. increases the crystal size decreases until at 510°-520° C. no spangle formation occurs. The spangle size and adherence of the coating are also influenced by (1) initial temp. and surface cleanliness of the material being coated; (2) time of immersion; (3) the interval between galvanizing and quenching; and (4) rate of quenching.—E. N.

***The Formation of Dross in Hot-Dip Galvanizing.** Phillip M. Fisk and F. F. Pollak (*Sheet Metal Ind.*, 1946, 23, (227), 490-492, 494).—Results are given of a study of the effect of the pickling operation, using hot sulphuric and hydrochloric acids, on the formation of dross in galvanizing. Variables considered included metal surface condition, type of acid, quantity of iron in the pickling solution, and variables in the washing tank.—R. Gr.

***The Magnitude of the Protection from Corrosion Afforded Zinc by Amalgamation.** C. Drotschmann (*Batterien*, 1940, 9, 36-37; *Chem. Zentr.*, 1941, 112, (I), 2007; *C. Abs.*, 1943, 37, 5359).—Samples of a standard commercial sheet zinc 0.33 mm. thick, and of zinc strip 0.25 mm. thick, were amalgamated in a boiling solution of 100 g. NH_4Cl , 10 c.c. 3.5% HCl, and 0.17 g. HgCl_2 in 1000 c.c. water. These specimens and untreated specimens were exposed to the action of 3.5% HCl. The decrease in thickness of the amalgamated sheet zinc was 0.03 mm. after 24 hr. in the acid; that of the untreated sheet zinc was 0.23 mm. for 3 hr. treatment in the acid. Neither the amalgamated nor the untreated zinc strip showed any reduction in thickness.

Protection by Zinc and Surface Coatings. K. Voss (*Werkstatt u. Betrieb*, 1942, 75, 112-116; *Chem. Zentr.*, 1942, 113, (II), 1174; *C. Abs.*, 1943, 37, 5359).—A review of established uses of varnishing, galvanizing, dipping, and electrolytic oxidation.

Metallizing as a Method of Waterworks Maintenance. C. E. Palmer (*Water and Sewage*, 1943, 80, (12), 20, 45; 81, (1), 16-18, 36-37; *C. Abs.*, 1943, 37, 6780).—A description of the plant at Erie, Pa., U.S.A. All steelwork is now metallized with zinc for protection against corrosion.

Protective and Decorative Coatings for Metals. A. F. Brockington (*Sheet Metal Ind.*, 1946, 23, (231), 1355-1362).—After a brief historical review, a general review is given of the use of paint, enamel, metals, and other substances to protect and decorate articles. Reference is made to the many various methods. Electrodeposition of metals is dealt with in more detail.—R. Gr.

VIII.—ELECTRODEPOSITION

Practical Experiences with Hard Chromium-Plated Cutting Tools. N. N. Sawin (*Maschinenbau*, 1942, 21, 11-14; *Chem. Zentr.*, 1942, 113, (II), 1174; *C. Abs.*, 1943, 37, 5659).—Hard chromium plate increased tool life three- to eight-fold; the use of chromium on cutting tools is limited to a tool operating temperature of 600° C. and is governed by chip pressure. The life of high-speed steel twist drills, 10-24-mm. dia., in cutting steel having a tensile strength of 57,000-85,000 lb./sq. in., was increased four-fold. Improvements with reaming and lathe tools were limited to certain applications. Tool life increases with chromium thickness up to 15 μ , and decreases with thicker deposits.

Chromium-Plating Practice in the Facing of Press Dies. E. A. Ollard and E. B. Smith (*Sheet Metal Ind.*, 1946, 23, (230), 1129-1141, 1159).—The advantages of chromium plating press dies and the technical difficulties of carrying out the operation satisfactorily are given. The dies should be within certain limits of hardness, pickled and demagnetized if necessary before plating. Practical details of the plating operation applied to materials of various compositions and shapes are described.—R. Gr.

On a Method for the Study of the Mechanical Properties of Chromium Coatings. F. F. Vitman and N. N. Davidenkov (*Zavod. Lab.*, 1945, 11, (9), 844-852).—[In Russian] Describes a method for the determination of the normal elastic modulus of chromium in coatings, the elastic limit in tension, the adhesion of the coatings to the basis metal, and their resistance to cutting action.—N. A.

Developments in Brass Plating. A. Pollack (*Chem.-Zeit.*, 1944, 68, (10), 191-192).—A review of developments in brass plating during the past 10 years. The composition of typical rapid-plating baths and their operating conditions are described. White brass deposits of high zinc content may be obtained by increasing the zinc cyanide content and pH value of the bath. By suitable additions of polyvinyl alcohol, piperonal, and molybdc oxide, polished surfaces are produced which are hard, free from porosity, tarnish-resisting, and which will protect iron from rusting. The use of brass as an intermediate deposit between chromium and nickel, as well as between chromium and steel, is discussed.—E. N.

Porosity Formation in Nickel Plating and Its Removal. Edmund R. Thews (*Chem.-Zeit.*, 1944, 68, (9), 159-162).—Pits and dimples in nickel plating are due to the formation of hydrogen bubbles at the cathode and the presence of foreign matter in the solution. As the working period of the bath increases, the density and surface tension of the solution increase while the viscosity decreases, these being factors which favour the adherence and prevent the dispersal of the gas bubbles. They can be overcome by making the following adjustments to the electrolyte: (1) keeping it agitated; (2) adding water at intervals, so that the density is kept constant; (3) increasing the temp., and working at 20°-65° C.; (4) decreasing the c.d.; (5) increasing the acid content, so that the pH value is 2.6-3.5; and (6) decreasing the surface tension. Foreign matter can be excluded from the bath by making additions of oxidizing agents and removing organic chemicals, sulphur, iron, &c., by continuous filtration or treatment with charcoal; the colloids can be stabilized by introducing sodium or potassium salts. In these conditions porosity and dimpling practically vanish.—E. N.

The Uses of Palladium and Platinum for Electrodeposits. M. A. Klochko and Z. S. Medvedeva (*Izvest. Sek. Platiny*, 1943, 19, 103-118).—[In Russian] A review.—N. A.

***Thickness Measurement of Thin Coatings by X-Ray Absorption.** H. Friedman and L. S. Birks (*Rev. Sci. Instruments*, 1946, 17, (3), 99-101).—The thickness of metal or pigment films (10^{-5} - 10^{-2} cm. thick) on a fine-grained polycrystalline metallic base may be determined by measuring, with a Geiger counter, the intensity of an X-ray beam reflected at a Bragg diffraction angle from the base and observing the absorption due to the double transmission through the coating. When the grain-size of the base is large, it may be necessary to oscillate the specimen, thereby increasing the number of crystallites irradiated. The application of the method to the determination of the thickness of silver electrodeposits on brass is outlined. The accuracy of the method may be checked by measuring the thickness of metal foils mounted in contact with a copper backing.—J. C. C.

Chemical Control of the Thickness of Tin Coatings. M. I. Zibel'farb and L. M. Shteinberg (*Zavod. Lab.*, 1945, 11, (11/12), 1042-1046).—[In Russian]

Five methods were examined. The most convenient for coatings on copper and brass appears to be removal with HCl (1.16). Solutions are suggested for use in the drop and jet methods.—N. A.

Production Plating of Zinc Die-Castings. Herbert Chase (*Iron Age*, 1946, 157, (22), 34–38).—The bright copper-nickel-chromium plating of zinc die-castings on a high production basis is outlined, the types of solution, current densities, and conveying arrangements being described in detail.—J. H. W.

Choice of Plated Coatings. H. L. Farber (*Metal Progress*, 1946, 49, (4), 754).—Factors in the cost of plated coatings include, in addition to the price of the coating material, the need for preliminary treatment of the basis metal, labour and chemicals used in plating, nature of basis metal as it affects plating costs, and range of equipment required to produce a given finish.—P. R.

Methods for the Removal and Recovery of Electrodeposited Coatings. Edmund R. Thews (*Chem.-Zeit.*, 1944, 68, (11), 205–210).—A survey of the chemical and electrolytic methods in use for the removal of deposits of chromium, nickel, copper, zinc, cadmium, tin, silver, and gold from commonly occurring basis metals. The composition and working details of baths are given, attention being directed to the necessity or otherwise for reclamation of the article being stripped.—E. N.

IX.—ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

The Potential Cu/Cu^{++} in Concentrated Copper Sulphate Solutions at Different Temperatures.—II. Friedrich Müller and Hellmut Reuther (*Z. Elektrochem.*, 1942, 48, 682–686; *C. Abs.*, 1943, 37, 5662).—Cf. *Z. Elektrochem.*, 1941, 47, 640–644; *Met. Abs.*, 1942, 9, 360. If a copper electrode is immersed in a Hg_2SO_4 solution, it is covered with a thin film of mercury, which apparently does not amalgamate and has no influence on the electrode potential. With such an electrode, potential measurements can be carried out in air. Results obtained in CuSO_4 solutions of different concentrations and at different temperatures are presented graphically and in tabular form. The most probable value of the N potential of Cu/Cu^{++} is $E_{25} = 0.3400 \pm 0.0003$ V.

Importance of Metal—Metal Contact Potential and the Electromotive Force of Primary Cells. W. G. Burgers (*Chem. Weekblad*, 1942, 39, 198–205, 215–220).—A review.

The Action of Solid and Liquid Metallic Particles in Solutions of Electrolytes. A. Frumkin and V. Levich (*Zhur. Fiz. Khim.*, 1945, 19, (12), 573–599).—[In Russian] A quantitative theory is developed of the action of an external electrical field in the case of an isolated particle, ideally polarized and not fully polarized.—N. A.

Electrolytic Etching of Electrotypes. A. I. Nevzorov (*Poligraf. Proizvodstvo*, 1940, (12), 20–22; *Chem. Zentr.*, 1942, 113, (I), 1435; *C. Abs.*, 1943, 37, 6559).—The etching of zinc electrotypes in a bath containing ZnSO_4 250, Na_2SO_4 50, and $\text{K}_2\text{Cr}_2\text{O}_7$ 25 g./litre was not satisfactory. Better results were obtained with a bath containing ZnCl_2 (200 g.) and CrO_3 10 g./litre, a c.d. of 30 amp./sq. dm., and an electrode spacing 5 cm. The areas not to be etched are coated with a cold enamel obtained by fixing a layer of shellac to the plate with a tanning solution ($\text{H}_2\text{CrO}_4 + \text{Cr alum}$).

XI.—ANALYSIS

***Quantitative Spectrographic Analysis of Aluminium Alloys with the Application of Objective Photometry.**—I. A. R. Striganov (*Zavod. Lab.*, 1940, 9, (2), 170–178).—[In Russian] S. indicates the optimum working conditions (as

revealed by his own experiments) for the determination of Cu, Mg, Mn, Fe, Si, and Ni in Al alloys. (For abstract of Part II, see *Met. Abs.*, this vol., p. 223.)
—N. A.

Spectral Lines for the Analysis of Aluminium Alloys by the Photographic Method Using the Microphotometer. A. R. Striganov (*Zavod. Lab.*, 1945, 11, (4), 311–314).—[In Russian] Recommends line pairs for use in analysis.
—N. A.

Very Rapid Spectrographic Analysis of Aluminium Alloys. A. R. Striganov (*Zavod. Lab.*, 1945, 11, (5), 415–419).—[In Russian] The method is described.
—N. A.

Very Rapid Spectrographic Analysis of Aluminium Alloys. N. N. Semenov and I. S. Fishman (*Zavod. Lab.*, 1945, 11, (5), 419–424).—[In Russian] A description of the method.—N. A.

Spectrum Analysis of Aluminium Alloys. A. R. Striganov and K. A. Sukhenko (*Izvest. Akad. Nauk S.S.S.R.*, 1945, 9, (6), 593–606).—[In Russian] Reviews methods used for the routine analysis of alloys in many aluminium works, in which approx. 300,000 quantitative determinations per month are being carried out in about 30 factories. For high-speed analysis, spectra taken through a stepped wedge may be estimated by visual interpolation, and tables of the lines used for this are given for Mg, Mn, Cu, Si, Fe, Ti, Co, Sn, Ni, and Be. For more precise work, microphotometric comparison with standard samples is preferred. Standard sparking conditions are given for the Feussner spark generator and for a simpler circuit and the methods of casting standard samples are described. Visual estimation with a polarizing spectrometer may also be used in some cases.—E. VAN S.

The Spectrographic Analysis of Aluminium Alloys. A. R. Striganov and K. A. Sukhenko (*Zavod. Lab.*, 1945, 11, (7/8), 685–694).—[In Russian] A review. (See also abstract above).—N. A.

Preparation and Investigation of Aluminium and Magnesium Alloy Standards for Spectrographic Analysis. K. A. Sukhenko and B. S. Krasil'shikov (*Zavod. Lab.*, 1945, 11, (11/12), 1125–1131).—[In Russian] The question of casting the standards is surveyed.—N. A.

The Influence of Third Elements on the Results of Spectrographic Analysis of Aluminium and Magnesium Alloys. K. A. Sukhenko and O. I. Mladentseva (*Zavod. Lab.*, 1945, 11, (11/12), 1132–1136).—[In Russian] The calibration curves of the standards of a number of alloys were compared among themselves, with the object of rationalizing the standards of alloys similar in chemical composition.—N. A.

The Spectrographic Determination of Copper in Aluminium Alloys. L. E. Vvedensky (*Zavod. Lab.*, 1945, 11, (2/3), 221–222).—[In Russian] Some directions based on V.'s own work.—N. A.

On the Application of the Method of Internal Electrolysis to the Determination of Small Quantities of Copper in Metallic Aluminium. N. A. Suvorovskaya (*Zavod. Lab.*, 1945, 11, (5), 474).—[In Russian] The proposed method is described.—N. A.

***Determination of Iron in Light Alloys by Means of Sulphosalicylic Acid.** E. I. Nikitina (*Zavod. Lab.*, 1940, 9, (5/6), 629–630).—[In Russian] N. recommends the introduction of NH_4Cl to prevent precipitation of $\text{Mg}(\text{OH})_2$.
—N. A.

The Determination of Silicon and Copper in Aluminium Alloys by the Lange Photocolorimeter. A. A. Tikhonova (*Zavod. Lab.*, 1945, 11, (6), 616–617).—[In Russian] Describes the method.—N. A.

***Contribution to Rapid Polarographic Analysis. III.—An Investigation of the Possible Errors in the Estimation of Zinc and Nickel in Aluminium and Secondary Aluminium Alloys.** F. Jablonski and H. Moritz (*Aluminium*, 1944, 26, (5/6), 97–99).—A study was made of the possible errors (mainly due to

high Si content) in the estimation of Zn in Al alloys, using the method previously described. Any residue left after the sample had been dissolved had a negligible volume effect on the determination, and there was no loss of Zn by precipitation as silicate or by adsorption in any undissolved SiO_2 . The solution was stable over a period of days, no precipitation of dissolved SiO_2 taking place. It is concluded that the errors due to analytical procedure are somewhat less than the subjective errors involved in the evaluation of the polarogram.—E. N.

***The Determination of Oxygen in Metallic Antimony.** A. S. Shakhov (*Zavod. Lab.*, 1945, 11, (9), 807–809).—[In Russian] Reduction by means of dry H_2 at 700°C . for $1\frac{1}{2}$ hr. is recommended. The accuracy is to 0.06%.—N. A.

The Quantitative Spectrographic Analysis of Bronze and Cast Iron in the Ultra-Violet Region of the Spectrum. N. N. Semenov (*Zavod. Lab.*, 1945, 11, (2/3), 215–218).—[In Russian] Describes experience in a works laboratory.—N. A.

***Polarographic Determination of Lead in Lead Bronzes.** V. F. Toropova (*Zhur. Priklad. Khim.*, 1945, 18, (3), 177–180).—[In Russian] T. points out the possibility of determining Pb in bronze in alkaline solution. The accuracy is about 2%, and the time required 8 min.—N. A.

Analysis of Cable-Sheathing Alloys. G. M. Hamilton (*Nature*, 1946, 157, (4000), 875).—The use of a new solvent for Pb alloy cable sheathing preparatory to wet analysis is described. The alloy contains approx. Sn 2.0, Sb 0.8, and Cd 0.25% and samples weighing 2.5 g. may be disintegrated in a few minutes in a mixture of 20 c.c. of 30% H_2O_2 and 5 c.c. glacial acetic acid. Sb is not attacked, but is later taken into solution with HCl, which also decomposes most of the peroxide and converts the other metals into chlorides. From this point the analysis follows conventional methods. Alloys rich in Sn are only slowly attacked by the recommended reagent.—G. V. R.

***Determination of Sodium in Calcium-Sodium Babbitts.** V. I. Kolosov and M. D. Trykov (*Zavod. Lab.*, 1940, 9, (5/6), 518–519).—[In Russian] 2–4 g. of Babbitt turnings in a porcelain boat are heated for 10–15 min. at 400° – 450°C . in a Mars furnace in a current of air. The oxidized product is then lixiviated with boiling water and a stream of CO_2 . The residue of CaCO_3 and PbO is filtered off, and Na is determined in the filtrate by titration with HCl.—N. A.

The Application of the Polarographic Method in the Analysis of Bearing Metals. S. I. Kaplansky, A. B. Gurevich, and I. A. Korshunov (*Zavod. Lab.*, 1945, 11, (10), 916–920).—[In Russian] Deals with the construction of directly calibrated apparatus for determining Cu, Cd, and Ni; the analysis of artificial mixtures; the analysis of works samples (for Cu, Ni, and Cd); and the determination of Sb and Sn.—N. A.

Colorimetric Determination of Zinc and Aluminium in Tin and Lead-Tin Solders. S. Yu. Fainberg and T. V. Zaglodina (*Zavod. Lab.*, 1945, 11, (11/12), 1109–1112).—[In Russian] A scheme of analysis is given.—N. A.

The Preparation of Standard Alloys for the Spectrographic Analysis of Zinc, Cadmium, Lead, and Tin. A. K. Rusanov (*Zavod. Lab.*, 1945, 11, (4), 349–350).—[In Russian] In the preparation of alloys of predetermined composition, R. recommends melting in a stream of hydrogen.—N. A.

***Spectrographic Estimation of Traces of Impurities in Zinc Alloys by Means of the Condensed Spark.** Paul Croissant (5me Congrès du Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métalliques, 1946, 65–67; discussion, 68).—Various spark-circuit conditions were tried on cylindrical samples of die-casting alloys to estimate traces of Mg, Fe, and Pb. By the use of unusually large capacity and self-inductance and 2-min. exposures, Pb and Cd can be estimated down to 0.003% and Sn down to 0.001%.

For Pb in the range 0.01–0.03% the coeff. of variation of the duplicate spectra was less than 2.5%. The precision for other elements is under investigation.

—E. VAN S.

The Analysis of Zinc Alloys. R. B. Golubtsova (*Zavod. Lab.*, 1945, 11, (11/12), 1112–1113).—[In Russian] A method is outlined for determining Al (5–20%), Cu (1–5%), and Fe (0.2–1.5%) in Zn alloys.—N. A.

***The Polarographic Determination of Cobalt in the Presence of Nickel. The Catalytic Evolution of Hydrogen in the Presence of Complexes of Cobalt with Dimethylglyoxime.** A. G. Stromberg and A. I. Zelyanskaya (*Zhur. Obsheh. Khim.*, 1945, 15, (4/5), 303–317).—[In Russian] A new method has been worked out for the polarographic determination of Co in the presence of Ni. It depends on the fact that dimethylglyoxime gives an insoluble complex with Ni, while forming a complex compound with Co, which remains in solution and can be determined polarographically. The necessary conditions for the determination of Co in the presence of Ni, Cu, Zn, and Fe have been examined.

—N. A.

***The Fractional Detection of Copper Ions.** N. A. Tananaev and V. N. Podchainova (*Zavod. Lab.*, 1940, 9, (2), 168–169).—[In Russian] CdS can be used for the fractional detection of Cu down to 10 mg./litre in the presence of Bi, Ag, and Hg, and down to 0.5 mg./litre in their absence.—N. A.

The Use of Na_2SO_3 for Volumetric Determination of Lead. T. Ya. Shevtsova (*Trudy Inst. Khim. Khar'kov. Gosudarst. Univ.*, 1940, 5, 283–291; *Khim. Referat. Zhur.*, 1941, 4, (7/8), 69; *C. Abs.*, 1943, 37, 6586).—[In Russian] For analysis of metallic Pb or of Pb alloys, dissolve the sample in HNO_3 (1 : 2), evaporate to dryness, moisten with N-HNO_3 , dissolve in water and filter; neutralize with NH_4OH in the presence of methyl red and titrate with Na_2SO_3 solution to a yellow colour. Cu and Cd interfere with the determination; less than 1% Fe does not interfere. The Na_2SO_3 must contain no sulphates; they can be removed by repeated crystallization from water.

Conductometric Titration of Lead with Iodate. C. Dragulescu and E. Latiu (*Z. anal. Chem.*, 1943, 126, 63–66; *C. Abs.*, 1943, 37, 6585).—The conductometric titration of Pb^{++} with IO_3^- can be carried out satisfactorily according to the reaction $\text{Pb}^{++} + 2\text{IO}_3^- = \text{Pb}(\text{IO}_3)_2$, but it is necessary to wait until the conductivity value is const. after each addition of reagent, particularly in the neighbourhood of the end point. The reason for this is the fact that $\text{Pb}(\text{OH})\text{IO}_3$ is formed at first and this basic iodate changes very slowly into the normal salt.

Potentiometric Titration of Lead with Iodate. C. Dragulescu and E. Latiu (*Z. anal. Chem.*, 1943, 126, 67–72; *C. Abs.*, 1943, 37, 6585).—To the neutral $\text{Pb}(\text{NO}_3)_2$ solution add an excess of KIO_3 solution and titrate the excess with standard $\text{Pb}(\text{NO}_3)_2$ solution. Use Ag wire as indicator electrode against a *N*-calomel electrode in the usual compensation apparatus. The excess KIO_3 can also be titrated with standard AgNO_3 solution. The results obtained by direct titration with KIO_3 were unsatisfactory, although some freshly precipitated AgIO_3 was added at the start and in some cases EtOH was added to decrease the solubility.

***A Simple Colorimetric Method for the Works Determination of Nickel.** J. Kinnunen (*Metall u. Erz.*, 1944, 41, (13/14), 158).—Ni may be estimated colorimetrically as NiSO_4 using a Lange photoelectric colorimeter with a red filter. The light adsorption is independent of time, temp., H_2SO_4 , H_3PO_4 , thiourea, and CuSO_4 concentration. When Cu is present, it may be decolorized with a solution containing 50 g. thiourea, 500 c.c. H_2SO_4 (1 : 1), and 1000 c.c. H_2O , 15 c.c. of which will decolorize 200 mg. Cu. Fe is decolorized with H_3PO_4 . The method may be applied to Ni and Cu electrolytes, NiSO_4 , NiO, nickel matte, nickel silvers, &c. The results obtained are in very good agreement with those from electrolytic estimations.—E. N.

***Direct Titration of Nickel with Dimethylglyoxime Solution.** A. K. Babko and M. V. Korotun (*Zavod. Lab.*, 1945, 11, (10), 896-899).—[In Russian] The chemical conditions of titration were studied and a method established for the determination of Ni in the presence of Fe and Cu.—N. A.

***On the Determination of Very Small Quantities of Platinum.** N. S. Poluektov and F. T. Spivak (*Zavod. Lab.*, 1945, 11, (5), 398-403).—[In Russian] A method has been evolved for the determination of small amounts of Pt in metallic Sb. It is based on the reaction with Sn chloride and ethyl acetate and enables 0.03 g. of Pt in 10 g. of specimen to be determined.—N. A.

***Gravimetric Method for the Determination of Potassium, Rubidium, and Cesium.** O. G. Sheintsis (*Zavod. Lab.*, 1940, 9, (2), 162-163).—[In Russian] K, Rb, and Cs are precipitated as hexanitrodiphenylamines, and the precipitate washed with dry ether and dried at 80°-90° C.—N. A.

***The Use of Anthraquinone- α -Arsonic Acid in the Determination of Small Quantities of Tin.** V. I. Kuznetsov (*Zavod. Lab.*, 1945, 11, (4), 263-266).—[In Russian] Sn is converted into the β -compound of Sn^{IV} by heating the alkaline solution acidified with HNO_3 . For precipitation from very dilute solutions, K. proposes a new reagent—anthraquinone- α -arsonic acid. The method is suitable for use in the presence of large quantities of W, Sb, Cu, Cr, Mo, V, and Nb.—N. A.

***Separation of Iron, Aluminium, and Chromium from Zinc by Means of Pyridine.** E. A. Ostroumov and R. I. Bomshtein (*Zavod. Lab.*, 1945, 11, (2/3), 146-149).—[In Russian] 200 ml. of an HCl solution, containing approx. 0.3 g. of the sesquioxides of the metals, are neutralized with NH_4OH until turbidity appears; this is cleared by addition of 3 or 4 drops of dil. HCl. Then 20-40 g. of NH_4Cl are added and the solution heated to boiling; after removal from the flame, Fe, Al, and Cr are precipitated by means of a 20% solution of pyridine.—N. A.

***The Reaction Range of Organic Reagents in Metal Analyses.** Hellmut Fischer (*Wiss. Veröff. Siemens-Werken*, 1940, 217-229; and *Mikrochem. ver. Mikrochim. Acta*, 1942, 30, 38-56; *Chem. Zentr.*, 1941, 112, (I), 2975; *C. Abs.*, 1943, 37, 6582).—A study was made of the way in which the reactivity of diphenylthiocarbazone (dithizone) varies with different groups of cations with the same electron configuration. The pH range of stability of the metal complexes formed was determined. Corresponding experiments were carried out with diphenylcarbazone dissolved in both CCl_4 and CHCl_3 . The reactions of α -nitroso- β -naphthol and β -nitroso- α -naphthol with heavy metal cations can be carried out in the same manner as the reactions of dithizone. The stability range was determined for these compounds also. A definite relationship was found to exist between the reactivity of the cations and the stability of the complexes formed on the one hand and the electron configuration of the reacting cations on the other. The tendency to form stable complexes is most pronounced in the case of the cations of the transition metals with incomplete electron shells; next to these are the group of cations with 18 external electrons. The reactivity of cations with $18 + 2$ electrons is essentially more limited, as is also the pH range within which they react.

The Sorting of Metals and Alloys by Means of Spot-Tests. E. I. Nikitina (*Zavod. Lab.*, 1945, 11, (2/3), 231-234).—[In Russian] Methods are indicated for Al alloys, Mg alloys, brasses, and steels.—N. A.

Quantitative Analysis by the Variation of the Specific Activity of an Added Radioactive Isotope. Pierre Sue (*Nature*, 1946, 157, (3993), 622).—The method of using radioactive isotopes in quantitative analysis is illustrated by reference to a new method of analysis for K in the presence of Li and Na. To the solution is added a radioactive potassium preparation of known activity. From the new mixture the K is precipitated as KClO_4 . From the measured

activity of a weighed sample of the precipitate, the quantity of inactive K in the original mixture is deduced. The precipitation need not be quantitative.

—G. V. R.

Metallurgical Analysis. Review of the Commonly Used Drying Agents. James Sanders (*Iron and Steel*, 1946, 19, (8), 423–425).—A review of drying agents commonly used in metallurgical analysis, dealing with CaCl_2 , B_2O_3 , conc. H_2SO_4 , P_2O_5 , $\text{Mg}(\text{ClO}_4)_2$, and silica gel. The properties of each substance are discussed and their method of use and rate of desiccation considered. Some examples of their use are also given.—J. W. D.

†**Modern Light Sources for Spectrographic Analysis.** N. S. Sventitsky (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Phys.], 9, (6), 677–690).—[In Russian] A critical review of light sources for spectrographic analysis, comparing the D.C. arc, different types of spark, and A.C. arcs of the usual types described in German and American papers. The A.C. arc with H.-F. ignition, favoured in the U.S.S.R., is supplied with an additional interrupter in the primary of the transformer, which enables an intermittent discharge to be used for samples of low melting point. The trend in the development of controlled sources is discussed and it is pointed out that there is only a limited value to theoretical calculations based on the assumption that thermal equilibrium is attained in the arc column. In seeking reproducibility of spectra, it may not be enough to aim at uniform electrical conditions which would restrict the scope of the light source; a more useful circuit would produce regular variations of potential gradient, so as to ionize the more difficult elements as well as those with lower excitation potentials.—E. VAN S.

The Impulse Source of Light and Its Application to the Spectral Analysis of Metalloids in Solids. I. I. Levintov (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Phys.], 9, (6), 699–702).—[In Russian] To estimate non-metallic elements in refractory solids which may be insulators, such as are used as catalysts, no existing spark- or arc source is suitable. To an ordinary arc circuit (A.C. or D.C.) a secondary circuit is added in which a condenser of 1 μF charged to 6000 V. by a transformer is discharged periodically by a rotary interrupter. The disc of the interrupter also carries an aperture which uncovers the spectrograph slit only at the time of the discharge impulse. This results in an arc of extremely high c.d. and can be used to excite non-metallic samples packed in a carbon crater. The resultant spectra give lines for the halogens and S, but have fewer air lines than a high-intensity spark; highly ionized lines abound. Very sensitive plates must be used as the sample is rapidly volatilized and if the plates are too slow spectra must be superimposed. Very short arc gaps (0.5–1 mm.) are essential, as an ordinary arc is often extinguished by the impulse discharge. An alternative circuit is given which is of simpler type but requires a 300 kW. transformer. Little quantitative work has been done with this light source but traces of halogens in copper oxide can be estimated (above 0.05%) by an internal-standard comparison method.—E. VAN S.

XII.—LABORATORY APPARATUS, INSTRUMENTS, &c.

(See also “Testing” and “Temperature Measurement and Control.”)

The Upright Type of Microscope in Metallography. J. R. Vilella (*Metal Progress*, 1946, 49, (1), 113–116).—The “inverted” microscope which permits the specimen to be placed on the stage with the polished side down, and which can be used in conjunction with a horizontal camera, is generally preferred for metallographic work. The upright microscope, however, has the advantages of superior comfort and adjustability in use, more extensive stage movement, possibility of determining the position on the specimen of the field under

examination, and a permanently attached and aligned source of light. Micrographs produced with both types of instrument illustrate the high order of clarity obtainable with the "inverted" type of instrument and recommendations are made regarding lighting, choice of film and filter, and developers.

—P. R.

A New Type of Focusing X-Ray Monochromator. W. H. Hall (*Nature*, 1946, 157, (3999), 842).—It is pointed out that single-texture copper strip, having a $\langle 100 \rangle$ direction lying in the rolling direction and a $\{100\}$ plane in the strip surface, may be used as the diffracting material in monochromators of the "bent crystal" type. The strip may be bent and ground more easily than the crystals usually employed and has other advantages.—G. V. R.

X-Ray Goniometer for Use with Micro-Specimens. L. V. Al'tshuller (*Zavod. Lab.*, 1941, 10, (3), 271-277).—[In Russian] A. describes a camera by means of which Debye photographs can be obtained from micro-specimens, using a focusing method.—N. A.

On Some Tests of Prism Spectrographs of High Dispersion. A. Gatterer and J. Junkes (*Ricerche Spettroscopiche*, 1940, 1, (3), 73-137).—[In German] A detailed examination of the resolving power and purity of the spectrum given by glass and quartz spectrographs. The instruments are considered in respect of the matching of their optical and photographic resolving power, as affected by focal length and aperture. Defects in the purity of the spectrum such as "ghosts" and broadened lines are discussed and methods of mechanical construction and adjustment are described. The paper is illustrated with actual photographs of spectra and micro-photometric traces of the photographs.—E. VAN S.

Comparative Light-Transmission of the Zeiss Qu24 Spectrograph and the Zélande Spectrograph of Jobin and Yvon. Edmond Lœuille (*5me Congrès du Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métalliques*, 1946, 46-64).—After a detailed comparison of the optical systems of two spectrographs, L. describes the effect on the relative intensity of two spectrum lines which results from changing from one spectrograph to another, a point which is of practical importance in the standardization of methods of analysis. The changed relative intensity is due to reciprocity-law failure of the photographic plate and disappears if the two spectrographs are similar in their light-transmissions.—E. VAN S.

Some Apparatus and Devices in the Spectrographic Laboratory. N. N. Semenov (*Zavod. Lab.*, 1945, 11, (2/3), 218-220).—[In Russian] S. describes the drying of photographic plates in a vacuum, a chill mould for casting specimens, and devices for sharpening electrodes.—N. A.

XIII.—PHYSICAL AND MECHANICAL TESTING, INSPECTION, AND RADIOLOGY

***On Random Sampling in Test Statistics.** Ludwig Küttner (*Metallwirtschaft*, 1944, 23, (22/26), 228-233).—Reports an extensive investigation of the classification, causes, and distribution of errors in experimental observations, the results being shown in tables, graphs, and error curves. K. discusses the choice of sample specimens and the probability that their properties shall be representative of the whole group. The interpretation and conclusions which can be reached from the results of small-scale sampling are dealt with.—E. N.

Practical Hardness Testing. K. Meyer (*Maschinenbau*, 1941, 20, 431-436; *Chem. Zentr.*, 1942, 113, (II), 1285; *C. Abs.*, 1943, 37, 5680).—Static, dynamic, and special methods for hardness testing are discussed and the equipment is described.

***Elastic Constants of Isotropic Media [Steel, Brass].** S. Bhagavantam and B. Ramachandra Rao (*Nature*, 1946, 157, (3993), 624).—A method for obtaining Young's modulus Y and the rigidity modulus n for very small quantities of isotropic materials is briefly referred to. The method depends upon the transmission of ultrasonic waves of continuously varying frequency through the specimen. Typical results for glass, steel, and brass are given and compared with the static values.—G. V. R.

Verification of Tensile-Testing Machines. W. Thompson (*Metallurgia*, 1946, 34, (201), 145-146).—Methods adopted for verifying the accuracy of tensile-testing machines including the calibrated test bar and the steel proving ring are described and discussed and attention is directed to the comparative ease of checking by means of the latter method.—J. W. D.

New Testing Machines of the TsNIITMASH Type. I. V. Kudryavtsev (*Zavod. Lab.*, 1945, 11, (2/3), 209-214).—[In Russian] New machines for the tensile testing of metals are described.—N. A.

Rapid Method for Determining, by the Amsler Machine, the Anti-Friction Properties of Alloys During Sliding Friction. A. I. Chipizhenko (*Zavod. Lab.*, 1945, 11, (7/8), 727-733).—[In Russian] C. proposes a method which gives significant amounts of wear in a comparatively brief time of testing.—N. A.

A Small-Scale Creep-Testing Unit. G. T. Harris (*Metallurgia*, 1946, 34, (201), 129-132).—A small-scale creep-testing machine has been developed to accelerate the development of new heat-resisting alloys for gas turbines. The machine is described, its accuracy defined, and some results are given to show the form of the strain/time curves obtained.—J. W. D.

Improving Design Through Life Testing. R. E. Peterson (*Machine Design*, 1945, 17, (8), 127-130).—"Life" tests are devised to test individual parts under an approximation to working conditions. Specially designed apparatus is illustrated for testing door latches and hinges, the durability of refrigerator trays, and turbine blades. Life tests are considered in two main groups including (a) fatigue and creep, and (b) wear, corrosion, and repeated impact, (a) comprising quantitative, and (b) qualitative tests. The merits and applications of *ad hoc* life tests are compared with those of conventional testing methods.—P. R.

The Supersonic Flaw Detector and Its Application in the Sheet-Metal Industries. A. C. Rankin (*Sheet Metal Ind.*, 1946, 23, (229), 893-898, 901).—A brief account is given of the apparatus and the principles on which it operates. The application of supersonic flaw detection to problems encountered in the sheet-metal industry is discussed, with some examples of its use. R. suggests that standardization and calibration for each application is necessary.—R. Gr.

RADIOLOGY

Prague Conference on the Use of X-Rays in the Metal Industries. V. Vand (*Nature*, 1946, 157, (3987), 415-416).—A summary report of a conference held in Nov. 1945 by the Society of Czechoslovak Mathematicians and Physicists.—G. V. R.

Sectional Radiography. Robert Taylor (*Metals and Alloys*, 1945, 22, (1), 99-101).—A method is outlined by means of which the defects in any given plane through a casting may be recorded sharply on an X-ray film and all defects on other planes thrown out of focus. This is accomplished by moving the X-ray tube and film in opposite directions during the time of the exposure. Methods of securing the desired movement are indicated diagrammatically.—J. C. C.

The Present State of X-Ray Radiology. A. K. Trapeznikov (*Zavod. Lab.*, 1945, 11, (4), 287-304).—[In Russian] A review.—N. A.

Two-Million Volt X-Ray Inspection Equipment. David Goodman (*Metal Progress*, 1946, 49, (2), 327-332).—A high-voltage installation for the non-destructive inspection of heavy ordnance permits the horizontal radiations to be used for inspecting pieces carried round the machine on a turn-table, while a beam, transmitted through the anode along the axis of the tube, is used for examining larger pieces in another part of the building, the transmitted beam being shut off by lead screening when not in use. The lay-out, safety devices, and method of operation are described.—P. R.

X-Ray Tube with Earthed Anode. L. P. Spassky (*Zavod. Lab.*, 1940, 9, (5/6), 655-657).—[In Russian] Describes the design of a new type of anode for X-ray tubes intended for radiology.—N. A.

XIV.—TEMPERATURE MEASUREMENT AND CONTROL

Anticipator Improves Temperature Control. M. J. Manjoine (*Machine Design*, 1945, 17, (6), 116).—The "anticipator" consists of two thermocouples of different characteristics and an electrical heating element, the assembly being enclosed in a vacuum tube. The thermocouples and the control thermocouple in the furnace are connected in series in such a way that the polarity of the couple with the less thermal capacity is additive, and that of the couple with greater thermal capacity is subtractive, with respect to the furnace couple. The heating element of the anticipator is actuated by the power source connecting the control mechanism with relays operating the main power supply. Differences in temp. between the two anticipator couples cause changes in the controller voltage and prevent the furnace temp. from increasing above or decreasing below the prescribed limits. It is stated that control of heat-treating furnaces within $\pm 2\frac{1}{2}^{\circ}\text{F.}$ ($\pm 1\frac{1}{2}^{\circ}\text{C.}$) is practicable with anticipators of this type.—P. R.

XV.—FOUNDRY PRACTICE AND APPLIANCES

German Aluminium Alloy Pistons. — (*Found. Trade J.*, 1946, 78, (1548), 414, 417).—A summary of British Intelligence Objectives Sub-Committee Report No. BIOS 238 (1946).

***Reciprocal Effects of Cuprous Oxide, Poling Gases (Hydrogen), and Deoxidizers in the Melting and Casting of Oxygen-Free and Oxygen-Containing Copper.** Paul Klare and Justus Kohlmeyer (*Metall u. Erz*, 1944, 41, (13/14), 149-155; (15/16), 174-180).—The production of high-conductivity copper was studied and the optimum conditions for poling, deoxidation, and casting ascertained. The action of the undermentioned deoxidizers on underpoled and overpoled copper was investigated: phosphorus, lithium, magnesium, calcium, strontium, barium, cerium, aluminium, and thorium. Copper of the highest conductivity is obtained by melting in a reverberatory furnace, poling at a temp. not exceeding 1200°C. until the oxygen content is reduced to 0.01%, and pouring the metal into ladles of 1800 kg. capacity which contain about 4 shovelfuls of glowing charcoal. The oxygen (0.002-0.005%) and hydrogen contents are thereby rendered so low that an addition of 0.0015% lithium effects the simultaneous removal of both elements. The molten metal, now at a temp. not exceeding 1160°C. , is poured into billets under a protective atmosphere of carbon monoxide.—E. N.

***Control of Shrinkage Porosity and Mechanical Properties in Chill-Cast Phosphor-Bronzes, Lead Bronzes, and Gun-Metals.**—I-II. (Pell-Walpole). See p. 317.

Relining Lead Alloy Journal Bearings. E. A. Wolfenden (*Metals and Alloys*, 1945, 22, (1), 81–84).—A detailed account of good practice. A lead–tin alloy containing 20–25% tin applied at 625°–675° F. (330°–357° C.) is recommended for tinning and the lining metal (tin 3–5, antimony >8%, lead balance) should be poured at 625° F. (330° C.).—J. C. C.

Technical Control in a Magnesium Foundry. G. B. Partridge (*Found. Trade J.*, 1946, 79, (1550), 5–8).—Read before the Birmingham, Coventry, and West Midlands Branch of the Institute of British Foundrymen. A general review dealing with melting technique, sand control, heat-treatment, and inspection.—J. E. G.

Magnesium Die-Castings. Ralph M. Heintz (*Machine Design*, 1946, 18, (2), 101–104, 180; (3), 111–115).—The problems of the design of magnesium die-castings differ entirely from those in sand casting. This is shown by tables comparing tolerances, limits of draught, shrinkage, and minimum hole sizes and wall thicknesses for the two processes. Die-cast parts are classed as “decorative” (appearance being the chief consideration); “functional” (for working parts, in which strength is less important than lightness and cheapness); “structural” (in which moderate stresses are involved); and “stressed” (in which working stresses are high). The composition and physical and mechanical properties of a common die-casting alloy are tabulated, with references to American specifications. The economy in machining effected by using die-cast parts is emphasized. Practical recommendations are made for dimensions of holes and fillets, uniformity of section, and protective treatment. H. suggests that the use of die-castings could be considerably extended if modifications were made in production practice.—P. R.

Mass-Production Precision Casting [of Stellite]. — (*Machinery (Lond.)*, 1946, 69, (1762), 65–73).—An illustrated account is given of the equipment of the Haynes Stellite Co. for producing precision castings by the lost-wax process. Castings weighing from 0.002 to 5 lb. each have been regularly produced to dimensional tolerances of ± 0.003 in.—J. C. C.

Lateral Die Movement with Positive Lock. N. Field (*Machinery (Lond.)*, 1946, 69, (1763), 118–119).—Details are given of the design of a die block in which the whole of the casting is formed within laterally sliding members.

—J. C. C.

Castings and Weldings. (Sir) Claude D. Gibb (*Found. Trade J.*, 1946, 79, (1557), 179–184, 191).—Edward Williams lecture to the Institute of British Foundrymen.—J. E. G.

Casting Reflections. D. Howard Wood (*Found. Trade J.*, 1946, 79, (1558) 219–222).—Presidential address to the Institute of British Foundrymen.

—J. E. G.

Ingot-Casting Machines. Karl Siedentopf (*Aluminium*, 1944, 26, (5/6), 88–93).—Typical automatic casting machines are described and their technical and economical requirements are discussed.—E. N.

On the Development of Water-Cooled Ingot Moulds. A. J. Stelljes (*Aluminium*, 1944, 26, (5/6), 94–95).—The development of water-cooled moulds for casting light metal ingots and their advantages over automatic casting machines are discussed. Details are given of a horizontal mould having a D section.—E. N.

Specification-Purchasing of Die-Castings. Herbert Chase (*Metals and Alloys*, 1945, 22, (1), 76–80).—Reproduces and discusses examples of American purchase specifications, some requiring radiographic and breakdown tests and others specifying the type of machine to be employed.—J. C. C.

Closer Co-operation Between the Foundry and Pattern Shop. (1) W. C. Meredith. (2) H. S. W. Brittain (*Found. Trade J.*, 1946, 79, (1550), 3–4; (1552), 49–52).—Two papers contributed to a discussion on the subject and read before the Sheffield Branch of the Institute of British Foundrymen.

Examples are given of typically bad pattern-making. The introduction is suggested of a "methods department" to contact the other departments and correlate information on castings for ultimate transmission to the pattern shop.

—J. E. G.

Core-Making Operations. J. H. List (*Iron and Steel*, 1945, 19, (8), 431-432).—Moulding and core operations used in the manufacture of castings for pump impellers are discussed, with special reference to the sand mixture used and the moulding technique employed for the production of good cores.

—J. W. D.

Determination of "Gas-Content" of Sand Cores. — (*Found. Trade J.*, 1946, 79, (1560), 273-275).—Report of Sub-Committee T.S. 13 of the Technical Council of the Institute of British Foundrymen. Detailed instructions are given for determining the "gas-content" of sand cores.—J. E. G.

Elevating Trucks in the Foundry. R. S. Turner (*Found. Trade J.*, 1946, 79, (1558), 225-227).—A general illustrated account.—J. E. G.

Apprentice Training in the Foundry. John B. Longmuir (*Found. Trade J.*, 1946, 79, (1552), 55-59).—Read before the Birmingham, Coventry, and West Midlands Branch of the Institute of British Foundrymen. A detailed account of a scheme used in a Works Foundry Training Centre.—J. E. G.

Opportunities for the Foundry Technician in India. J. Blakiston (*Found. Trade J.*, 1946, 78, (1547), 393-397).—Read before the London Branch of the Institute of British Foundrymen. A review of recent metallurgical developments in India and future prospects for United Kingdom technicians.

—J. E. G.

XVI.—SECONDARY METALS: SCRAP, RESIDUES, &c.

Use and Working of Light Metal Scrap. G. Gauthier (*Usine*, 1942, 51, (5), 31-33; (6), 37-39; (7), 35-39; (8), 37-39; *Chem. Zentr.*, 1942, 113, (II), 453; *C. Abs.*, 1943, 37, 6612).—Reviews the origin and types of scrap, refining and purifying methods, melting furnaces, fluxes, and further uses of residues from melting.

XVII.—FURNACES, FUELS, AND REFRACTORY MATERIALS

The Ellipsoid Melting Furnace. A. J. G. Smith (*Metallurgia*, 1946, 34, (201), 153-154).—Briefly describes a modification of the horizontal-roller- or rocking-type furnace which has been used on the European continent in recent years, and gives particulars of some typical performances in melting red-brass swarf, nickel-silver grinding swarf, and Argentan alloy.—J. W. D.

The Use of [Iron-Chromium-Aluminium] Alloy No. 2 for the Heating Elements of High-Temperature Furnaces. I. I. Kornilov and N. V. Nikol'sky (*Izvest. Akad. Nauk S.S.S.R.*, 1945, [Tekhn.], (9), 825-830).—[In Russian] An alloy of iron with chromium 23-27 and aluminium 4-6% is recommended for use in furnaces operating at 1200°-1270° C.—N. A.

XVIII.—HEAT-TREATMENT

***Blister Formation on Rolled [Aluminium Alloy] Plate.** Heinz Borchers and Othmar Castellani (*Metall u. Erz*, 1942, 39, 407-409; *C. Abs.*, 1943, 37, 6230).—The alloy studied had the composition aluminium plus copper 2.60, magnesium 1.80, manganese 0.65, silicon 0.29, iron 0.40, and titanium 0.17%.

Slight changes in the time and temperature of heat-treatment often have great effect on blister formation.

Improved Treatments of Aluminium Alloys. Max E. Tatman (*Metal Progress*, 1946, 49, (2), 323-327).—Recent developments in heat-treatment are reviewed with special reference to alloy 24S (sheet, strip, and plate). Warping during a quench has been reduced by (1) air-cooling after removal from the furnace, (2) "fog-" or "spray-" quenching, (3) quenching the part in an upright position, and (4) stretching and/or rolling the quenched material. Methods (1) and (2) are found, however, to decrease strength and corrosion-resistance. A practice widely adopted in aircraft production is to roll annealed hydropress blanks into coils of sheet, which are then solution-treated and quenched with little or no distortion. The purpose of solution-treatment is explained and recommendations are made regarding duration, temp., and rate of cooling. The mechanical properties of aluminium-clad 24S-O are tabulated for various degrees of hardening, the alterations produced by a stated solution-treatment being shown in each case, and similar data for solution-treatment only are given for three other commercial "clad" alloys.
—P. R.

Bright Annealing of Brass. Ernest F. Elliott (*Metal Progress*, 1946, 49, (2), 346).—A brief illustrated account of a bottom-electrode salt-bath furnace for the full annealing of brass cartridges. The material enters at one end, is conveyed through a long ceramic pot containing commercial nitrates, enters the washer, and returns to the starting point for unloading.—P. R.

Atmospheres for Annealing Metals. C. E. Peck (*Metals and Alloys*, 1945, 22, (1), 85-91).—Describes the production, characteristics, and applications of various atmospheres produced by American proprietary fuel-oil and ammonia burners.—J. C. C.

XIX.—WORKING

The Rolling of Metals.—VIII-XI. L. R. Underwood (*Sheet Metal Ind.*, 1946, 23, (227), 475-489, 494; (228), 677-684, 696; (229), 883-892; (230), 1097-1106; (231), 1291-1306).—[VIII.]—Nadai's solution of Von Karman's equation is discussed and the effect of the application of front and back tension on roll pressure and neutral plane are considered. Orowan's theory of rolling is discussed in some detail. This theory enables the pressure distribution between the rolls and the material to be obtained by a graphical method of integration of the equation of friction hill for the case of variable coeff. of friction and/or variable constrained yield stress. In the theory the assumptions of plane sections remaining plane and of slipping friction are abandoned. [IX.]—Quantitative use is made of the various theories and the means for allowing for the effect of work-hardening, front and back strip tension, and roll flattening are considered. [X.]—The calculation of specific roll pressure with flattening and the derivation of Hitchcock's expression for the length of the flattened arc of contact is considered. [XI.]—The general nature of lateral spread is indicated and the chief experimental work on the subject reviewed to show the influence on spread of such factors as width/thickness ratio, roll diameter, coeff. of friction between the rolls and the material, &c. The principal formulæ that have been proposed to calculate spread are discussed and the special case of wire flattening is considered briefly.—R. GR.

*The Relationship Between the Working Properties and Final Rolling Reduction Before Annealing of Sheets of Duralumin-Type Alloys Containing 3% and 4% of Copper, Made from Chill- and Continuously Cast Billets. (Hartmann). See p. 316.

Magnesium Practice at I.G. Farben. — (*Iron Age*, 1946, 157, (26), 68). — Abstract of a report of the Combined Intelligence Objectives Sub-Committee on German magnesium production and fabrication.—J. H. W.

A Reliable Method for Estimating the Technological Plasticity [of Magnesium Alloys]. E. M. Savitsky (*Zavod. Lab.*, 1945, 11, (10), 978-984).—[In Russian] It is shown that for magnesium and its alloys the magnitude of the greatest possible reduction in height during compression is a reliable guide to the capacity of the material for working under pressure.—N. A.

***The Spinning of Zinc Ribbon from Ingot Zinc for Use in Extractor Boxes in the Reduction Works at Government Gold Mining Areas, Limited.** J. C. Phillips (*J. Chem. Met. Min. Soc. S. Africa*, 1945, 46, (3/4), 119-123; discussion, 123-124; also (abridged) *Metal Ind.*, 1946, 68, (20), 392).—Before the war the zinc shavings used as a precipitant for gold were made from imported sheet zinc. P. discusses the difficulties encountered and the apparatus finally evolved for producing zinc ribbon by an alternative method. Molten zinc was poured on to a water-cooled rotating mandril and strips of zinc ribbon approx. 0.002-0.004 in. thick were produced.—H. J. A.

Working of Zinc and Zinc Alloys. O. H. C. Messner (*Schweiz. Arch. angew. Wiss. Techn.*, 1944, 10, (12), 387-388).—Technological aspects of working of zinc and zinc alloys are discussed.—V. K.

***The Effect of Die Friction on the Temperature Distribution in the Drawing Material and on the Heating of the Die.** Erich Siebel and Robert Kobitzsch (*Mitt. K.-W. Inst. Eisenforsch.*, 1943, 26, (7), 91-95).—Thermodynamic considerations of the flow of heat generated by the deformation and by the friction on the surfaces of the drawn material show that its surface temperature is proportional to the coeff. of friction, resistance to deformation, and the square root of the product of speed of drawing and die length. At high speeds of drawing most of the heat generated remains in the drawn material.—V. K.

"Reactive" Wire Drawing. John E. Parton (*Wire Ind.*, 1946, 13, (148), 210).—A letter, suggesting that the term "reactive wire drawing" should be restricted to the process in which the back tension is a fixed proportion of the forward tension and is applied by means of differential gearing. For processes in which the back tension is independent of the forward tension, "pre-stressed wire drawing" or "drawing pre-stressed wire" are considered to be more suitable descriptions.—J. C. C.

Wire-Drawing Machines—Past, Present, and Future. N. Davidson (*Wire Ind.*, 1946, 13, (148), 211-214).—An address to the staff school of Frederick Smith and Co., Ltd., briefly reviewing progress in machine design since 1930 and referring to the use of individual motors for each block and to Vaughan D.C. and Morgan Connor A.C. machines. There is a future for special machines designed for specific products. Mention is made of the possibility of developing extrusion or swaging as methods for wire production.—J. C. C.

German Practice in Extruding Zinc Alloys. Sam Tour (*Iron Age*, 1946, 157, (26), 54).—Abstract of a Joint Intelligence Objectives Agency report on the German war-time use of extruded bars, shapes, and tubes of high-stress zinc-base alloys as substitutes for free-cutting brass.—J. H. W.

Production Processes: Their Influence on Design. VII.—Deep Drawing. Roger W. Bolz (*Machine Design*, 1946, 18, (1), 119-126).—Practical limits for reduction are indicated for various types of die, work, and material. In many instances tapering, necking, upsetting, &c., can be combined to obviate expensive machining operations. The rubber-die-forming process, which is described, often permits the blanking and forming of parts in one operation; it is especially successful with light alloys, copper, and light-gauge steel sheet. The stretch press is often used for parts with fairly symmetrical contours and without considerable reverse bends, but the marginal allowances are so great that scrap may reach 33%. B. states that (1) the high-speed mechanized

press is best suited to the production of large numbers of drawn parts having plain shapes requiring a single operation, (2) round cross-sections are the most easily produced, (3) the drawing of flanged parts results in increased wear on dies: the radius of the drawing edge should be carefully controlled and recommendations are made on the relation between radius and gauge in stainless steel, magnesium, and aluminium alloys, nickel, Monel metal, and Inconel, (4) corner radii should be as great as possible (at least 5 times the gauge) and especially generous allowances should be made in rubber-die forming, (5) the depth of hollows and height of projections should not exceed twice the gauge, (6) drawn parts should be flat-bottomed, and (7) in deep-drawing irregular parts the tendency to form bulges or wrinkles should be counteracted by making depressions, &c., before drawing. Desirable ductilities for 19-gauge soft sheet are given for a number of metals and alloys and recommendations are made regarding the framing of specifications, the deep drawing of magnesium, drawing speeds for different materials, and tolerances.—P. R.

Production Processes: Their Influence on Design. VIII.—**Swiss Automatic Machining.** Roger W. Bolz (*Machine Design*, 1946, 18, (2), 117–119).—The high-precision automatic screw machine developed for use in the clock and watch industries is now widely used to produce small parts in which high precision is required. The machine and available capacity ranges are described and the speed and accuracy of Swiss and American machines compared. Suitable attachments permit centering, boring, reaming, tapping, slotting, and light milling.—P. R.

The Manufacture of [Steel-Backed Sintered Lead-Bronze] Wrapped Bushes. — (*Machinery (Lond.)*, 1946, 68, (1756), 713–718).—Describes press operations in the forming of a standard range of wrapped bronze bushes from strip steel, coppered on one side and coated on the other with lead-bronze alloy (applied in powder form and bonded by sintering).—J. C. C.

Brazed Stampings Reduce Engine Weight. Paul Klotsch (*Machine Design*, 1946, 18, (2), 123–127).—An illustrated description of a 26-h.p. automobile engine, which weighs 58 lb. and is constructed largely from sheet-metal stampings. The sheet-steel cylinder-block assembly is brazed with copper, which is applied to the joints as sheet, wire, or paste. The crankshaft pulley, fan assembly, and fan pulley are also copper brazed. The crank-case is an aluminium alloy casting and the pistons are of heat-treated aluminium alloy.
—P. R.

Combination Method of Forging Magnesium [Alloys]. A. L. Rustay and F. B. Roto (*Machinery (Lond.)*, 1946, 69, (1764), 129–136).—All standard magnesium alloys can be finish-forged under a hammer, provided that the initial metal movement is accomplished by blocking in a slow-moving hydraulic press. By this procedure, the need for using very large hydraulic finishing presses is obviated. Forging should be carried out at 700°–725° F. (371°–385° C.) and the blocker dies should be maintained at 600° F. (315° C.). Reheating, when necessary, should be at 600°–650° F. (315°–343° C.). Some typical forgings are illustrated and reference is made to the use of special coating treatments for protecting magnesium alloys.—J. C. C.

Recent Research in Metal Machining. D. F. Galloway (*Proc. Inst. Mech. Eng.*, 1945, 153, 113–127; discussion, 128–132).—In reviewing recent research on metal machining, G. discusses chip formation under three headings: (1) continuous chips formed *without* built-up edge, (2) continuous chips formed *with* built-up edge, (3) discontinuous chips. Chip formation is shown to be a process of deformation and ultimate failure by shear, and an approximate quantitative analysis of the process is given. The action of cutting fluids and the results of practical research in rough machining, finish-machining, finish-turning, and boring operations are discussed.—H. J. A.

Oxidation by Friction of Machined Metals. K. Lipacher (*Anz. Maschinenwesen*, 1941, 63, (71), 21-22; *Chem. Zentr.*, 1942, 113, (I), 415; *C. Abs.*, 1943, 37, 6226).—No pairs of metals are yet known which do not undergo oxidation during friction, but some react better than others, e.g., brass/brass, brass/steel, and brass/chromium. Rust-free steels always give very poor results. If it is impossible to eliminate unfavourable combinations, it is recommended that a galvanic coating be applied. The use of lubricants such as graphite and high-molecular oils is also helpful. Other means are discussed.

Machining Powder-Metal Parts. Harry L. Strauss, Jr. (*Machinist (Eur. Edn.)*, 1946, 90, (13), 539).—The tool materials recommended and the coolants required for boring, drilling, reaming, and milling powder-metal parts are given. Variations in structure of these parts may be met by: (1) spots of different density, (2) porosity and voids, and (3) poor core structure. Carbide is recommended for machining copper-nickel and copper-chromium alloys, and high-speed steel for brass and copper.—J. H. W.

XX.—CLEANING AND FINISHING

Soft-Water Rinse Improves Aluminium Finishes. Robert S. Herwig (*Iron Age*, 1946, 158, (2), 58-59).—Increased production and fewer rejects result from the use of rinse water of zero hardness in the electroplating or anodizing of aluminium products.—J. H. W.

Cleaning and Sterilization of Receptacles and Utensils Used in Dairy Work. Georges Ray (*Chim. et Ind.*, 1946, 55, (1), 23-30).—The respective advantages and disadvantages of steel and aluminium alloys as used in the dairy industry are discussed.—H. J. A.

The Influence of Manufacturing Defects Due to Melting Technique on the Ability of Some Alloys to be Polished. Edmund R. Thews (*Oberflächentechn.*, 1942, 19, 3-4; *Chem. Zentr.*, 1942, 113, (I), 2192; *C. Abs.*, 1943, 37, 3384).—Iron-containing bronzes and special brasses sometimes show hard spots which cannot be polished with the usual agents. Addition of aluminium prevents the formation of such spots, which are caused by iron carbides. The lack of ability of copper alloys to take a good polish is often due to insufficient attention to the properties of the alloying materials with regard to melting technique.

***The Electrochemical Polishing of Metals. I.—The Electrochemical Polishing of Nickel.** V. I. Lainer (*Zhur. Priklad. Khim.*, 1945, 18, (4/5), 236-246).—[In Russian] Nickel can be successfully polished electrochemically in H_2SO_4 solution in a very short time; the optimum concentration of acid is 68.7% (sp. gr. 1.6) and the temp. 40° C.—N. A.

Ultra-Fine Surfaces on Metals. Kenneth Rose (*Metals and Alloys*, 1945, 22, (1), 70-75).—A general review of methods of producing fine surface finishes and measuring surface roughness.—J. C. C.

Metal Finishing. VIII.—Paints, Varnishes, and Enamels. H. Silman (*Sheet Metal Ind.*, 1946, 23, (229), 949-953, 956).—The reasons for the use of pigments in paints, lacquers, and enamels are explained. Of the white pigments, the chief non-reactive metallic are antimony trioxide, titanium oxide, and lithopone, with white lead (basic lead carbonate essentially) and zinc oxide as the leading reactive whites. Methods of producing these materials and their properties in relation to paints are described.—R. GR.

XXI.—JOINING

Bonding of Rubber to Metal. G. Proske (*Gummi-Zeit.*, 1942, 56, 183-184, 195-196, 207-209; *C. Abs.*, 1943, 37, 5619).—A general review, with quantita-

tive data and recipes, of various processes for bonding natural and synthetic rubber mixtures to metals.

A New Method for Obtaining More Secure Adhesion Between Rubber and Aluminium. G. Génin (*Ind. Chim.*, 1941, 28, 266-267).—A discussion of the Yates process.

"Araldit"—A New Plastic Material for Joining Light Metals. E. Preiswerk and A. von Zeerleder (*Schweiz. Arch. angew. Wiss. Techn.*, 1946, 12, (4), 113-119).—Describes a new thermoplastic material "Araldit-Ciba," developed for joining light-metal parts such as sheets, tubes, &c. The pieces to be united are cleaned of dirt and grease and then held together to prevent movement. The powdered plastic is spread around the joint, which is then warmed to 120°-220° C. for 5-120 min., when the powder melts and fills the joint, final hardening taking place on cooling below 100° C. An intimate bond is formed without the application of pressure. Properties of pure Araldit are: tensile strength 7-8 kg./sq. mm., bending strength 1200-1300 kg./sq. cm., impact strength 13-14 kg./sq. cm., and elastic modulus 30,000-31,000 kg./sq. cm. The strength of joints made between light metals, and measured in terms of shear strength, is 2.3 kg./sq. mm. from -58° to 80° C., decreasing slightly between 80° and 100° C. and rapidly to 0.7 kg./sq. mm. at 120° C., after which the Araldit again becomes liquid. Joints subjected to alternating-bend tests at a load of 11 kg./sq. mm. have a life of more than 20 million reversals; failure, when it occurs, usually takes place outside the bond. The plastic is resistant to water, aliphatic and aromatic compounds, and anodic oxidation, if not heated above its softening temp., 80° C.; it is tasteless, odourless, and non-poisonous, and can therefore be used in the construction of food containers. Flanged and butt joints between light metals, light metals and aluminium bronze, and light metals and steel are illustrated.—E. N.

Soft Solders and Soldering Technique. L. G. Earle (*Sheet Metal Ind.*, 1946, 23, (227), 515-521; (228), 719-724; (229), 945-948, 954).—The history and properties of solders are described, the fundamentals of joint formation and solder penetration discussed, and the practical application of these principles is considered. Advice is given on the technique of soldering various types of articles.—R. Gr.

***A Study of the Behaviour of Ruthenio-Palladium in Torch Flames, with the Object of Improving Soldering Technique.** R. H. Atkinson and G. P. Gladis (*Metals Technol.*, 1946, 13, (3); and *A.I.M.M.E. Tech. Publ. No. 1982*, 9 pp.).—Palladium-ruthenium alloys used for jewellery were found to undergo surface damage in areas which had been repeatedly heated when ornaments were built up by soldering. This was regarded as due to heating in atmospheres which were alternately oxidizing and reducing. Laboratory samples of ruthenio-palladium (4.5% ruthenium) were repeatedly heated and cooled using (a) oxy-acetylene, (b) air-acetylene, (c) oxy-city gas, and (d) oxy-hydrogen flames, in reducing, oxidizing, and neutral conditions. The samples were then examined microscopically and submitted to bend tests. The results are tabulated and photomicrographs are reproduced. The oxy-acetylene flame produced the least damage and the air-acetylene flame was the next most satisfactory. In all cases reducing flames caused most damage. The oxy-acetylene flame can also be used without fear on the normal platinum solders.—W. H. R.

Silver Simplifies Shunt Soldering. U. L. Smith and Paul MacGahan (*Machinist (Eur. Edn.)*, 1946, 90, (9), 339).—The soft soldering of ammeter shunts was often imperfect. Silver soldering, although producing a much better joint, was too slow for volume production until high-frequency induction heating was used. This method of silver soldering is briefly described.—J. H. W.

Silver Alloys Braze Auto Parts. Adolph Bregman (*Machinist (Eur. Edn.)*, 1946, 90, (6), 208-209).—The use of low-m.-p. alloys, improved methods of application, and better heating methods and equipment increased the application of silver alloys for brazing automobile parts. The particular applications of these alloys are tabulated.—J. H. W.

Electric-Furnace Brazing. H. D. Hendrick (*Machinery (Lond.)*, 1946, 68, (1752), 585-591).—Furnace brazing with copper, brass, and silver solders is described. Furnaces, atmospheres, brazing alloys, fluxes, and questions of joint design are briefly considered.—J. C. C.

***On a New Type of Defect in Light-Metal Welds and Its Elimination.** H. Mäder (*Aluminium*, 1944, 26, (5/6), 83-84).—Blisters in the vicinity of welds in aluminium-magnesium-manganese alloys are attributed to the trapping of evolved gas. They can be eliminated by the addition of further constituents to the alloy, which increase its freezing range, thereby allowing the gases to escape before complete solidification. The properties of the new alloy, "Berghaus-Aluminium, Al-Mg 75," were studied and compared with those of the alloy Al-Mg 7. After normal solution-treatment at 580° C. for 30 min., the new alloy has a layer-corrosion-resistance equal to, and a stress-corrosion-resistance superior to, that of Al-Mg 7.—E. N.

The Practical Significance of Recent Work on the Pressure Welding of Light Alloys. R. F. Tylecote (*Sheet Metal Ind.*, 1946, 23, (230), 1165-1167, 1178).—In a short historical survey of the subject it is stated that pressure welding of aluminium by hammering while hot was used extensively in Germany over 40 years ago. The principles and technique of the modern method of pressing together the cold metals are described. Pickling or, preferably, scratch brushing the surfaces is necessary prior to welding, which is accomplished by a minimum critical deformation to break up the thin oxide film followed by recrystallization. Applications of the method are cited.—R. Gr.

The Welding of Non-Ferrous Metals.—VII. E. G. West (*Sheet Metal Ind.*, 1946, 23, (227), 553-564; (228), 755-761; (229), 961-970; (230), 1168-1177; (231), 1375-1382).—W. considers the resistance welding of aluminium alloys in a series of sections on the specific requirements of aluminium and its alloys for satisfactory welding, machine data, properties of spot welds, seam welding, and resistance butt welding. The effect of aluminium oxide films as an advantage owing to their high electrical resistance, and disadvantages in other directions, are indicated. Aluminium and its alloys require up to eight or nine times the power required for steel spot welds and the requirements of machines for this purpose are high currents, accurately controlled, flowing for short times, with exact synchronization of the required pressure. Information is given about the preparation of the metal surfaces, machine setting, electrodes, and control-test methods. The results of metallurgical examination of spot welds by various authorities are described, and it is stated that the resistance to corrosion is reduced by spot welding. The possibilities of seam welding are discussed, and it is stated that the advantage of high speed for thicknesses up to 0.1 in. almost outweigh the plant cost. Alloys are easier to weld than aluminium. The technique of seam welding and the properties of the welds are described. Experiments on pressure welding, and the trends in flash and butt welding are referred to briefly. Aluminium and several of its alloys are now brazed on a production basis, although only on a small scale in this country. The subject is dealt with in the sections on: fundamentals of aluminium brazing, fluxing, filler metal, methods used—i.e. flame, furnace, dip, electric—properties of brazed joints, and applications. The welding of magnesium alloys requires particular types of joint, care in fluxing, and cleaning of the metal surfaces. The properties of the alloys are described. The welding processes that may be used are chiefly the oxy-gas methods, while the heliarc or argon-arc methods are being developed commercially; resist-

ance welding is confined to spot welding up to thicknesses of about 12 S.W.G.; pressure welding is also possible. These methods are described and details of technique are explained.—R. Gr.

How to Weld Aluminium. — (*Iron Age*, 1946, 157, (25), 50-55; (26), 60-64; 158, (1), 67-71; (2), 53-57).—A comprehensive review of modern techniques in welding aluminium, including gas, carbon-arc, metallic-arc, atomic-hydrogen, inert-gas-shielded-arc, and resistance welding, brazing, and soldering. The weldability of various aluminium alloys, electrode size and type, joint edge preparation, pre-weld cleaning, and the final inspection of joints are discussed.—J. H. W.

***Spot Welding of Light Metals in the Manufacture of Aircraft.** W. Heiz (*Schweiz. Arch. angew. Wiss. Techn.*, 1944, 10, (8), 241-252).—Technical and economic factors controlling the production of spot-welded joints are reviewed. The effects of the spot diameter and distances between spot joints on the strength of the joints were studied and the relationship between these quantities is graphically presented. It is shown that the strength of a spot joint increases with the spot diameter and that an optimum distance between spots gives highest strength values. Macrostructures illustrating various types of spot-weld defects are given and various methods of revealing and controlling these defects are described.—V. K.

***Welded Magnesium Alloys and Their Alternating Tension-Compression Fatigue Strength.** (Thum and v. Manteuffel). See p. 317.

Cleaning of Magnesium Alloy Sheet for Spot Welding. K. E. Dorcas and N. H. Simpson (*Aero Digest*, 1945, 48, 107, 112).

Welding Small Platinum Heaters and Electrodes. A. R. Morris (*J. Sci. Instruments*, 1946, 23, (4), 84-85).—A note describing a device for welding thin platinum wires. The wires are accurately located on a base-plate and are heated by an oxy-coal-gas flame. The weld is accomplished by hammering a small punch held accurately above the wires.—W. H.-R.

Welding with Powdered Metal. — (*Iron Age*, 1946, 157, (25), 69).—A new welding technique involves the use of a spray of powder metal instead of the usual welding rod. The equipment consists virtually of an ordinary welding torch, but with a special nozzle and sixteen independent controls, many of them in the handle of the torch. The process is somewhat restricted at present by the number of powder metals and fluxes available, but interesting results have been obtained in welding, brazing, and hard and plastic surfacing with lead, tin, zinc, cadmium, aluminium, copper, iron, magnesium, manganese, nickel, silver, and alloys of these containing beryllium, bismuth, carbon, cobalt, molybdenum, and tungsten.—J. H. W.

Joining of Zinc and Zinc Alloys. O. H. C. Messner (*Schweiz. Arch. angew. Wiss. Techn.*, 1944, 10, (11), 356-358).—Various joining methods, viz. riveting, soldering, and welding of zinc and zinc alloys are briefly reviewed.—V. K.

The Repair of Casting Defects by the Slavyanov (Arc) Method of Welding. I. F. Mart'ynenko (*Sudostroenie*, 1945, (2), 19-21).—[In Russian] Problems arising in the welding of copper-base casting alloys are reviewed. Satisfactory results in welding lead-free alloys were obtained with electrodes containing: aluminium 7-8, manganese 1.5-2, iron 0.6-1, zinc 3-4%, and copper remainder. The electrodes were coated with fluxes containing: ferrosilicon 60, manganese ore 40, and sand 10 parts by weight. The weld metal obtained possesses mechanical properties equal to those of the parent metal.

—V. K.

Improvements in Manual Arc Welding. W. J. Conley (*Metal Progress*, 1946, 49, (1), 103).—Improved practice has permitted the use of larger electrodes and heavier currents, with the economy in time and material which is shown for T joints and butt joints. It requires a larger capacity in welding machines, but results in easier welding.—P. R.

Measurement and Effect of Contact-Resistance in Spot Welding. Robert A. Wyant (*Elect. Eng.*, 1946, 65, (1), 26-33).—Static and dynamic methods of measuring contact-resistance are described, with special reference to the spot welding of aluminium alloys.—E. V. W.

Application of Quality Control to Resistance Welding. L. S. Hobson, R. S. Inglis, and R. P. McCants (*Elect. Eng.*, 1945, 64, (8), 573-575).—Standard samples of material identical with production parts are welded under production conditions and then tested to destruction in a torsion device. The diameter, torque, and angle at failure are measured and combined into a single figure indicative of weld quality. This figure is plotted on control charts.

—E. V. W.

Projection Welding for Short Runs. R. O. Klenze (*Machine Design*, 1945, 17, (11), 101-106).—Pressure-type resistance welding machines can be economically adapted for production of limited numbers of parts by the temporary mounting of inexpensive fittings. Illustrated examples are considered in detail. K. recommends that (a) the part be carefully designed and the welding procedure accurately established and followed, (b) parts must be clean, this is of the first importance, (c) a satisfactory heat balance must be maintained between parts of different thickness, and (d) preliminary experimental welds should be made in the case of large projections.—P. R.

***The Strength and Safety of Welded Connections.** M. Roš (*Schweiz. Arch. angew. Wiss. Techn.*, 1946, 12, (1), 1-12; (2), 48-63; (3), 73-92).—A study of the working stresses in welded joints from considerations of experimental fatigue results and analytical methods. Assuming values for riveted joints and the Mises-Hencky relation for failure under combined stresses, expressions and curves are obtained for the permissible limiting-stress conditions in the fusion zone of: (1) longitudinal, circumferential, and spiral butt seams in pressure lines, boilers, and gas cylinders, and (2) V- and X-butt welds, longitudinal and transverse fillet welds in plates and beams. Both (1) and (2) are further subdivided into Classes I and II, which may or may not have been machined or heat-treated. The new formulæ, together with specifications for static tests, are designated EMPA 1945; they are compared with, and replace, those in use from 1935 to 1944.—E. N.

XXII.—INDUSTRIAL USES AND APPLICATIONS

Light Alloys in Shipbuilding. A. A. Pravdin (*Sudostroenie*, 1945, (3/4), 31-36).—[In Russian] Economic and engineering aspects of the application of light alloys in shipbuilding are reviewed.—V. K.

Progress Report on Light-Metal Bearings. R. Sterner-Rainer (*Motortech. Z.*, 1941, 3, 259-262; *Chem. Zentr.*, 1942, 113, (1), 260; *C. Abs.*, 1943, 37, 3378).—The application of light-metal alloys for slide bearings is reviewed. Special attention is given to the use of alloy KS 1275, a eutectic aluminium-silicon alloy with copper 1, nickel 1, and silver 1%. Light metals can replace bronze as bush metals when the characteristics of the light metals are properly taken into consideration.

Deformation Under Pressure of Brazed Aluminium Tank. L. W. Combs (*Product. Eng.*, 1943, 14, 581-582).

Aluminium and Its Alloys in Canning. T. D. Smith (*Sheet Metal Ind.*, 1946, 23, (228), 699-706; (229), 909-912).—A brief account of the use of aluminium and its alloys for canning in various countries. The materials available and their advantages are indicated, the methods of fabrication are described and illustrated, and methods of protection are given. The method of impact extrusion of cans of aluminium and its alloys is shown. Examples of the use of the materials are listed.—R. Gr.

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