Vol. 3 Part 7

The Monthly Journal of the INSTITUTE OF METALS

and

METALLURGICAL ABSTRACTS

CONTENTS

JULY 1936

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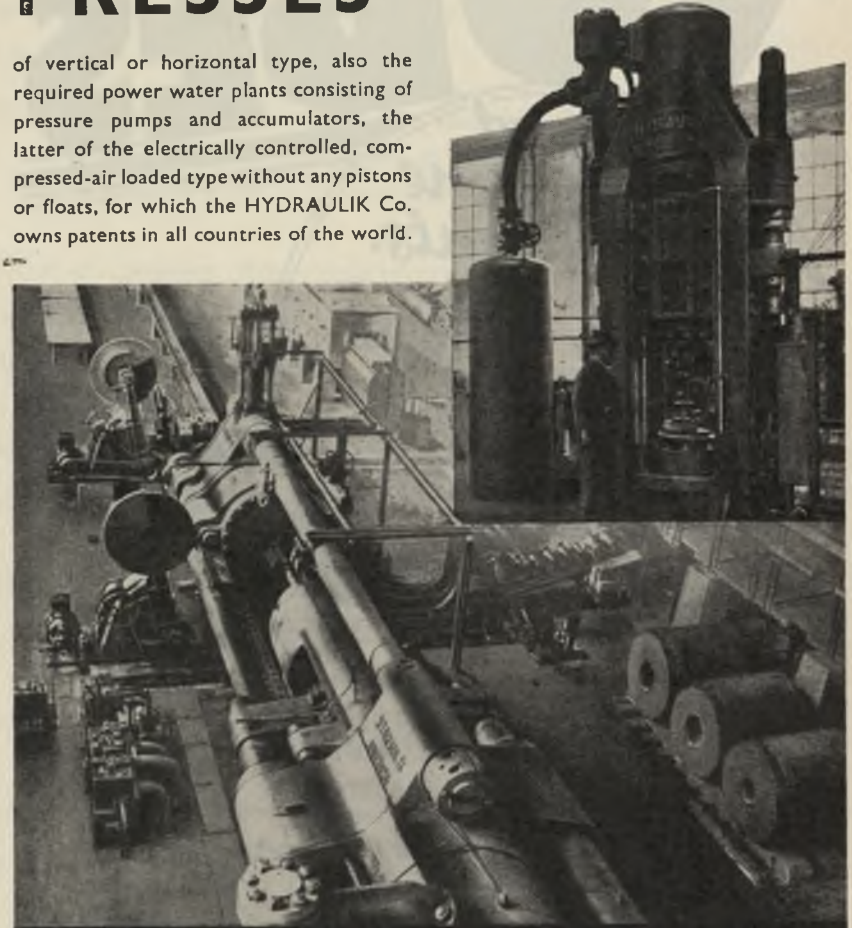

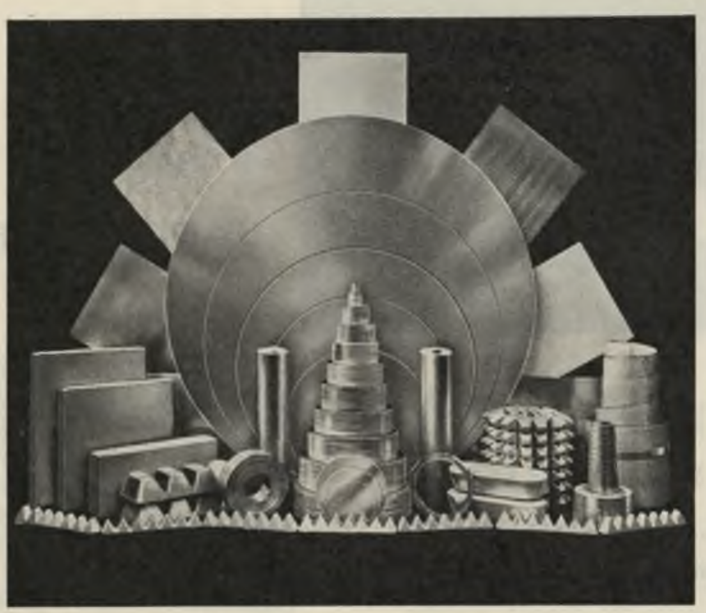

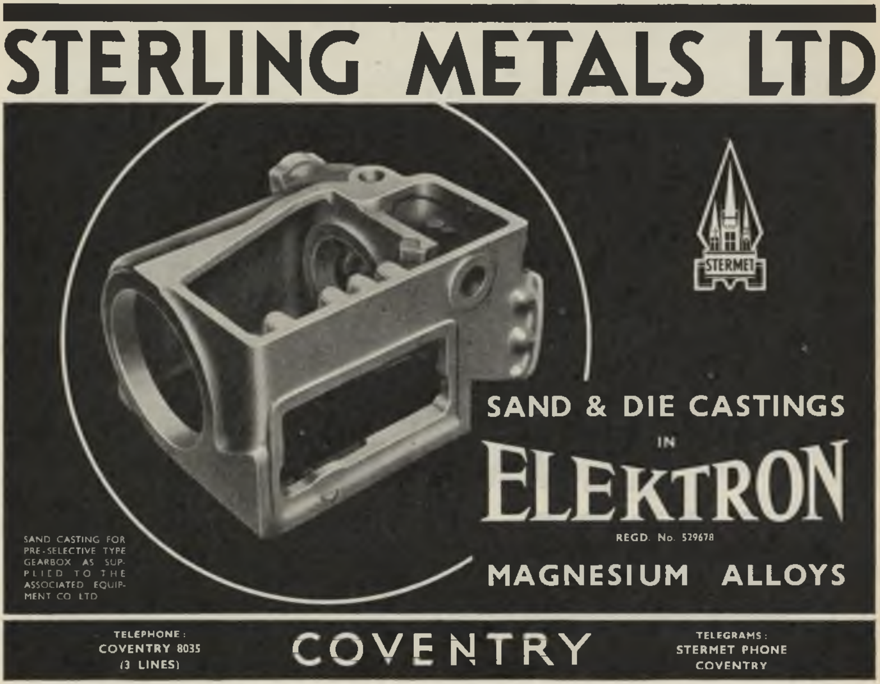

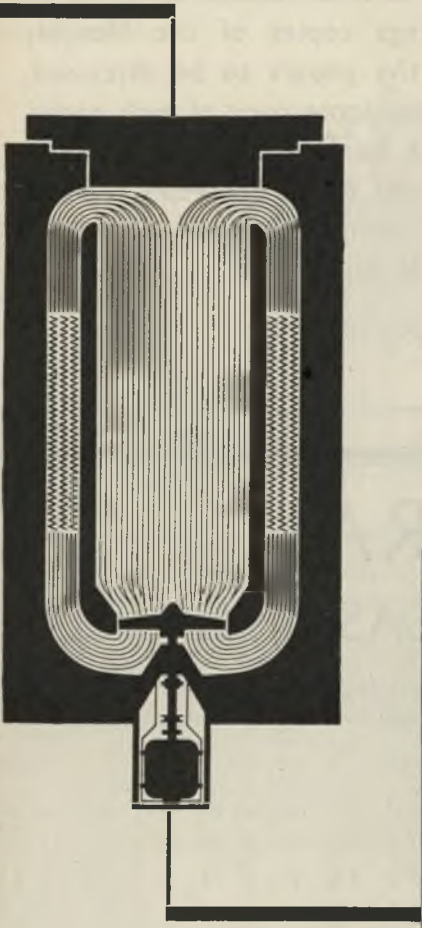

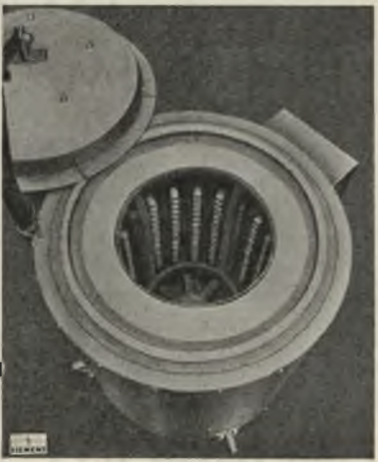
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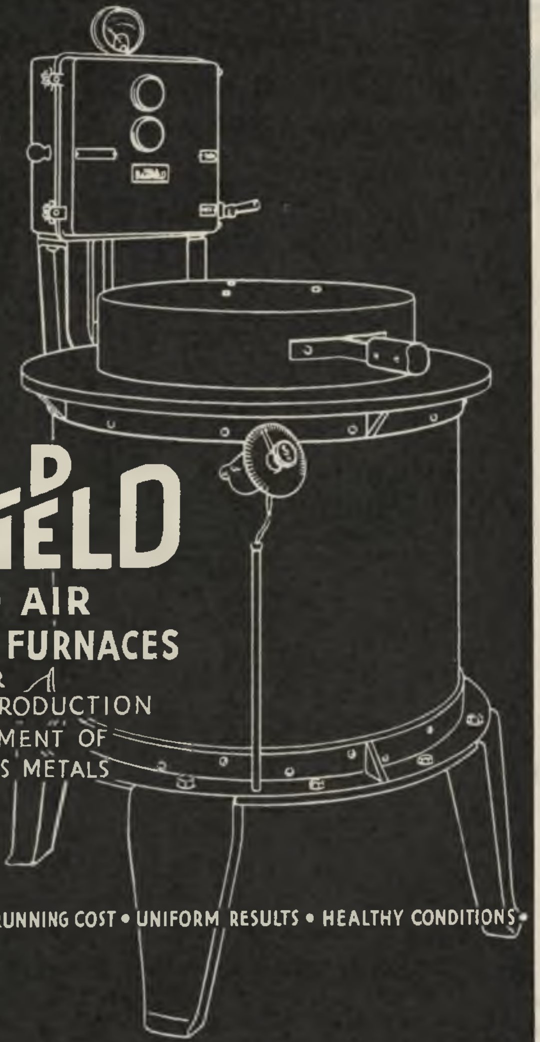

EDITORIAL OFFICE:
36 Victoria Street,
London, S.W.1.

TELEPHONE:
Victoria 2320.

Editor:
G. SHAW SCOTT,
M.Sc., F.C.I.S.

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Volume 3 JULY, 1936 **Part 7**

CONTENTS

METALLURGICAL ABSTRACTS

The monthly issue of *Metallurgical Abstracts* may be cut up for card indexes, as members will receive early in 1937 the year's abstracts in bound form.

INSTITUTE NEWS AND ANNOUNCEMENTS

Paris Autumn Meeting, Sept. 14–18.

On July 1 there was sent to every member a detailed programme of the Institute's Twenty-Eighth Annual Autumn Meeting, to be held in Paris from September 14 to 18, 1936. The programme was accompanied by a reply form, which each member attending the meeting is asked to return to the Secretary not later than August 15. In order than a member may retain in convenient form a record of the functions for which he has applied for tickets, a duplicate reply form, printed on pink paper, accompanied the programme, this form being marked " To be retained by member for information." There was also included with the programme a circular from Messrs. Thos. Cook & Son, Ltd., giving prices for travel and hotel accommodation in connection with the Paris Meeting. Members who desire to take advantage of one of the itineraries set out in this circular should return Messrs. Thos. Cook's " Travel Reply Form " also not later than August 15.

Members are reminded that the earlier they return their forms the better will be their chance of obtaining admission to the works that they particularly desire to visit (admittance to most of these works being limited to small parties), whilst the best hotel accommodation usually goes to early applicants.

Overseas members who do not wish to take advantage of one of the planned itineraries, can have similar arrangements made for them through the offices of Messrs. Thos. Cook & Son in their respective countries. Members travelling independently will find of service the list of hotels given in the " Travel " circular.

Any member who has not received the above-mentioned documents should apply immediately to the Secretary for duplicates.

In connection with the Paris meeting there has been arranged an attractive excursion to Rouen and Havre on September 18. Members will leave Paris by the 8.20 a.m. express, arriving at Rouen at 9.43 a.m. During the morning a tour of Rouen will be made with competent guides, and the following places of interest will be visited : Joan of Arc's Tower; the Old Market Place, l'Hôtel de Bourgtheroulde; the Clock Tower; the Palais de Justice; Notre-Dame Cathedral; the Church of St. Maclou; the Cloisters of St. Maclou and the Church of St. Ouen. After lunch, which will be served at mid-day in a first-class hotel, the party will divide. One group will proceed to Havre, arriving at 2.29 p.m. After a short trip by motor-coach round the Port there will be a visit to the wire-drawing and metal-rolling mills of the Usines des Tréfileries et Laminoirs du Havre, lasting from 3 to 4.45 p.m. The party will leave Havre by the 5.25 train for Paris, where they will arrive at 7.49 p.m.

The second party will spend the afternoon in making a motor-coach trip to the Abbey of St. Wandrille. The excursion will pass through beautiful country of historic interest. Returning to Rouen at 5.30 p.m. members will have tea and will join the first party on the express returning from Havre, which calls at Rouen at 6.30 p.m., both parties thus being back in Paris at 7.49 p.m.

The charge for the excursion is £1 14*s.* (Rouen) or £1 17*s.* (Havre). Bookings should be made either on the previously circulated reply form or direct to the Secretary. A detailed account of the excursion accompanied the Paris Meeting programme.

Co-operation with American Institute of Mining and Metallurgical Engineers.

A subject of special importance to younger metallurgists on both sides of the Atlantic has recently been under consideration by the Councils of the Institute of Metals and the American Institute of Mining and Metallurgical Engineers (Institute of Metals Division), and as a result of negotiations a scheme has now been concluded whereby younger members of either body who have not attained the age of 33 years can join the other on specially reduced terms.

To those English metallurgists who are eligible the advantages of contact with the American Institute and the receipt of its publications are obvious, and it is hoped the scheme will receive the cordial reception it deserves.

A further announcement, with details, will be made at as early a date as possible.

Council Nominations for 1937–1938.

In accordance with Article 22 of the Institute's Articles of Association, "At the ordinary General Meeting preceding the Annual General Meeting" (*i.e.* at the Paris Meeting on September 15) the Council will present the following list of members nominated by them for election to the Council :

President :

(Retires; eligible for re-election.)

W. R. Barclay, O.B.E.

Vice-Presidents :

Engineer Vice-Admiral Sir Robert Dixon, K.C.B., D.Eng.
Lieutenant-Colonel The Hon. R. M. Preston, D.S.O.

Members of Council :

W. T. Griffiths, M.Sc.
Stanley Robson, M.Sc.
A. J. G. Smout.
D. Stockdale, M.A., Ph.D.

In accordance with Article 22: "Any ten members may also, at or before such ordinary General Meeting, nominate in writing, with the written consent to act if elected of the person nominated, any duly qualified person other than one of those nominated by the Council . . . "

The following constitute the present Council (for 1936–1937) :

President :

W. R. Barclay, O.B.E.

Past-Presidents :

Harold Moore, C.B.E., D.Sc., Ph.D.
Richard Seligman, Ph.nat.D.
Professor Thomas Turner, M.Sc., A.R.S.M.

Vice-Presidents :

H. W. Brownsdon, M.Sc., Ph.D.
C. H. Desch, D.Sc., Ph.D., F.R.S.
A. G. C. Gwyer, B.Sc., Ph.D.
Professor D. Hanson, D.Sc.
H. C. Lancaster.
E. L. Morcom, M.A.

Honorary Treasurer :

John Fry.

Members of Council :

Professor J. H. Andrew, D.Sc.
Lieutenant-General Sir Ronald Charles, K.C.B., C.M.G., D.S.O.
Engineer Vice-Admiral Sir Robert Dixon, K.C.B., D.Eng.
S. F. Dorey, D.Sc., Wh.Ex.
R. Genders, M.B.E., D.Met.
Kenneth Gray.
Lieutenant-Colonel J. H. M. Greenly, C.B.E., M.A.
H. H. A. Greer, J.P.
J. L. Haughton, D.Sc.
D. J. Macnaughtan.
A. H. Mundey.
Lieutenant-Colonel The Hon. R. M. Preston, D.S.O.
C. J. Smithells, M.C., D.Sc.
F. Tomlinson.
H. B. Weeks.

and

The Chairmen of the Six Local Sections.

The following retire in March, 1937, in accordance with Article 21 (as revised on March 10, 1936) :

H. H. A. Greer, J.P.
A. G. C. Gwyer, B.Sc., Ph.D.
J. L. Haughton, D.Sc.
H. C. Lancaster.

Papers for the "Journal."

Members are reminded that papers approved by the Publication Committee are accepted for publication in the "Monthly Journal" (and subsequently in the "Journal") but *not necessarily for presentation at any meeting of the Institute.*

Prior to a general meeting of the Institute, a selection is made from the papers considered most suitable for oral discussion and which have been printed in the "Monthly Journal" since the previous meeting. These and all other papers published may of course be discussed by correspondence.

Normally the last dates for the receipt of papers that could be considered for discussion at either the Autumn Meeting or the March Meeting are, respectively, April 30 and October 31. It is, however, impossible to give an undertaking that any accepted paper, even if received on or before one of these dates, will necessarily be published prior to the next following General Meeting, as the date of publication is also dependent on the number of papers previously received and accepted.

Financial considerations make it impossible for the Institute to publish more than a pre-determined amount of matter in each half-yearly volume.

Already several papers have been received which cannot be published before the Autumn Meeting, but which will be published before, and may be selected for discussion at, the Annual General Meeting. Authors are requested, therefore, promptly to advise the Secretary of papers which they intend to submit during the next few months.

Intending authors of papers are referred to the " Notice to Authors of Papers " appearing on p. 298.

Joint Meeting with Manchester Metallurgical Society.

The usual annual joint meeting of the Institute of Metals with the Manchester Metallurgical Society will be held in Manchester on Wednesday, November 4, 1936, when some of the Institute's Paris Meeting papers will be re-presented for discussion. The President of the Institute of Metals, Mr. W. R. Barclay, is expected to preside.

" The Structure of Metals and Alloys."

Some weeks ago the Institute issued the first of its new Monograph and Report Series of publications entitled " The Structure of Metals and Alloys," by W. Hume-Rothery, M.A., Ph.D. Published at 3*s*. 6*d*. net, post free, the book has had an extensive sale in all parts of the world. About one thousand copies have been distributed free to members in response to the Council's offer, made when the book first appeared, to send a copy without charge to any member of the Institute on application being received by the Secretary.

Any member who has not already received a copy of the book, and desires to possess one, should write to the Secretary. If additional copies are required these can conveniently be ordered at the same time, the order being accompanied by a remittance at the rate of 3*s*. 6*d*. per copy.

Reinstatement of Former Members

During the recent period of trade depression some members of the Institute were unable to continue their subscriptions and their names, therefore, were removed from the Roll. Lately several such members have applied for reinstatement, which the Council has been glad to arrange—in the case of members who resigned in good standing—simply on the basis of the payment of the current year's subscription.

As a new financial year of the Institute began on July 1, it is possible that other former members may wish to resume their membership on this basis, and this notification is, therefore, made at the suggestion of the Membership Committee in the hope that it will be communicated by present members to any of their friends who were formerly on the membership roll and who might welcome the opportunity to resume their Institute membership.

Subscriptions for 1936–1937.

The present issue of the *Monthly Journal* will reach all members, even though their subscriptions for the current year may not have been paid.

Many members have already been good enough to remit their subscriptions for the new financial year, which commenced on July 1, without receiving the usual subscription " reminder." The subscriptions of over 550 other members were paid on July 1 by means of Banker's Orders : in this way a very considerable saving of labour and expense has been effected. It is the hope of the Council that still more members will take advantage of this convenient method of payment. All that the member has to do is to fill in a form, to be obtained from the Secretary, instructing his banker to pay his subscription to the Institute's banker every year. This saves the member writing of cheques, stamp duty on cheques, addressing envelopes, and postage, and it saves the Institute the preparation of receipts, addressing envelopes, and postage, for, as the money passes direct from one bank to another, no receipt is necessary. It also obviates the need, in certain cases, for issuing several notices regarding subscription arrears.

To those members whose subscriptions had not been paid by the date due, July 1, there have already been sent letters from the Secretary reminding them that the amount payable for the year 1936–1937 is £3 3*s.* in the case of Ordinary Members or £1 1*s.* in the case of Student Members. MANY POUNDS WILL BE SAVED TO THE INSTITUTE IN SENDING FURTHER REMINDERS IF MEMBERS WHO HAVE NOT YET PAID THEIR SUBSCRIPTIONS WILL BE GOOD ENOUGH TO FORWARD THEM WITHOUT DELAY.

Researches on Lead.

Professor J. NEILL GREENWOOD, D.Sc., of Melbourne University, Australia, has obtained a grant of £1000 per annum from the Broken Hill Associated Smelters in order to extend the research work on the physical properties of lead that is in progress in the metallurgical laboratory of the University of Melbourne.

Professor Greenwood would be glad to get into touch with fellow members of the Institute in other countries—particularly in France, Germany, U.S.S.R., and U.S.A.—who may be working on the physical properties of lead. Such members are invited, therefore, to write to Professor Greenwood with regard to their own researches—or researches which they may be aware are now being carried out on this subject. Letters should be addressed to Professor J. Neill Greenwood, D.Sc., the University, Melbourne, N.3, Australia.

Loose Copies of Papers.

Many members do not wish to bring with them to the meetings of the Institute their copies of the *Monthly Journal* containing the papers to be discussed. To meet the convenience of such members the Council has authorised the supply of specially prepared loose prints of the papers.

Orders for complete sets of the papers appearing for 1 year in the *Monthly Journal* may now be sent to the Secretary accompanied by a remittance for 5*s.* This payment will ensure the despatch, post free, of loose copies of all the papers to be read at the Paris Meeting in September and at the Annual General Meeting next March.

PERSONAL NOTES

The Editor requests that his attention be directed to items of interest to members that might be included under the "Personal Notes" heading. All contributions for the August issue of the Monthly Journal *should reach him not later than July* 25.

SIR HERBERT AUSTIN was created a Baron on the occasion of the King's Birthday. He has assumed the title of Lord Austin of Longbridge. Lord Austin, who has been a member of the Institute since 1928, is head of the motor-car company that bears his name.

SIR HAROLD CARPENTER, F.R.S., Past-President, has been elected a corresponding member of the Ingeniörs Vetenskaps Akademien, Stockholm (the Royal Swedish Institute for Engineering Research).

MR. O. W. ELLIS, M.Sc., Director of Metallurgical Research, Ontario Research Foundation, has been nominated to the office of Trustee of the American Society of Metals for the ensuing two years.

MR. T. J. HARTIGAN, Commissioner for Railways, State of New South Wales, was gazetted C.M.G. in the Birthday Honours List.

MR. D. F. PHILLIPS has left Birmingham to take up an appointment as Chief Chemist to Messrs. E. Hind (South Bank), Limited, South Bank-on-Tees, Yorkshire.

PROFESSOR ALBERT PORTEVIN was recently in London in connection with the International Congres of Acetylene, Oxy-Acetylene, Welding, and Allied Industries and visited the offices of the Institute.

Personal Notes

Mr. Gilbert Rigg is now on a visit from Australia to London. Letters will be forwarded to him if addressed C/o The English, Scottish, and Australian Bank, Limited, 5 Gracechurch Street, E.C.3.

Professor Dr.-Ing. Max Schlötter, who for some time past has been associated with Dr.-phil. G. Lüttke as joint editor of *Metallwirtschaft*, has succeeded Dr.-Ing. Paul Rosbaud as Director of the technical section of that publication.

Dr. Alan E. W. Smith, B.Sc., has accepted an appointment as junior lecturer in metallurgy and chemistry at the Military College of Science, Woolwich. Dr. Smith served for many years as Honorary Secretary and Treasurer of the Swansea Local Section of the Institute of Metals.

Dr. Cyril S. Smith, B.Sc., of Waterbury, Conn., U.S.A., recently visited England, returning to America in the R.M.S. *Queen Mary*.

Mr. James L. Wick, Jr., President and General Manager of the Falcon Bronze Company, Youngstown, O., has been elected President of the American Foundrymen's Association.

Mr. T. Wolverson has been appointed Foundry Manager to Messrs. Leeds Piston Ring and Engineering Company, Limited.

Marriage.

The marriage of Dr. L. H. Callendar, F.I.C., D.I.C., to Miss W. M. Copeland, of Malone Road, Belfast, took place at Belfast on June 29.

Birth.

Heywood. To Dr. and Mrs. Harold Heywood (*née* Frances Weaver), B.Sc., Ph.D., on June 13, 1936, a daughter.

NOTICE TO AUTHORS OF PAPERS

1. Papers will be considered for publication from non-members as well as from members of the Institute. They should be addressed to The Secretary, Institute of Metals, 36 Victoria Street, London, S.W.1.

2. Papers are accepted for publication in the *Monthly Journal* and subsequently in the *Journal*, and not necessarily for presentation at any meeting of the Institute.

3. 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.

4. *Manuscripts* and illustrations must be submitted in duplicate, and an abstract, for publication, must accompany each paper. Manuscripts 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. MS. not accepted will be returned within 6 months of receipt.

5. *References* should be collected at the end of the paper, and set out in the following style: (1) W. Hofmann and W. Jäniche, *Z. Metallkunde*, 1936, **28**, 1 (*i.e.* year, vol., page).

6. *Illustrations.* Each illustration must have a number and description; only one set of numbers must be used in one paper. *Line figures* must be drawn in India ink on smooth white Bristol board, tracing paper, 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 *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 2¼ in. wide (four to a plate); 2 in. deep by 2¼ in. wide (six to a plate). Magnifications of photomicrographs should be given in each case.

7. *Tables or Diagrams.* Results of experiments, &c., may be given in the form of tables *or* figures, but (except in exceptional circumstances) 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). Authors may elect to receive their reprints from the *Monthly Journal* (without discussion) or from the *Journal* (complete with discussion). Additional reprints are obtainable at rates to be obtained from the Editor.

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VEINING AND SUB-BOUNDARY STRUCTURES IN METALS.*

By L. NORTHCOTT,† Ph.D., M.Sc., MEMBER.

SYNOPSIS.

A network structure, termed veining, which occurs within the crystals has been studied in several pure metals and in many alloys of copper, and methods for its production are discussed. The structure is shown to be due to the precipitation from solid solution, in a network form, of the oxide of the metal during cooling. The influence of alloying additions on veining in copper has been examined and the effect of heat-treatment and recrystallization on veining correlated with changes in hardness. Similar structures referred to as sub-boundary structures have been observed in a number of other alloys. These structures are due to the precipitation, also during cooling, of a constituent of the alloy.

(1) INTRODUCTION.

VEINING structures have been considered in the past mostly in connection with fairly pure irons or low-carbon steels. The structure takes the form of a network, usually irregular, inside the α-iron crystals and has been termed α-veining or the sub-boundary structure. In a recent investigation,[1] it was concluded that the structure is due to the precipitation of iron oxide during cooling and that the solid solubility of the oxide in iron decreases with decrease in temperature, a conclusion which has been confirmed by analytical methods.[2-4] In the same paper,[1] examples were given of veining in copper and nickel, and with both these metals it was also concluded that veining results from the precipitation of the oxide of the metal during cooling. The decrease in the solid solubility of copper oxide in copper with decrease in temperature has since been confirmed by chemical analysis.[5]

The investigation of veining structures has now been extended to non-ferrous metals and alloys generally, and the results are described in this paper.

(2) PRODUCTION OF VEINING OR VEINING-FREE METALS.

As the earlier investigation [1] had shown the dependence of veining, in the case of iron, nickel, and copper, on the oxidation of the metal,

* Manuscript received April 15, 1936.

† Research Department, Woolwich.

Note to Abstractors and Other Readers.—This paper will be published, in permanent form, in the *Journal of the Institute of Metals*, Vol. LIX, 1936. Reference should accordingly be as follows: *J. Inst. Metals*, 1936, **59** (Advance copy).

methods of increasing the oxide content were selected for the purpose of obtaining material in which veining might be present. The methods adopted were : (*a*) Melting the metal in an oxidizing atmosphere. In some instances it has been found sufficient to melt the material without using a flux or reducing agent. With lower melting point metals, the practice commonly adopted was to pour the melt from one crucible to another several times in air; alternatively air has been blown through the melt, using a silica tube to introduce the air well below the surface of the metal; (*b*) Annealing the metal in air; (*c*) Heating the metal in contact with its oxide. In this case the oxide is packed either around the polished metal or in a hole drilled in it.

In all cases it has been found necessary to arrange for a slow rate of cooling.

A comparison sample of many of the metals was prepared by passing hydrogen through the melts and solidifying in an atmosphere of hydrogen. The intention was to obtain material free from oxide.

(3) VEINING IN PURE METALS.

All the metals examined were of the highest purity commercially available. Of the metals which had been oxidized according to one or more of the methods described above, veining has been observed in cadmium, zinc, antimony, magnesium, aluminium, silver, copper, nickel, and iron.

Veining has not been observed, in the present investigation, in either tin or lead. The structures observed in the individual metals (taken in the order of their melting points) are briefly described below. A reference to recommended etching reagents and methods will be found in the Appendix (pp. 332–333).

Cadmium.—A typical structure is shown in Fig. 1 (Plate I). The two crystals in the photograph are easily distinguished by the difference in orientation of the twins. It will be noted that the direction of the veins does not appear to be affected by the twins. No veining was observed in the original cadmium or in the hydrogen-treated sample.

Zinc.—Two qualities of zinc were examined containing 99·99+ per cent. and approximately 99·98 per cent. zinc, respectively. Neither sample showed veining in the condition as received or after the hydrogen treatment, but much veining was observed after treating the melt with air. The microstructure of zinc is similar to that of cadmium, and it was also observed with zinc that twinning did not appear to cause any discontinuity in the veins.

Magnesium.—A sample of 99·95+ per cent. magnesium showed no veining as received, whereas the air-treated material showed veining.

PLATE I.

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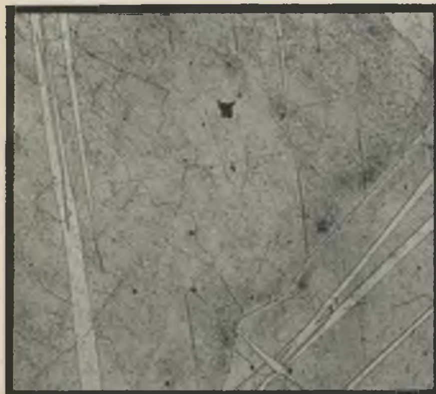

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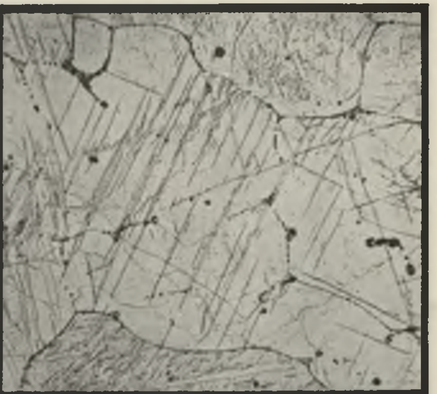

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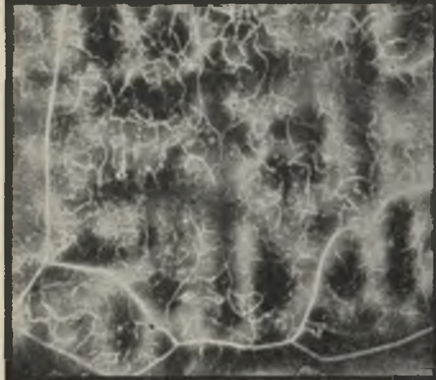

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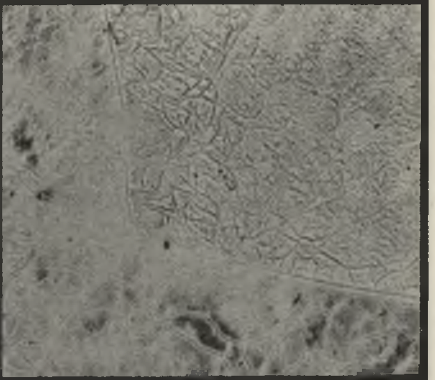

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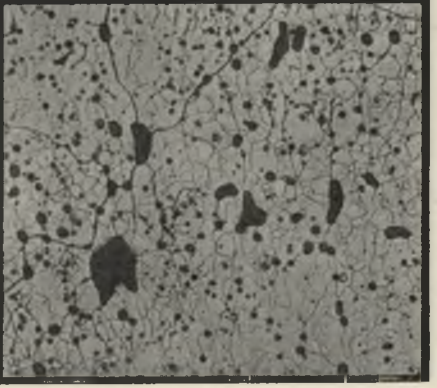

FIG. 1.—Cadmium. × 50.
FIG. 2.—Magnesium. × 50.
FIG. 3.—Copper, Dendritic Structure and Veining. × 15.
FIG. 4.—Copper. × 25.
FIG. 5.—Copper, Hydrogen Treated in the Melt, Veining Absent. × 20.
FIG. 6.—Wrought Iron. × 100.

PLATE II.


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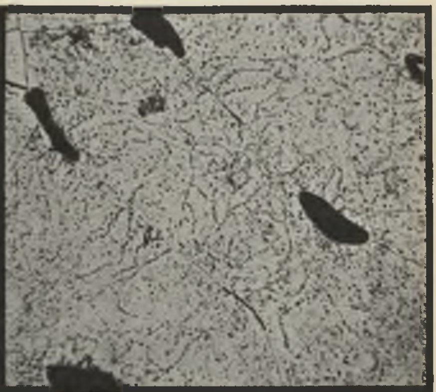
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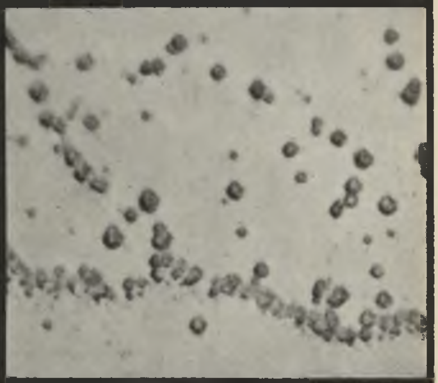
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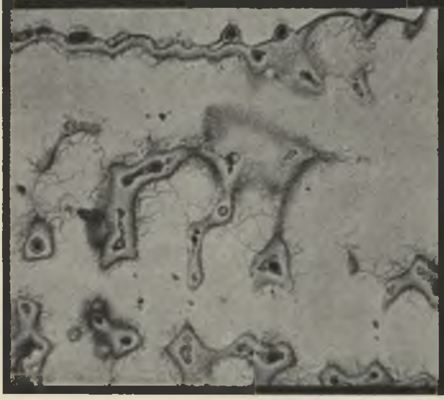
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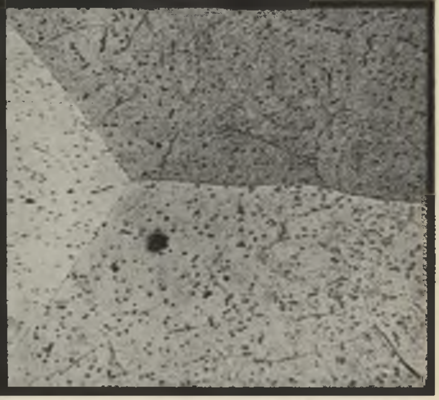
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FIG. 7.—Mild Steel, 0·3% C. × 250.
FIG. 8.—Silver. × 100.
FIG. 9.—95 : 5 Copper–Zinc Alloy. × 40.
FIG. 10.—70 : 30 Brass. × 2500.
FIG. 11.—70 : 30 Brass. × 50.
FIG. 12.—52 : 48 β-Brass. × 50.

PLATE III.

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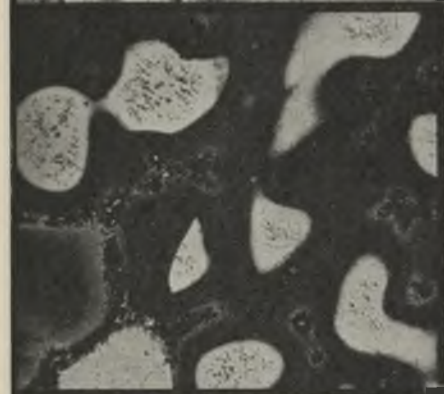

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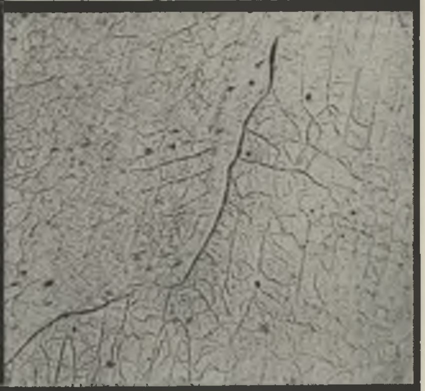

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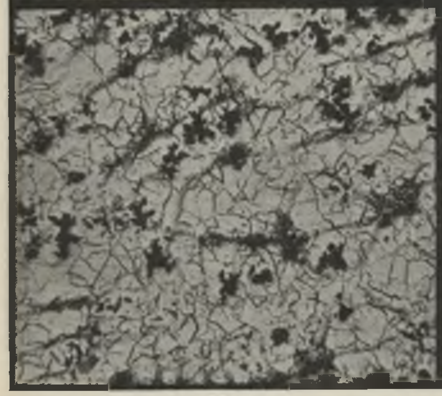

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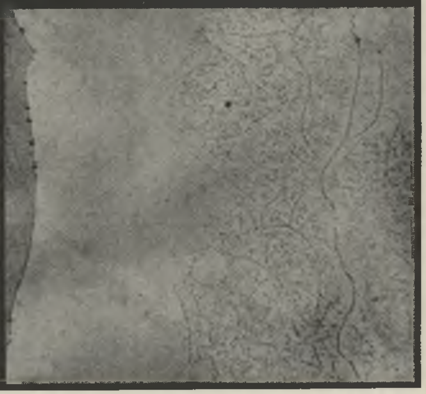

FIG. 13.—93 : 7 Copper–Tin Alloy. × 100.
FIG. 14.—93 : 7 Copper–Tin Alloy. × 2500.
FIG. 15.—Copper Containing 0·5% Phosphorus. × 50.
FIG. 16.—92 : 8 Copper–Aluminium Alloy. × 50.
FIG. 17.—70 : 30 Nickel–Copper Alloy. × 100.
FIG. 18.—Copper Annealed for 2 Hrs. at 800° C. and Quenched.

PLATE IV.


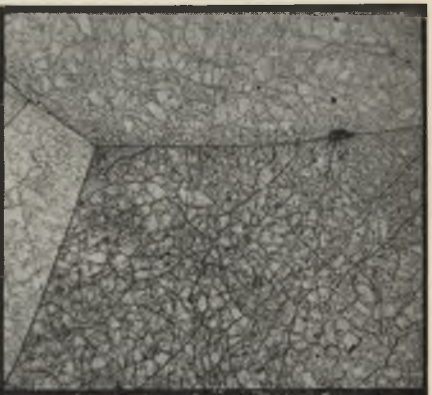
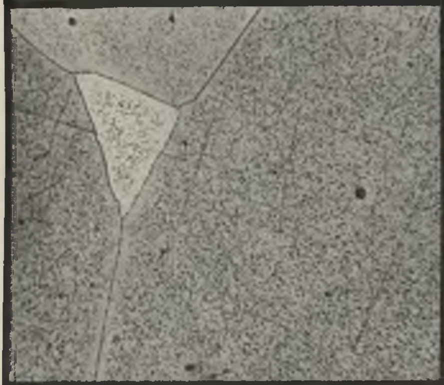
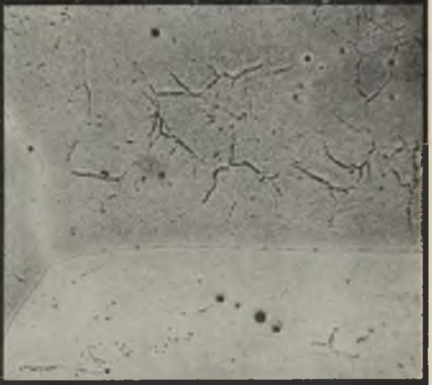
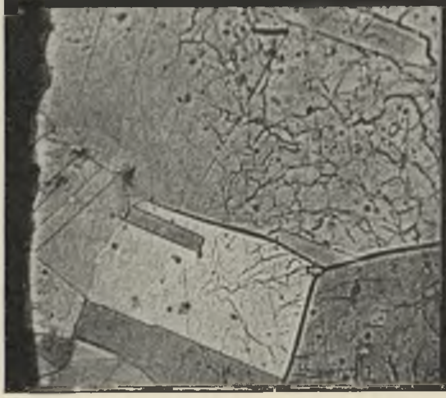
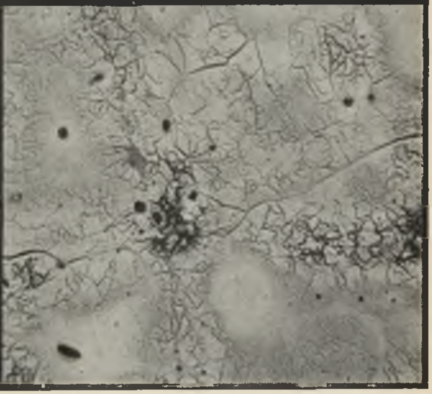

FIG. 19.—70 : 30 Nickel–Copper Alloy, Annealed for 2 Hrs. at 750° C. and Quenched. × 100.
FIG. 20.—Copper Quenched from 1000° C. and Tempered for 20 Hrs. at 400° C. × 50.
FIG. 21.—Copper Quenched from 1000° C. and Tempered for 20 Hrs. at 750° C. × 50.
FIG. 22.—95 : 5 Copper–Zinc Alloy, Slowly Cooled 1000°–600° C. and Quenched. × 50.
FIG. 23.—98 : 2 Copper–Tin Alloy, Annealed in Hydrogen at 900° C. for 22 Hrs. Transverse Section, Showing Surface Removal of Veining. × 25.
FIG. 24.—90 : 10 Copper–Tin Alloy, Annealed 3 Hrs. at 800° C. and Slowly Cooled. × 50.

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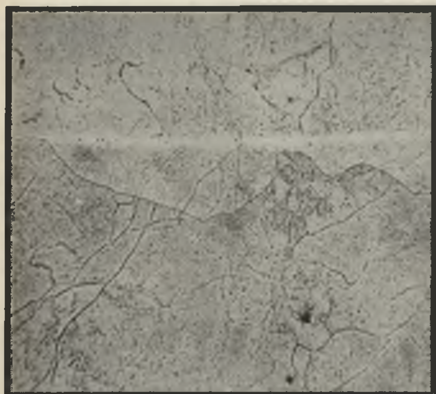


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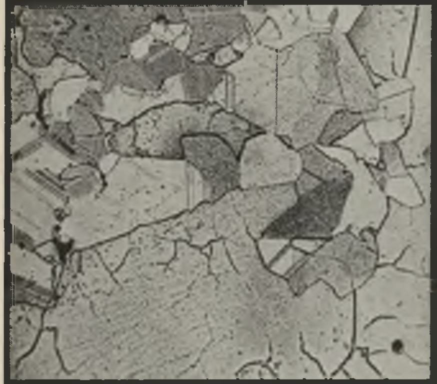


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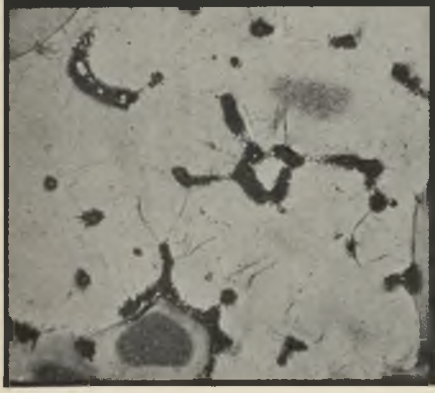

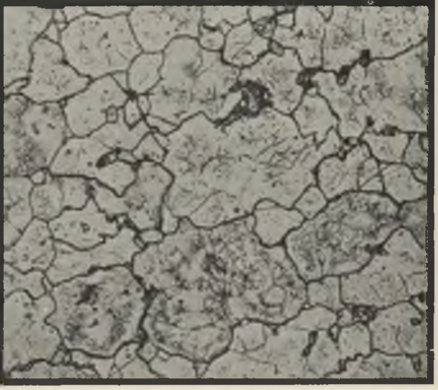
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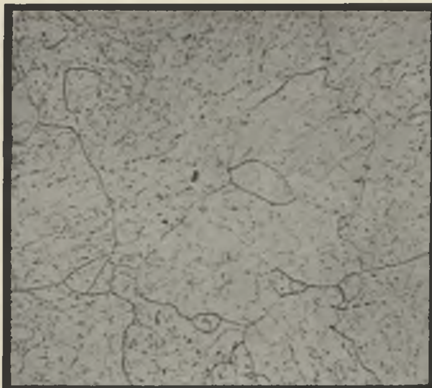
FIG. 25.—Grain-Growth in Copper. × 50.
FIG. 26.—Veining in Recrystallized Copper. × 15.
FIG. 27.—95 : 5 Copper–Zinc Alloy, Partially Recrystallized. × 100.
FIG. 28.—Interdendritic Recrystallization in 95 : 5 Copper–Zinc Alloy. × 25.
FIG. 29.—Copper Containing 0·5% Phosphorus. × 40.
FIG. 30.—Molybdenum–Iron Alloy. × 100.



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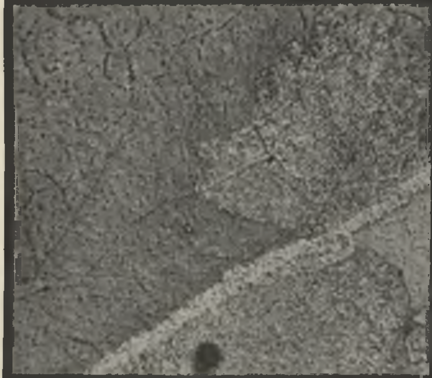

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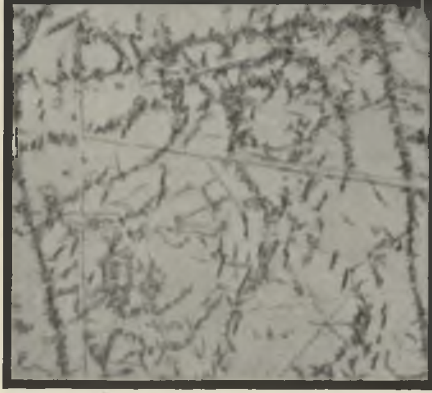

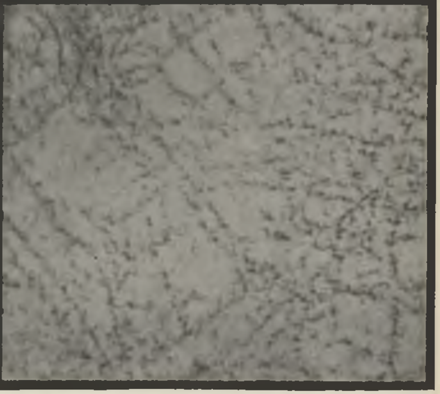

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FIG. 31.—Tantalum–Iron Alloy. × 100.
FIG. 32.—Network in Two Crystals of β-Constituent of Nickel–Iron–Aluminium–Copper Alloy. × 100.
FIG. 33.—Nickel–Iron–Aluminium–Copper Alloy, $\gamma \longrightarrow \delta$. × 250.
FIG. 34.—Air-Cooled Beryllium–Copper Alloy. × 100.
FIG. 35.—Air-Cooled Beryllium–Copper Alloy. × 2500.
FIG. 36.—Beryllium–Copper Alloy Air-Cooled for 1½ Minutes and Quenched. × 2500.

The appearance of both veining and crystal structure of magnesium, Fig. 2 (Plate I), was similar to that in the case of cadmium.

Aluminium.—Veining, which was not observed in the original sample, was found in the air-treated sample, and was still observed after treating this with hydrogen. Hydrogen treatment of the melt for ½ hr. was sufficient to remove veining from most metals, but was found to have no effect on the structure in aluminium.

Copper.—The commercial varieties of cast or wrought copper examined all showed oxide present in the usual globular form but not as veining. Such copper remelted without charcoal or flux showed an excess of the copper–copper oxide eutectic and no veining. When excessive oxidation of the melt was prevented, however, and the metal slowly cooled, veining was present. Several samples of copper were obtained showing veining; all consisted of large crystals, up to 1 cm. in diameter, due to the slow solidification, and the veining was both coarse and well-defined. Typical structures are illustrated in Figs. 3 and 4 (Plate I). The actual appearance of veining, especially in copper, depends very much on the conditions of polishing and etching as well as on the magnification and focussing. The veins may either appear light against a dark background or dark against a light background. Occasionally it has been necessary to photograph the structure slightly out of focus in order to obtain sufficient contrast for reproduction. At low magnifications the veins appear as continuous lines, but usually the discontinuous nature of the veins can be shown in copper at quite moderate magnifications. In addition to the dotted appearance of the veins at higher magnifications, the apparent absence of any connection between veining and dendritic structure will be noticed. When treated in the liquid state with hydrogen for a sufficient length of time, veining was removed (see Fig. 5, Plate I).

Nickel.—Veining was induced in pure nickel by annealing the metal in air for many hours, as described in a previous paper.[1] The depth of penetration was considerably less with nickel than with those metals described previously.

Iron.—Good examples of veining have frequently been observed in different samples of wrought irons and in steel (Figs. 6 and 7, Plates I and II), but not in electrolytic iron as deposited. No difficulty was experienced in inducing veining in the latter either by annealing in air or in contact with iron oxide at a high temperature. The study of the veining in iron and steel has been described in some detail in the previous paper.[1]

Silver.—Silver is considered last, as the veining in this metal is of a different type from that described above. Two qualities of silver were examined, the first was 99·9+ per cent. pure and the second, which was obtained through the courtesy of Mr. D. McDonald of Messrs. Johnson, Matthey & Company, Ltd., was of 99·99 per cent. purity. No veining was observed in samples prepared by bubbling hydrogen through the melt and allowing solidification to take place in an atmosphere of hydrogen, but veining was present after the silver had been poured to and fro in air, then slowly cooled in the furnace (Fig. 8, Plate II). Veining in silver may be considered as a special type on account of the fact that although silver oxide Ag_2O exists, it decomposes at temperatures in the neighbourhood of 300° C., although the dissociation temperature is known to increase with increase in pressure. Further reference to the structures in silver will be made later in the paper.

Of the pure metals examined, tin, lead, and antimony failed to show veining. Some difficulty was expected with the first two metals in view of the comparatively small range of temperature between their melting points and atmospheric temperature; their extreme softness and the facility with which they recrystallize only added to difficulties in examination. Greater uncertainty was felt about antimony; several samples were prepared by oxidizing the melts, but the oxide occurred too readily in globular form similar to the oxide eutectic in copper. The addition of an alloying element was, therefore, made for the purpose of deoxidizing, on the lines described below for copper. Antimony, unlike copper, forms only a few alloys in which one phase alone occurs over any appreciable range in composition at the antimony end of the diagram. An alloy was prepared containing 2 per cent. tin, and this resulted in the suppression of excess globular oxide and the formation of a discontinuous veining network somewhat similar to that observed in silver. When the rate of cooling was increased by air cooling from 590° C. a sample of dimensions approximately 2 in. × 1 in. × $\frac{1}{2}$ in., the network structure was more pronounced. It is probable, therefore, that pure antimony would also show veining if the state of oxidation were more closely controlled than it had been in the present work.

(4) VEINING IN ALLOYS.

An investigation was made to determine the effect of a second constituent on the formation of veining. Copper was selected as the basis metal in view of its industrial importance and the comparative ease with which it was possible to develop veining in suitably prepared samples. The general procedure adopted was to melt 4–6 lb. of electrolytic copper under charcoal, add the required amount of the second element, pour the alloy to and fro in air about six times, using a second pre-heated crucible, and then to allow solidification to take place in the crucible in the furnace. The ingot was sectioned along its axis and the axial face examined for veining. Particulars of the alloys prepared are given in Table I.

TABLE I.—*Copper Alloys Prepared by Air-Treating the Melt.*

Added Element.	Percentage of Addition.	Constitution.	Veining?
Zinc	5	α-solid solution	pronounced coarse veining
	10	α	ditto
	15	α	less pronounced veining
	30	α	slight veining
	39	α (+ β)	very slight veining
	40	$\alpha + \beta$	trace of veining in β
	48	β	chiefly random deposit, some network
	50	β (+ γ)	veining present
	61	γ	possibly ditto
Tin	2	α	pronounced coarse veining
	5	α	much veining
	6	α (+ δ)	ditto
	7	$\alpha + \delta$	ditto
	10	$\alpha + \delta$	slight veining
Phosphorus	0·2	α (+ Cu_3P)	veining present
	0·55	$\alpha + Cu_3P$	P network present
	0·5	$\alpha + Cu_3P$	veining and P network present
Silicon	0·5	α	much veining present
	4	$\alpha + \gamma$	slight veining
	5	$\alpha + \gamma$	very slight veining
Aluminium	2	α	veining present
	6	α	ditto
	8	α	ditto
Arsenic	2	α	veining present

The effect on veining of the various additions to copper may separately be described briefly as follows:

Copper–Zinc Alloys.—Additions of small quantities of zinc to copper resulted in veining which was more pronounced than in copper alone, as is shown in the 5 per cent. zinc alloy in Fig. 9 (Plate II). Comparatively light etching developed the structure chiefly in the interdendritic areas, but continued polishing and etching showed veining to exist throughout the crystals. The veins appeared as fairly coarse lines, especially after deep etching, but a light treatment and examination at higher magnifications (see reference to Fig 10 below) showed the discontinuous nature of the veins. The addition of zinc eliminated the coarse globules of Cu_2O which are a common feature of most commercial coppers. Increasing the zinc content of the α-solid solution had little effect other than to make the structure less coarse and much less easily developed in the dendritic axes of the crystals (Figs. 10 and 11 (Plate II), 70 : 30 brass) but veining could still be observed. In the 60 : 40 alloy, consisting of approximately equal proportions of the α- and β-solid solutions, a possible trace of veining was observed in the β-solid solution only. In the next sample, containing sufficient zinc to ensure an all β-alloy, the distribution of the deposit was random (similar to that in Fig. 21, Plate IV) and was developed equally throughout the crystals as the freezing range was insufficient to cause much " coring." Another sample prepared by allowing the crucible containing the melt to cool in air showed a somewhat similar structure, except that there was some evidence of a network formed by a linking up of the separate particles (Fig. 12, Plate II). A more complete network structure was obtained as described later by cooling a small sample rapidly in air. With sufficient zinc to saturate the β-solid solution and to ensure the presence of a small quantity of γ, veining was again observed, similar to, but not so pronounced as, that shown in Fig. 16, Plate III. The alloy consisting solely of the compound Cu_5Zn_8, which is the principal constituent of the γ-solid solution, was exceptionally difficult to prepare for micro-examination, but there was again a definite suggestion of veining.

Copper–Tin Alloys.—When small additions of tin were made to copper, pronounced veining of the type found in the low-zinc brasses was again observed. Further additions of tin appeared to reduce the quantity and also to refine the veining, and greater difficulty was experienced in developing it. Veining was present, however, even in alloys containing the δ-solid solution and Fig. 13 (Plate III) shows a bronze containing sufficient tin to ensure the presence of the δ-constituent, which appears as small lakes. The lines of the network in this photograph appear singularly continuous, and it was only by examination under very high magnifications that the discontinuity became apparent. This is shown in Fig. 14 (Plate III), at a magnification of × 2500 obtained by the use of the objective of N.A. 1·60 developed by H. Wrighton [6] for high-power work. With this objective all the veins were easily resolved; there appeared to be no connecting lines between the individual particles constituting the veins. In the 10 per cent. tin alloy the veining was most prominent in the dendritic areas of the crystals.

Copper–Phosphorus Alloys.—The first two alloys, containing 0·2 and 0·55 per cent. phosphorus, respectively, were prepared by adding the required amount of phosphor–copper (15 per cent. phosphorus) to the copper melted under a thin layer of charcoal, and allowing it to cool down slowly in the furnace without pouring to and fro in air. Veining was observed in the 0·2 per cent. phosphorus alloy, of a type similar to, although not so pronounced as, that found in the original copper, but no oxide veining was observed in the 0·55 per cent. phosphorus alloy. In this there were several isolated particles of free phosphide, and in addition a secondary deposit of phosphide indicative of veining was observed. The latter structure was somewhat irregular and was confined to the interdendritic areas adjacent to the major phosphide deposits. The addition of 0·55 per cent. phosphorus appeared to have been sufficient to deoxidize the copper completely. The third alloy containing 0·5 per cent. phosphorus was poured to and fro in air six times before being allowed to cool, and the structure was for the most part similar to that of the second alloy containing 0·55 per cent. phosphorus. Several crystals in the third alloy, however, showed both veining and the phosphide network, the former in the middle, dendritic areas and the latter in the interdendritic, phosphide-rich areas of the crystals. The rate of cooling was not the best for the production of veining in a good network form (Fig. 15, Plate III).

Copper–Aluminium Alloys.—The veining in the 2 per cent. aluminium alloy was similar to that observed in the low alloy brasses and bronzes. Veining in the 8 per cent. alloy was finer (Fig. 16, Plate III) and more general throughout the crystals, and required prolonged etching to develop the structure as may be seen from the coarse nature of the grain boundary. Subsequently, long annealing at a temperature just below the solidus failed to remove the structure which, therefore, should not be mistaken for dendritic segregation.

Copper–Silicon Alloys.—The 0·5 per cent. silicon alloy showed veining similar to that in the low-zinc or low-tin copper alloys. Increasing the silicon content to 4 per cent. introduced the γ-constituent and considerably increased the difficulty of developing veining, but veining was definitely present although it was confined to the dendritic areas. This feature is of interest in view of the general tendency for impurities to concentrate in the interdendritic spaces, and would appear to be due to the smaller solubility of the oxide in the silicon-rich portions of the cored α-solid solution.

Copper–Arsenic Alloy.—Veining was present in an alloy to which 2 per cent. of arsenic had been added, and was similar in appearance to that in the low-tin alloy except that it was less pronounced.

Nickel–Copper Alloys.—In addition to the specially prepared alloys described above, veining has been observed in other alloys. As two of these have been employed in some of the heat-treatment experiments, brief descriptions may be made here. One alloy, ECN, was a cupro–

nickel containing 70 per cent. nickel and 0·6 per cent. graphite, the remainder being copper; the samples examined were cut from a chill-cast ingot of dimensions 36 in. × 3 in. × 2 in. The structure of the alloy as received is shown in Fig. 17 (Plate III); in addition to abundant veining, coring and graphite deposits were also in evidence. The dimensions of the veining network were much smaller than in the majority of the other non-ferrous alloys examined. The other alloy, LAQ, containing 20 per cent. nickel, was also received in the " as cast " condition; the structure was pronouncedly dendritic but no veining was observed. Slow cooling from 1020° C. resulted in abundant veining.

General Effect of Alloy Additions on Veining.

It is worth recording that whereas the addition of a small quantity of an alloying element to copper facilitates the development of veining in suitably oxidized specimens, large additions have the reverse effect. The actual percentage of the added element required to show these effects is different for the different elements, and will presumably depend on (*a*) the range of composition of the particular α-solid solution; (*b*) the deoxidizing power of the added element, and (*c*) the solubility in the alloy of the oxide which is formed. Much care was required in preparing for micro-examination most of the alloys containing appreciable alloy additions. So pronounced is the difficulty of developing the veining structures in some alloys, that in a few, for example the 70 : 30, $(\beta + \gamma)$, and γ-brasses, and the higher silicon alloys in particular, it had almost been decided that no veining was present, but etching and light polishing were repeated for a large number of times, and eventually the structure was developed. Fortunately, when veining is present, it is often possible to detect some suggestion of the structure in comparatively early stages of preparation.

The present study has shown the definite lack of homogeneity of oxide content in the cored structures of the majority of the cast alloys. Although in a few instances the veining is evenly distributed throughout each crystal, the alloying addition usually tends to concentrate or even confine the veining in certain portions of the crystal presumably owing to differential solidification which leads to variations in concentration of the alloying element across the crystal. This in turn controls the distribution of the oxide, since the solubility of the oxide is affected by the concentration of the solid solution. In addition, where the alloying element is strongly deoxidizing, the deoxidation will be greatest where the alloying element is most concentrated. Thus, in the low alloy copper alloys, veining occurs mostly in the interdendritic areas, but it occurs in the dendritic areas in some of the high-alloy copper alloys and in intermediate positions in the intermediate types of alloys. The position of the veining in any one crystal is affected by two opposing influences : (*a*) the tendency for oxide-rich material to concentrate in the interdendritic spaces according to the natural law of differential solidification, and (*b*) the reduced solubility of the oxide in the alloy-rich portions which also occur in the interdendritic spaces owing to differential solidification.

Of the deoxidizing agents which have been considered, it is clear that phosphorus is the only one which when present in slight excess is capable of completely removing oxide from the metal, and this is presumably due to its being volatile at the temperature of the melt.

The facility with which veining may be produced in those alloys containing only a small alloying addition is at least partly due to the deoxidizing effect of the addition in removing excess oxide from the melt. Oxide contents greater than that required to saturate the α-solid solution at the temperature corresponding to its maximum oxide content result, during the final stages of solidification, in precipitation of comparatively coarse particles, usually in the form of a eutectic. Further precipitation of material during cooling in the solid state is liable to result in deposition on that already present, so that little evidence of veining may be shown.

(5) HEAT-TREATMENT EXPERIMENTS.

Effect of Annealing Temperature on Veining.

The effect of temperature was determined by annealing samples of the copper (LJY) and cupro–nickel (ECN) at temperatures from 250° to 1000° C. for 2 hrs. and then quenching. Similar effects were noticed in the two materials. Two-hour treatments caused little change in the appearance of the veining until 600° C. was reached, when it was apparent that solution and coalescence of the veining particles were taking place. These processes increased with increasing temperature until complete disappearance of the veining occurred. No veining could be detected after quenching from 900° C. in the case of the copper and 1000° C. in the case of the cupro–nickel. The disappearance of veining took place first in the dendrites, especially in the case of the cupro–nickel, with the result that the dendritic structure, arising from differential solidification, appeared to be more prominent after annealing at the intermediate temperatures than it was originally. The coalescence and partial solution of the veining constituent after treatment of the copper at 800° C. are shown in Fig. 18 (Plate III) and of the cupro–nickel at 750° C. in Fig. 19 (Plate IV). A comparison of Fig. 19 with Fig. 17 shows the intensifying effect of solution and partial precipitation on the casting structure. Heating to still higher temperatures caused the solution of the veining constituent from the interdendritic spaces, with the result that the dendritic structure again became less pronounced as it was originally. It was also noticed, especially in the case of the copper sample, that the veining persisted in some crystals more than others; after treatment at 800° C., for example, a few crystals appeared free from veining, whereas the veining in other crystals was still prominent although the type and quantity were different from the original structure.

The hardness of both materials was determined by the Vickers diamond pyramid after the different heat-treatments; the results are given in Table II.

TABLE II.—*Effect of Heat-Treatment on Hardness (Vickers Diamond Pyramid).*

Annealing and Quenching Temperature, ° C.	Hardness Value.	
	Copper (LJY).	Cupro-Nickel (ECN).
...	60	140
250	63	140
500	52	142
600	48	141
750	42	116
800	36	107
900	39	122
1000	40	127

These figures confirm the results of the microscopical examination: coalescence at the intermediate temperatures leads to loss of hardness, and solution at the higher temperatures has a hardening effect. The first two or three values for each material are high on account of a small amount of work-hardening to which the original materials had been subjected.

Effect of Time of Annealing.

The effect of prolonging the annealing was determined on fresh samples of the copper (LJY) and the cupro-nickel (ECN). One sample of each material was annealed for 20 hrs. at 400° C. and quenched, and another sample was given the same time at 750° C. and quenched. The treatments at 400° C. caused slight coalescence and possibly solution of the veining particles. A considerably greater change was naturally observed after the treatment at 750° C. Much solution and coalescence had occurred in both alloys, similar to that observed after 2 hrs. at 800° C. (cf. Fig. 18, Plate III; copper after 2 hrs. at 800° C.), although there was more coalescence after the treatment for 20 hrs.

The hardness values after annealing at 400° and 750° C. were 48 and 37, respectively, for the copper, and 130 and 111 for the cupro–nickel alloy. The temperatures of the 2-hr. treatments corresponding to these figures were approximately 600° and 800° C. in the case of the copper and 680° and 780° C. in the case of the cupro–nickel. The lower temperature is apparently more affected by the duration of the annealing, the greater hardness changes observed in each alloy indicating that equilibrium is less rapidly attained.

Release of Suppressed Veining.—I.

The temperatures favouring the solution of the veining constituent have already been considered, and it has been shown that when the metal is quenched at temperatures not too far below the melting point, veining is no longer to be observed. This " suppression " of veining by severe quenching is a characteristic property of the type of veining at present under discussion. It has also been shown that heating the metal to a high temperature and cooling at a suitably slow rate results in veining once more. In view of this, two questions may be asked: (1) what happens when a quenched specimen is reheated? and (2), during slow cooling at what temperatures is the veining formed?

The effect of re-heating a quenched specimen was determined on samples of the copper (LJY). Two specimens were first quenched in water from 1000° C. after 1 hr. at that temperature and the specimens annealed for 20 hrs., one at 400° C. and the other at 750° C., and then quenched. Pronounced veining was present in the specimen treated at 400° C., as shown in Fig. 20 (Plate IV), the meshes of the network being considerably smaller than in the original copper. Some difficulty was experienced in sharply defining many of the veining " lines," as if movement of the veining constituent had commenced in a direction away from the veins. The same specimen was re-quenched from 1000° C. and again annealed at 400° C., this time for only 2 hrs. The dimensions of the veining network were the same as after treatment for 20 hrs., but the veins were much more sharply defined.

The other specimen, quenched after 20 hrs. at 750° C., showed no veining but only a random precipitation (Fig. 21, Plate IV), although occasional examples suggestive of line formation of the deposit could be observed. (It was noticed in this case and also in one of the β-brass samples where random precipitation took place, for the " deposit " to appear very much like etching pits. A similar effect was observed in a few other specimens, especially when very deep etching had been employed, and has also been noted by Rhines and Mathewson [5] in their study of cuprous oxide in copper.) The specimen showed the same structure when given a further treatment by quenching from 1000° C. and heating for only 2 hrs. at 750° C. A third treatment was therefore given; the specimen was quenched from 1000° C., heated at 750° C. for only 10 minutes, and then quenched. Pronounced veining was now observed similar to that in the specimen given the treatment for 2 hrs. at 400° C. It is apparent, therefore, that the random deposit previously observed after the longer 750° C. treatments does not indicate the method of precipitation at that temperature but must be due to coalescence subsequent to precipitation. The experiment at 750° C. confirms in a remarkable manner the tendency towards coalescence observed at the treatment at 400° C.

Hardness Results.

The hardness of the specimens was determined at each stage using the Vickers diamond pyramid on the surfaces prepared for micro-examination; the results are given in Table III. For comparison there are also given the results of hardness determinations on two other samples of the copper, one slowly cooled from 1000° C. at the rate of $\frac{1}{4}$° per minute and referred to later, and the other a sample hydrogen treated in the melt and then annealed in hydrogen at 1000° C. and slowly cooled, and in which veining was completely absent. Thus, although the effect of the veining constituent is slight, the hardness of the copper is definitely influenced by the form in which the veining constituent occurs.

TABLE III.—*Hardness of Heat-Treated Copper (Vickers Diamond Pyramid).*

Heat-Treatment.	LJY 1.	LJY 2.
Quenched from 1000° C. . .	40	40
ditto + 20 hrs. at 400° . .	45	...
ditto + 20 hrs. at 750° . .	...	39
Quenched from 1000° C. . .	39	39
ditto + 2 hrs. at 400° . .	41	...
ditto + 2 hrs. at 750° . .	...	39
ditto + 10 minutes at 750° .	...	41
$\frac{1}{4}$° per minute, cooled from 1000° C.	36	
Hydrogen treated, and annealed .	30	

Release of Suppressed Veining.—II.

The determination of the temperatures at which veining is formed during cooling was made on two alloys : the 5 per cent. zinc–copper alloy and the copper–nickel alloy (LAQ) containing 20 per cent. nickel. Separate samples of each alloy were annealed at 1000° C., one pair quenched at that temperature and the remainder cooled at the rate of

2° per minute and quenched at intervals of 200° C. No veining was observed in the copper–zinc samples quenched at 1000° or 800° C. A coarse type of veining had commenced to form during cooling to 600° C. (Fig. 22, Plate IV) although some crystals showed very little of the structure. By the time 400° C. had been reached much veining was general throughout the specimen, as in Fig. 4 (Plate I), and little difference was observed between this and the one quenched at 200° C.

The effect of the treatments on the copper–nickel alloy was rather different. No pronounced veining could be observed in any of the samples, although a possible trace occurred in those quenched at 400° C. and below. There was, however, a slight random deposit in samples quenched below 800° C., and it was concluded that the cooling rate had not been the best to develop the network structure. The specimens were therefore re-treated, using a cooling rate about four times that previously adopted. The structures then corresponded with those of the 5 per cent. zinc–copper alloys at the different temperatures; a trace of veining was observed after quenching at 800° C., and the veining increased in quantity as the quenching temperature was lowered.

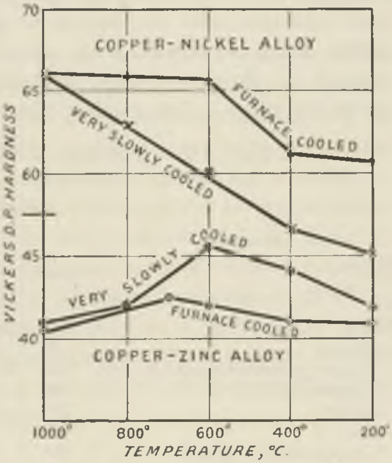

Fig. 37.

In view of this difference in structure, the samples of the 5 per cent. zinc–copper alloy were also subjected to the more rapid cooling. The only difference observed was that the veins in the more quickly cooled specimens were possibly slightly finer. The results of hardness tests are plotted in Fig. 37. The agglomerated precipitate in the copper–nickel alloy cooled at the lower rate resulted in a lower hardness figure than was obtained when the constituent occurred in a veining form. The copper–zinc alloy showed veining at both rates of cooling, and the higher hardness range of the very slowly cooled samples was presumably due to more complete precipitation.

Heat-Treatments in Hydrogen.

In a previous section it was mentioned that no veining was observed in samples prepared by bubbling hydrogen through the liquid metal and allowing to cool down in an atmosphere of hydrogen. The only exception to this was aluminium, whose oxide is known not to be reduced except at very high temperatures. Heat-treatments in hydrogen were, therefore, carried out on several metals and alloys in the solid state to determine whether and to what extent veining could be removed from materials in which it was originally present. Particulars of the treatments are summarized in Table IV.

Of the materials treated, only two types failed to show complete removal of veining from the surface of the specimen. In order to determine the depth of removal and to check that the heat-treatment alone was not responsible for the removal of veining, several specimens were sectioned after the treatment and the transverse faces examined. Before being sectioned, these specimens were given a thick electrodeposit of copper to enable the transverse faces to be examined right up to the edges. The general effect of the hydrogen treatment on the

TABLE IV.—*Hydrogen Treatments of the Solid Alloys.*

Specimen.	Treatment.		Veining Present?	
	Hours.	° C.	Before Treatment.	After Treatment.
Zinc	20	360–400	yes	no
Silver	3	900	yes	yes, as before
Copper (electrodeposited)	24	950–1000	no	no
Copper (LJY)	46	950	yes	no
Nickel	50	1000	yes	no
Iron	45	950	yes	no
Copper–zinc (5% Zn)	20	900	yes	no
ditto (10% Zn)	20	900	yes	no
Copper–tin (2% Sn)	20	900	yes	no
Copper–aluminium (2% Al)	3	1000	yes	yes, or before
Copper–silicon (0·5% Si)	23	1000	yes	yes, larger than before
The undermentioned specimens were plated and a transverse section examined after the hydrogen treatment.				
Copper (LJY)	22	900	yes	almost completely removed
Copper–zinc (5% Zn)	3	950	yes	veining removed from edge
ditto	22	900	yes	ditto
Copper–tin (2% Sn)	3	950	yes	ditto
ditto	22	900	yes	almost completely removed
Copper–nickel (70% Ni)	3	950	yes	veining removed from edge
ditto	20	950	yes	ditto
Copper–phosphorus (0·2% P)	3	950	yes	ditto
ditto	20	950	yes	ditto
Copper–aluminium (2% Al)	23	1000	yes	yes, as before
Copper–silicon (0·5% Si)	3	950	yes	ditto

structure was the same in the different specimens. Veining was present in the middle of the specimen, but there was an outer rim completely free from veining. The transition between the two fields occurred independently of the crystal boundaries. A typical structure is illustrated in Fig. 23 (Plate IV) which shows a portion of the transverse section of the low-tin bronze up to the edge. There was little difference in the thicknesses of the veining-free layers in different alloys given the same treatment, and the rim was, in general, only slightly thicker after the longer treatments. Unsoundness aided diffusion of the hydrogen, since in one or two bronze samples which showed pin-hole unsoundness "as cast," the removal of veining was almost complete throughout the specimen.

As mentioned above, there were two prominent exceptions to the rule by which veining is removed by hydrogen at high temperatures; a third exception, a type not included in the present section, will be referred to later. The first exception is offered by alloys containing either aluminium or silicon, and is apparently due to the inability of hydrogen to reduce aluminium or silicon oxide except at temperatures considerably in excess of those used in the present work. In the alloy of copper containing 0·5 per cent. silicon, for example, veining persisted even on the surface after a treatment of 23 hrs. in hydrogen at 1000° C.; the veining in this copper-silicon sample was on a much larger scale in most of the crystals than it was originally. Presumably alloys of other materials containing oxides not reduced by hydrogen will also be unaffected by the hydrogen treatment.

Veining in Silver.

The second exception is offered by pure silver. In this, the original veining had the appearance of a succession of cavities as due to gas (Fig. 8, Plate II), and no change was observed as a result of the hydrogen treatment. It was mentioned earlier in the paper that silver represents a special type, since its oxide decomposes at relatively low temperatures. The solubility of oxygen in molten silver has been measured by Sieverts and Hagenacker [7] and in solid silver by Steacie and Johnson.[8] It was shown that whereas silver in the liquid state (973° C.) can dissolve twenty times its volume of oxygen, the solubility in the solid state is considerably less, being 0·56 at 923° C., and decreases to a minimum of 0·087 at 400° C. Steacie and Johnson [9] consider this minimum to be due to the state in which the oxygen exists in solution, and point out that this minimum corresponds with the minimum in the heat of formation of silver oxide calculated by Keyes and Hara [10] from measurements of the dissociation pressure of oxygen in equilibrium with silver oxide.

The large difference between the oxygen solubilities of liquid and solid silver is responsible for the well-known " spitting " phenomenon which is liable to occur during the solidification of silver. Sieverts and Hagenacker also calculated the changes in internal gas pressure during the solidification of silver saturated with air at 1030° C., assuming that no gas escaped, and concluded that values well exceeding 200 atm. would be obtained. Normally, of course, some gas escapes, and such a value is not likely to be reached. Partial solidification hindering the escape of gas would lead, however, to high values of internal pressure. It was calculated that under atmospheric pressure oxygen should not be liberated from the melt before 59 per cent. of the metal had solidified. This figure agrees very well with the 63 per cent. found experimentally by Allen.[11] Gas evolution then gives rise to " spitting." The porosity due to this evolution which occurs during solidification is intercrystalline or interdendritic in character and is shown as large black areas in the photomicrograph of silver, Fig. 8 (Plate II). From their position the minute cavities indicative of veining must have occurred, however, subsequent to solidification. Although the gas pressure necessary for the formation of a bubble in solid silver will be considerably greater than that when the silver is in the liquid condition, the high internal gas pressures which can be realised and the mechanical weakness of the silver at temperatures not far removed from its melting point render this not impossible. The conditions for the appearance of gas blisters in solid metals have been considered by Allen.[11] Alternatively, partial combination between the silver and oxygen may have taken place. Allen has shown that the compound can exist at temperatures as high as 600° C. when the external pressure is sufficient to prevent dissociation. It may be concluded, however, that whether its veining constituent was initially a gas or whether it was a solid which later dissociated, silver can undoubtedly be classed amongst those metals which are capable of showing veining structures.

Quenching Experiments.

It has been shown that by increasing the temperature, the veining structure of several alloys may be made to disappear. Similar intermediate quenching experiments have not been carried out for all the other alloys in which veining was observed, but check experiments have been made by quenching several of them at temperatures not far removed from their respective solidus temperatures (except iron, which was quenched well above the upper critical temperature). All samples were maintained at the quenching temperature for not less than 1 hr. The particulars of these experiments are summarized in Table V.

TABLE V.—*Quenching Experiments.*

Specimen.	Quenching Temperature, ° C.	Veining Present?	
		Before Treatment.	After Treatment.
Zinc	400	yes	no
Copper	1000	yes	no
Iron	1050	yes	no
Copper–tin (2% Sn) . .	800	yes	no
Copper–zinc (5% Zn) . .	950	yes	no
Copper–phosphorus (0·2% P)	950	yes	no
Copper–silicon (0·5% Si) .	1000	yes	no
Copper–aluminium (2% Al) .	1020	yes	no
Copper–nickel (70% Ni) .	1020	yes	no

Thus, in every instance veining may be suppressed by quenching the solid alloy at a sufficiently high temperature; it may be concluded, therefore, that in these alloys also the veining constituent is more soluble at the higher temperatures. Subsequently, slow cooling resulted, once more, in the formation of veining.

Effect of Very Slow Cooling.

When veining was developed by oxidizing the melt and slowly cooling in the melting furnace, the rate of cooling of the copper alloys was of the order of 6°–2° C. per minute between the solidifying temperature and 400° C., and this rate was adopted in most of the heat-treatment experiments except when the specimens were quenched.

A sample of copper (LJY) and of fairly pure iron were annealed for 20 minutes at 1000° C. and then cooled at the rate of only $\frac{1}{4}$° per minute. Typical veining occurred in both samples before the heat-treatment, but only in the iron sample afterwards. There was a random distribution of particles in the copper sample and practically no trace of a network structure. The appearance was similar to that of the copper annealed for several hours at 750° C. and quenched (Fig. 21, Plate IV), as described previously, and was undoubtedly due to the annealing effect exerted by the extremely slow cooling. In the iron sample the veining constituent showed some evidence of coalescence, but the metal was presumably too rigid to permit the degree of movement found in the copper.

A nickel–copper alloy containing 5 per cent. nickel, when cooled in the furnace at the rate of 1°–2° per minute from 1000° C. also showed a random distribution; cooling the same sample in air caused some alignment of the particles into the form of a network. Similar results were obtained with the higher phosphorus alloys.

Heat-Treatment of β-Brass.

The first β-brass (copper 52, zinc 48 per cent.) prepared showed a random distribution of particles, and only slight evidence of a network occurred when the melt was air-cooled. Cooling small samples in a tube furnace from about 820° C. at different rates failed to develop veining, but a tendency towards line formation was observed after air-cooling (Fig. 12, Plate II). An increase in the proportion of veining was obtained by cooling in an air blast or better still by cooling in contact with two cold copper blocks.

Effect of Annealing.

A number of the alloys in which the dendritic structure was pronounced were annealed for short periods at high temperatures and cooled in the furnace to remove the "coring" and to determine the influence on veining. Veining was more easily developed after annealing. In the 10 per cent. tin–bronze annealed for 3 hrs. at 800° C., for example, there was abundant veining, which was more pronounced in the dendritic axes than elsewhere (Fig. 24, Plate IV), and this resulted in the dendritic structure being actually more pronounced after annealing than before. The dendritic axes of the cored α-solid solution in the copper–tin alloys are normally etched more rapidly than the tin-rich interdendritic material, and this differential etching effect was accentuated by the duplex character of the dendrites in the annealed material owing to the separation of the veining constituent.

(6) THE EFFECT OF WORK AND RECRYSTALLIZATION ON VEINING.

Cold-Work.

Samples of a number of materials were given various degrees of cold-rolling, and sections were examined to determine whether any relationship existed between veining and strain marks. In addition, small flat tensile test-pieces were prepared from a few of the alloys and various degrees of strain applied. The test-pieces were polished and etched before straining. In most instances the slip-lines were not affected by veining, but examples were observed where a slip-line appeared to be arrested at the veins or where a small kink occurred in the slip-line as it crossed a vein. Such a kink or slight directional change may have been due, however, to the presence of a "furrow" or slight difference in level of the surface at the position of the veining line, and the only conclusion which could be reached from these experiments was that if the veining had any influence at all on the slip-lines, there was very slight evidence of it at the low magnifications at which the structures were most prominent in the samples examined.

Recrystallization Experiments.

Low-Temperature Recrystallization.—Recrystallization of a metal which does not undergo an allotropic change may be achieved either by hot-work or by cold-work followed by annealing. Control over recrystallization may be obtained by adjusting the degree of cold-work and the temperature to which the material is heated. Annealing an unworked specimen of the copper (LJY) for a short time at 500° C. has been shown to have no observable effect on the veining. In the present experiments the thickness of the specimen was reduced 10 per cent. by cold-rolling, and this was followed by annealing at 500° C. for 20 minutes. This treatment resulted in slight grain-growth, which took place by an undulating movement of the boundaries of one crystal into its neighbour. A typical example, Fig. 25 (Plate V), shows the old boundary, demarcated by isolated globules of copper oxide and by the abrupt ending of the veins, and also the new boundary of which the wavy form characterizes grain-growth as opposed to recrystallization, which is usually shown by straight or step-like boundaries of twins or new crystals. The lightly etched crystal in the photograph is extending into the dark crystal; scratch tests confirmed the continuity of the light crystal. The veining present in both crystals remained in the position in which it occurred before annealing, with the result that the new grain-boundary cuts across several of the veins.

High-Temperature Recrystallization.—The effect of recrystallization at high temperatures was determined on another sample of the same copper. Solution of the veining constituent has previously been shown to occur in this copper at temperatures in excess of 900° C., and veining may be produced again by slow cooling. A fresh sample was, therefore, reduced in thickness 20 per cent. by cold-rolling, annealed at 1000° C. for 1 hr., by which time it was considered recrystallization would be reasonably stabilized, and then slowly cooled at the rate of 2° per minute. Complete recrystallization had taken place and the high annealing temperature had resulted in fair-sized crystals about $\frac{1}{8}$ in. or more in width. Examination for veining showed that each of the new crystals had its own system of veining. This is well illustrated in Fig. 26 (Plate V), from which it will be seen that the networks in the different crystals are independent and do not extend beyond the boundaries of their own crystals. Apart from the twinning which signifies this type of recrystallization, the structure is similar in all respects to that found in the original copper slowly cooled from the melt. Similar results were obtained with many of the copper alloys.

Intermediate Stages in Recrystallization.—A tapered sample of the 5 per cent. zinc–copper alloy, in which there was abundant veining, was prepared and cold-rolled until it was flat, so that a specimen was obtained in which the internal strain varied over a wide range in different parts of the specimen. Pieces cut from this were then annealed at from 300° to 600° C. As would be expected, the crystal structure varied across the specimens, the portion which had had most cold reduction consisted of wholly recrystallized material, and that which had had least work was unchanged.

At first, especially when examined at low magnification, there appeared to be no veining in the recrystallized areas although veining was still abundant in the unrecrystallized material, but further preparation and examination at higher powers showed remains of veining principally as isolated particles more or less in line formation as if due either to coalescence or to the solution of the smaller particles constituting the veins. The effect on the sample annealed at 500° C. for 1 hr. is shown in Fig. 27 (Plate V). Another feature which was observed was the tendency for a twinning plane in the new crystals to spread from the position where a veining line touched the crystal. It is not possible to say whether this has any significance or whether it is merely a coincidence, but many examples were noticed.

The structure of the sample annealed at 450° C. changed gradually from the fully recrystallized at one end to the unchanged original crystals at the other, with mixtures of the two structures in varying proportions at intermediate positions. It appeared that recrystallization first took place in the interdendritic areas as shown in Fig. 28 (Plate V). This was unexpected as the impurities, including oxide, are concentrated in such areas, and impurities are known to increase the recrystallization temperature. The crystals were too small to permit sampling for analysis from the dendritic and interdendritic areas as has been done with large steel crystals,[12] but the quantity of impurities (excluding zinc and oxide) in the interdendritic areas must be extremely small since the alloy was prepared from electrolytic copper and 99·99+ per cent. zinc. The following explanation is put forward. Since the adjacent dendritic and interdendritic areas may be assumed to have undergone the same degree of rolling, the less pure interdendritic material will have work-hardened slightly more than the purer dendritic areas, and this increased work-hardening had the usual effect of reducing the recrystallization temperature. In order to check this view, Vickers diamond pyramid hardness tests were carried out on the different areas using a load of 5 kg. to obtain a small impression, and the results are given in Table VI. Most of the impressions were made on two adjacent crystals (*a* and *b*) which were conveniently each nearly 1 cm. in length and in which slight recrystallization had occurred at one end only.

TABLE VI.—*Variation of Hardness within the Crystals.*

Crystal Examined.	Position in the Crystal.	Vickers Diamond Pyramid Hardness.
Crystal *a*, unrecrystallized	dendritic areas	94·6, 96·5, 94·0
	interdendritic areas	99·0, 100·0
Crystal *a*, partial recrystallization	interdendritic areas, slight recrystallization increasing in quantity with decreasing hardness	77·5, 73·2, 68·8
Crystal *b*, unrecrystallized	dendritic areas	91·7, 93·4, 93·4
	interdendritic areas	98·4, 99·7, 100·0
Crystal *b*, partial recrystallization	as in *a* above	82·6, 71·2
Fully recrystallized area at end of specimen, hardness varying with size of new crystals.		49·7–53·3

The hardness of the alloy is thus different in different parts of the same crystal, and the initiation of recrystallization in the interdendritic areas would appear to be due to their increased work-hardening. The true hardness differences are most probably greater than the figures indicate since some overlapping of the impression with the different areas occurred. The effect of such overlapping is seen in the hardness determination made on the partially recrystallized areas, the higher values corresponding with overlapping of the impression beyond the new crystals.

(7) SUB-BOUNDARY STRUCTURES DUE TO AN ALLOYING CONSTITUENT.

During the examination of a wide variety of alloys, structures have occasionally been observed which show some resemblance to veining but which have obviously not been due to the precipitation of oxides. It is not proposed to consider this type of structure in detail, but a few examples are given to show the difference between this structure and veining. Simply to distinguish between the two structures in the present discussion, the one not due to oxides will be termed the sub-boundary structure as it appeared less characteristic of “ veins.” The origin of both structures is the same, namely, the precipitation from solution, during cooling, of a substance which is less soluble at lower temperatures. It has been found that the network structure is likely to be more pronounced when the composition is such that there is no excess of the solute at elevated temperatures in the solid state.

An example of a network structure due to an intentional constituent of the alloy has already been mentioned in the discussion of the copper–phosphorus alloys. Veining was observed in the alloy containing a small quantity (0·2 per cent.) of phosphorus. With larger additions, however, the veining was displaced by a similar structure (Fig. 29, Plate V) which was associated with free phosphide and which evidently resulted from precipitation of copper phosphide from solid solution.

Several examples of sub-boundary structures have been observed in alloy steels. One type is represented by steels in which a closed γ-loop is formed. When the composition is such that the alloy during cooling does not cross this loop, that is it remains in the α state throughout, sub-boundary structures may be observed if the alloying constituent forms a compound with iron in which its solid solubility decreases with decrease in temperature, and this solubility boundary is crossed by the alloy in cooling to atmospheric temperature. Examples of this type may be found in the alloys of iron with vanadium, molybdenum, tungsten, or chromium. The precipitation of the iron–molybdenum compound in line or network form is shown in Fig. 30 (Plate V). In this type of alloy, each crystal contained its own system of sub-boundaries, *i.e.* the network did not cross the crystal boundaries. There are other alloys steels, however, in which it is possible to select a composition where the sub-boundary constituent is thrown out of solution in a network form from the γ-modification of iron, which subsequently undergoes the allotropic change to α-iron at a lower temperature. In this instance, the sub-boundary network is independent of the finally occurring α-iron crystal boundaries. A good example of this is shown in Fig. 31 (Plate VI) of the 2·2 per cent. tantalum–iron alloy, in which the compound Fe_2Ta starts to separate at a temperature nearly 200° C. above the γ–α change point. This particular sample had been annealed in hydrogen at 1220° C. for 1 hr. and cooled in the furnace.

An interesting example of sub-boundary structures in a non-ferrous alloy has been observed in a series of complex aluminium–" bronzes " containing appreciable quantities of nickel and iron. Under certain conditions of cooling from a high temperature, a compound which is probably rich in iron separates from the β-solid solution in a network form as shown in Fig. 32 (Plate VI); the coarse black mark is a quenching crack pointing along the crystal boundary, and it will be seen that each crystal has its own network system. With a higher aluminium content, the stable phase at high temperatures is the γ-solid solution which undergoes a change to δ-solid solution on cooling, but the temperature at which this change occurs is below that at which the constituent is deposited from solution. The sub-boundary constituent remains undisturbed by the phase change as is shown by Fig. 33 (Plate VI), which illustrates an intermediate stage in the breakdown of $\gamma \longrightarrow \delta$ with the " lines " of the network continued across the γ–δ boundary.

Another alloy in which sub-boundary structures may be developed is that of copper containing a little over 2 per cent. beryllium. Cooling in the furnace from 820° C. resulted in isolated deposits of the γ-constituent. By cooling in air small samples about $\frac{1}{2}$ in. × 1 in. square, the second phase occurred in the form of a small network as shown in Figs. 34 and 35 (Plate VI).

Attempts to obtain similar structures by tempering quenched samples were not successful. Specimens tempered for 1 hr. at 200°–600° C. all showed a " martensitic " structure, although at 600° C. there was evidence of some coalescence, and this was more pronounced after tempering at 700° C. when partial re-solution had occurred. In order to examine the structure of the network at an early stage in its formation, samples were quenched at different intervals during air-cooling. There were only isolated deposits in the sample quenched after 25 seconds, but samples cooled for longer periods showed a network structure which

TABLE VII.—*Effect of Heat-Treatment on Structure and Hardness of* 2·2 *Per Cent. Beryllium–Copper Alloy* (*Vickers Diamond Pyramid*).

Heat-Treatment.	Structure.	Hardness.
Forged and fully heat-treated, as received	$\alpha + (\alpha + \gamma) +$ martensitic structure in α	406
Furnace-cooled from 800° C. at 1° per minute	$\alpha + (\alpha + \gamma) +$ random distribution of γ	144
Air-cooled from 800° C. in about 12 minutes	$\alpha + (\alpha + \gamma) +$ network of γ in α	178
Air-cooled from 820° C. for 25 seconds and quenched	$\alpha + (\alpha + \gamma) +$ isolated needles of γ forming a broken network	113
Air-cooled from 820° C. for $1\frac{1}{2}$ minutes and quenched	$\alpha + (\alpha + \gamma) +$ small fine network of γ in α	142
Air-cooled from 820° C. for 3 minutes and quenched	ditto, slightly thicker network and some isolated particles	164
Air-cooled from 820° C. for 5 minutes and quenched	ditto, network coarser and more isolated particles	184
Quenched after 1 hr. at 840° C.	$\alpha + \beta$	118
Ditto and tempered for 1 hr. at 200° C. and quenched	$\alpha + \beta$ (decomposing) + faint martensitic structure in α	164
Quenched and tempered at 300° C.	ditto, martensitic structure more pronounced	342
Quenched and tempered at 400° C.	ditto	389
Quenched and tempered at 500° C.	ditto	234
Quenched and tempered at 600° C.	ditto	166
Quenched and tempered at 700° C.	ditto + isolated particles of β	128
Quenched and tempered at 800° C.	$\alpha + \beta$	120

was finest in that quenched at a barely visible red heat after cooling for 1½ minutes in air; this structure is shown in Fig. 36 (Plate VI). In order to correlate structure with mechanical properties, hardness determinations were made on the different samples; the results are summarized in Table VII.

The " martensitic " structure thus covers a wide range of hardness, and although there were certain differences in this structure in the samples of different hardness, this aspect was not explored. The network structure corresponded with only moderate hardness values, but such samples were slightly, though definitely, harder than those in which the second phase showed a random distribution.

The main difference between veining and sub-boundary structures as considered in the present paper is that whereas veining is normally removed by a long hydrogen treatment at high temperatures, the sub-boundary structure remains unaffected. The sub-boundary structures also usually consist more obviously of discreet particles; the individual particles constituting the veining structure, on the other hand, are usually only resolved under quite high magnifications. A third difference, only occasionally noticed, is that whereas there is no great difficulty in suppressing veining by quenching from high temperatures, in a few of the alloy steels for example, the sub-boundary structure tended to persist after quenching, apparently on account of the inability of the solid solution to remain highly supersaturated.

(8) DISCUSSION OF RESULTS.

Before considering the significance of veining in metals, the results obtained will be summarized and an explanation offered as to the cause of veining. In the first instance the discussion will not include those structures of the type referred to as sub-boundaries which represent an intentional constituent of the alloy.

Veining has been shown to be developed in several pure metals and alloys as a result of slow cooling after the metal or alloy has undergone an oxidizing treatment, either by deliberately oxidizing the melt or by annealing the solid at a high temperature in air or in the presence of its oxide. Conversely, treating the liquid metal with hydrogen or annealing the solid metal at high temperature for a sufficient length of time in an atmosphere of hydrogen results in a material free from veining,* although veining has been observed under similar conditions

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* One exception to this action is given by aluminium as mentioned on p. 307. The temperature of the experiment was well below the temperature above which Al_2O_3 is reduced by hydrogen. No reducing experiments were made with magnesium which is known to form a chemical compound with hydrogen.

of treatment when an oxidizing atmosphere has been employed. This type of veining is intimately associated therefore with the oxygen or oxide content of the material. Secondly, microscopical examination of the veins at a sufficiently high magnification shows that the veining lines are discontinuous and appear to be made up of a series of small particles of a constituent, which, from the previous conclusion, must consist of an oxide of the metal concerned. Thirdly, since severe quenching from a temperature slightly below the melting point prevents the formation of veining, whereas slow cooling develops it, the solubility of the veining constituent in the metal is evidently greater at elevated than at atmospheric temperatures.

The remaining results, namely, (*a*) the effects of low-temperature recrystallization where the veining is left undisturbed by the formation of new crystals (or the growth of the original crystals) at temperatures below the solution temperature of the veining constituent; (*b*) the formation of a fresh system of veining with the recrystallization and growth of new crystals during the high-temperature annealing after cold-work; (*c*) the ability to differentiate between veining and strain-lines after cold-work; (*d*) the action of a powerful deoxidizing agent such as phosphorus in reducing or even eliminating veining; (*e*) the effect of the form in which the veining constituent is present on the hardness of the metal; (*f*) the coalescence of the particles during long annealing at intermediate temperatures; and (*g*) the effect of work in aiding the solution of the veining constituent at intermediate temperatures; these results are all consistent with the view that veining is due to the presence of metallic oxides whose solubility in the metal increases with increase in temperature.

One of the results shown by the present work is that many metals dissolve, especially at high temperatures, small quantities of their own oxides. The action is not limited to pure metals but also occurs with many alloys. Further, although the addition of a deoxidizing agent changes the composition of the oxide, it does not necessarily completely remove the oxide from the alloy. The beneficial influence of deoxidizing agents, therefore, depends on the properties of the oxide formed, and particularly of that in excess of the quantity which can exist in solid solution. It is also evident, from the small quantity of the veining constituent present and from the chemical purity of the metal even when much veining is shown, that the quantitative solubility is very small indeed.

It is of interest to consider the position of veining structures relative to twinning. Twinning may be classed either as stress twinning which is due to release of stress at atmospheric or moderate temperatures, or annealing twinning, which, although also associated with stress, does not develop until the temperature is increased sufficiently to permit recrystallization and, commonly, grain-growth. In the case of stress twinning, the veining as would be expected, passes through the twinned crystals unchanged. On the other hand, twin crystals formed by annealing at high temperatures were observed more often than not to have their own system of veining, although examples were observed where a vein crossed the twin plane. A study of the relationship between veining and twins would probably yield some information as to the temperature of twin formation and growth.

The results of the present investigation confirm those of the previous investigation on veining in iron [1] in which it was concluded that veining was due to the precipitation of iron oxide during cooling and that the solid solubility of the oxide in iron decreases with decrease in temperature. The results also explain the irregularities found in single crystals of copper by W. H. George,[13] who stated that the appearance of the structure (the same as that described in the present paper) suggested that the crystal had broken down into a large number of small ones.

It may be enquired why it is that although the majority of alloy systems contain one or more phase boundaries representing a decreased solubility of a constituent with decrease in temperature, sub-boundary structures are not more frequently observed. There are probably two reasons for this. First, it appears to be necessary to have only a certain small critical amount of the sub-boundary constituent. If much more is present there is too strong a tendency towards either a fine shower precipitate or else a massive one where the constituent is deposited on the initial particles as mentioned previously. Secondly, the rate of cooling is of some importance. In the present work, network structures have been most prominent when cooling has been slow, although an excessively slow cooling may result in too much coalescence to show any network. The optimum cooling rates differ for the different alloys, and it has been shown that whereas furnace cooling which corresponds to a rate of from 6° to 2° C. per minute is satisfactory for many alloys, others such as the β-brass and beryllium–copper require to be air-cooled at as high a rate as 300°–100° C. per minute in order to prevent coalescence.

Although the conditions favouring the precipitation of oxides or other compounds in a network form have been considered, no reason has been adduced as to *why* the deposits should assume a network form. This aspect of the subject and the possible relation between veining and mosaic structures of metals are discussed elsewhere.

SUMMARY.

(1) A network structure, termed veining, which occurs within the crystals has been studied in several pure metals and in many alloys of copper.

(2) Veining has been shown to be due to the precipitation from solid solution, in a network form, of the oxide of the metal during cooling.

(3) Many metals in the solid state can dissolve small quantities of their own oxides.

(4) The solid solubility of a metallic oxide in the metal is very small and decreases with decrease in temperature.

(5) Alloys are also capable of dissolving only small quantities of an oxide; the oxide which is formed depends on the character of the added element.

(6) Structures somewhat similar in appearance to veining and referred to as sub-boundary structures have been observed in a number of alloys, and are due to the precipitation during cooling, not of oxide, but of a compound of the metals concerned.

(7) Veining can be removed by a hydrogen treatment at high temperatures, in the solid or in the liquid state, provided that the oxide which is responsible for the veining can be reduced by hydrogen.

(8) The formation of veining structures can be suppressed by quenching the alloy from a sufficiently high temperature in the solid state.

(9) Veining can be released in a quenched specimen by slow cooling from the quenching temperature or by annealing at a suitable intermediate temperature.

(10) The addition of a small quantity of an alloying element of the deoxidizing type to copper was found to facilitate the development of veining. Large additions had the reverse effect.

(11) Solution of the veining constituent in material recrystallized at intermediate temperatures is aided by preliminary deformation.

(12) When an alloy having a pronounced dendritic structure is rolled and then annealed, recrystallization is initiated in the interdendritic areas.

(13) The hardness of the alloy has been shown to be affected by the form in which the oxide occurs, a network form corresponding with higher hardness.

ACKNOWLEDGMENTS.

The author wishes to acknowledge the help of Mr. H. Wrighton, B.Met., F.R.M.S., who was responsible for the photographic work.

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APPENDIX.

PREPARATION OF SPECIMENS FOR MICROEXAMINATION.

Polishing.

Standard methods of metallographic polishing have been followed, a liquid metal polish being used on a soft cloth for the final hand polishing. Repeated polishing and etching have nearly always been necessary, as the presence of many scratches seriously interferes with the development of veining.

Etching Reagents.

As some etching reagents develop the veining structure more easily than others, those found most suitable are indicated below. Except with certain "easy" alloys, prolonged etching is usually necessary, and re-polishing with metal polish must be resorted to as soon as the etching-pit structure is developed.

Copper.—10 per cent. ammonium persulphate in 50 per cent. ammonium hydroxide very satisfactory. Diluting the solution reduces the tendency towards etching-pit formation but the veining structure is then left in poor contrast. Ferric chloride (10 grm. $FeCl_3$, 30 c.c. hydrochloric acid, 120 c.c. water) gives poor results and veining is not developed by nitric acid, caustic potash, or ammonia solution.

Cadmium.—1 per cent. nitric acid in alcohol followed by ammoniacal ammonium persulphate followed by a light etch with dilute alcoholic ferric chloride to clear the surface. This triple etching proved more satisfactory that a single etch.

Aluminium.—Dilute hydrofluoric acid in water.

Zinc.—10 per cent. sulphuric acid containing one or two crystals of chromic acid is very good; 5 per cent. nitric acid in alcohol is fairly good.

Silver.—10 per cent. sulphuric acid containing one or two crystals of chromic acid is good; nitric acid and ammonium persulphate are unsatisfactory.

Magnesium.—5 per cent. nitric acid in alcohol.

Antimony.—30 per cent. hydrochloric acid containing about 5 per cent. hydrogen peroxide.

Nickel.—Aqua regia.

Iron.—5 per cent. nitric acid in alcohol.

Mild Steel.—Alternate etching with ½ strength Stead's reagent and picric acid.

Stainless Steels.—10 per cent. ammonium persulphate in 10 per cent. hydrochloric acid. The acid concentration can be increased if the action is found too slow.

Alloy α-Steels.—Prolonged etching in 5 per cent. nitric acid in alcohol.

High-Nickel Copper–Nickel Alloys.—Ferric chloride as described for copper.

Low-Nickel Copper–Nickel Alloys.—10 per cent. ammonium persulphate in 20 per cent. ammonium hydroxide or ferric chloride.

Copper–Zinc Alloys.—Ammonium persulphate solution as for copper was found most satisfactory. The ferric chloride solution required prolonged etching and developed the dendritic structure excessively.

Copper–Tin Alloys.—Ammonium persulphate as above. For the high-tin alloys, acid persulphate solution (10 per cent. ammonium persulphate in 30 per cent. sulphuric acid) is preferable.

Copper–Aluminium Alloys.—Prolonged etching with the acid persulphate solution.

Copper–Silicon.—Ammoniacal ammonium persulphate as for copper–zinc, followed by washing in very dilute hydrofluoric acid solution. Acid persulphate is satisfactory for alloys containing up to about 1 per cent. silicon.

Copper–Beryllium.—Dilute alcoholic ferric chloride solution or 5 per cent. alcoholic nitric acid.

Copper–Phosphorus and Copper–Arsenic.—Ammoniacal persulphate followed by a light etch with acid persulphate to clear. Acid persulphate solution slightly less satisfactory.

False Networks.

It may be as well to point out that examples have been observed of a network structure due to cracks in a surface oxide film formed during or after etching with ammoniacal persulphate solution. Some alloys, *e.g.* the alloys of copper containing aluminium, silicon, or tin, are more prone than others to film formation. The film is easily removed by washing with dilute hydrofluoric acid or a strongly acid solution of persulphate. A simple test is to examine the specimen at low magnification at the same time trying to remove the " network " with a needle or sharp knife; a film is easily detected.

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MECHANICAL PROPERTIES OF ALUMINIUM AND ITS ALLOYS AFTER PROLONGED HEATING.*

By Professor Dr.-Ing. A. von ZEERLEDER,† Member, and Dr.-Ing. R. IRMANN,‡ Met.Eng.

Synopsis.

The strength of pure aluminium and various aluminium alloys has been investigated after heating for long periods at 75°–300° C. The strength properties of the same materials were determined also in the state of complete stabilization. The heating periods employed extended over approximately 2 years. After treatments of different duration at elevated temperatures, some specimens were cooled and tested at room temperature; others were tested at the treatment temperature. For complete softening at 250° C., for instance, *i.e.* to obtain the properties observed in completely stabilized samples, heating periods of at least 6 months are required in the case of pure aluminium and Anticorodal, and of more than 2 years for Avional and " Y " alloy. A still longer heating time is necessary in the case of the alloy " R.R. 59."

To determine the decrease in strength of any material after heating for long periods at elevated temperatures it is necessary, therefore, to carry out tests of long duration.

The values of the yield-point observed in the normal short time test after heating periods of 1 year form a basis for calculations for engineers. The permissible loads can be ascertained only by observation of the creep limit, and this property is being studied by the authors.

Introduction.

The increasing use of aluminium and its alloys at elevated temperatures makes a precise knowledge of the properties at those temperatures of increasing importance. The present paper describes extensive investigations on pure aluminium and some of its alloys in this connection.

On consideration of the influence of temperature on the tensile strength of aluminium (Fig. 1) as observed by different investigators, serious discrepancies are found in the results, especially for temperatures between 100° and 300° C.—a range of special importance to the engineer. In view of the experience of the various investigators, the discrepancies cannot be attributable to errors of observation, but rather to differences in the test conditions employed.

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† Chief Engineer, Aluminium-Industrie A.G., Neuhausen, Switzerland; Professor of Technology of Aluminium at Federal Technical High School, Zürich.

‡ Head of the Mechanical Testing Department, Aluminium-Industrie A.G., Neuhausen, Switzerland.

Note to Abstractors and Other Readers.—This paper will be published, in permanent form, in the *Journal of the Institute of Metals*, Vol. LIX, 1936. Reference should accordingly be as follows: *J. Inst. Metals*, 1936, **59** (Advance copy).

As will be shown later, a slow, gradually increasing decrease in strength occurs on long heating within the temperature range mentioned above. The decrease in strength properties depends on the nature of the alloy. Anticorodal, for instance, softens within a comparatively short time, whereas Avional needs much more time. This is made clear in Fig. 2, which shows the time necessary at different temperatures to cause a 50 per cent. decrease in hardness. The heating time is represented logarithmically in this and the other illustrations.

Fig. 1.—Tensile Strength of Pure Aluminium as Depending on the Temperature. Tested at the Annealing Temperature.

Systematic investigations of these effects have been carried out during the past 15 years, but appropriate and reliable technique has been developed only recently. At the Autumn Meeting of the Institute in Düsseldorf in 1929, the first paper on the influence of operating temperatures on the strength properties of overhead electric transmission cables was read by one of the present authors.[1] It was stated in that paper that a hard wire of pure aluminium softens noticeably at a temperature of 80° C., and it was further stated in error that complete softening had occurred after heating at 80° C. for 1000 hrs. More accurate tests in a subsequent investigation have shown that statement to be incorrect.[2] The results of the subsequent investigation, which included tests of wires in different materials after heating for long periods at 77° C. are given in Fig. 3, and those obtained after heating at 160° C. are given in Fig. 4. These results show that the percentage decrease of strength after heating at 77° C. is much greater for copper than for pure aluminium and Aldrey, while Anticorodal and Avional show a noticeable age-hardening effect. After heating at 160° C. the latter alloys showed a distinct decrease of strength, but after heating at 100° C. no softening was observed in these materials.

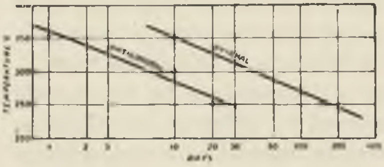

Fig. 2.—Time Necessary for a 50 Per Cent. Decrease in Hardness of Anticorodal and Avional after Heat-Treatment at Different Temperatures. Tested at 20° C.

EXPERIMENTS.

Additional experiments have been made to determine the temperature at which the softening commences. Heating at temperatures between 77° and 160° C. was carried out in an oil-bath. The strength of the samples was determined after they had been heated for suitable periods at different temperatures, and cooled to 20° C. The time required for completion of the tensile test varied between 2 and 5 minutes, according to the elongation of the material.

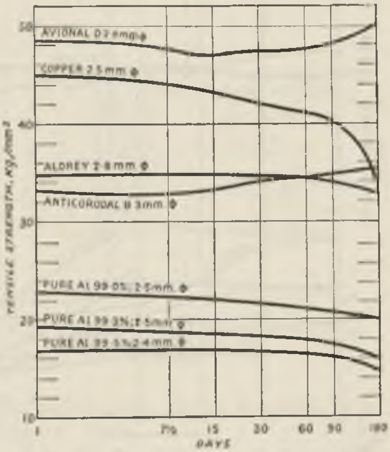

FIG. 3.—Influence of Continuous Heating at 77° C. on the Tensile Strength of Various Wire Materials. Tested at 20° C.

Tests of wires of pure aluminium of 99·0 and 99·5 per cent. purity, Anticorodal, Avional D, Aldrey, copper, nickel, and iron, at 20° after heating for 1 year at 50°, 75°, 100°, and 160° C. confirm the earlier results. The tensile strength curves are given in Fig. 5. In Fig. 6 the values are given as percentages of the original strength, and it is quite clear that copper, after heating for 1 year at temperatures up to 160° C., shows a much higher percentage decrease in strength than pure aluminium.

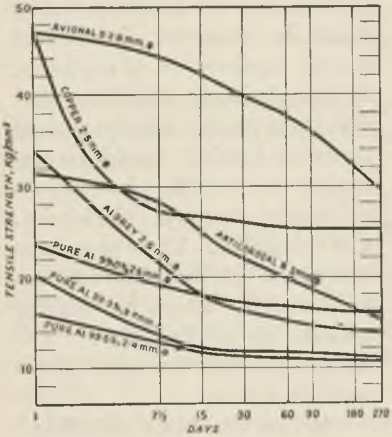

FIG. 4.—Influence of Continuous Heating at 160° C. on the Tensile Strength of Various Wire Materials. Tested at 20° C.

In previous work[2] a study was made of the influence of complete and incomplete stabilization on the strength properties of heat-treated aluminium alloys. The material was in the form of extruded bars of 20 mm. diameter.

The term "incomplete stabilization" is used in reference to samples which, before being tested, had first been given normal solution-treatment, then heated for 2 hrs. at 100°, 200°, 250°, 300°, 350°, 400°, or 450° C., and finally tested at that temperature.

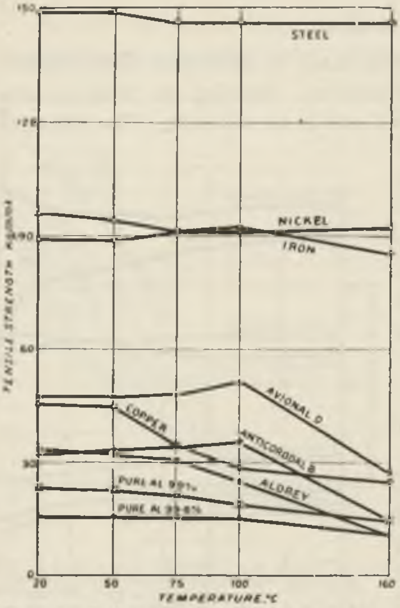

FIG. 5.—Influence of Heating for 1 Year at Different Temperatures on the Tensile Strength of Various Wire Materials. Tested at 20° C.

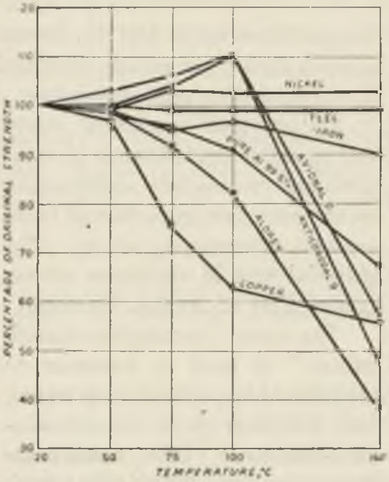

FIG. 6.—Influence of Heating for 1 Year at Different Temperatures on the Tensile Strength of Various Wire Materials Expressed as Percentage of the Original Strength. Tested at 20° C.

The term "complete stabilization" is used in reference to samples which had been first heated for 4 hrs. at their solution-treatment temperature, then allowed to cool in 24 hrs. by stages of 50° C. until the temperature of testing—450°, 400°, 350°, 300°, 250°, 200°, or 100° C. —was reached.

In connection with the earlier investigations, the decrease of the strength resulting from prolonged heating was only determined by means of the Brinell hardness test and it was found that after heating to temperatures exceeding 200° C., the Brinell hardness only approaches that of completely stabilized samples when the heating has extended over lengthy periods; at 300°–350° C. this is usually reached after a heating period of 100–300 days. Information on hardness values alone is, however, insufficient for the requirements of engineers. It was desirable, therefore, that the tests on the effects of long duration heating should be extended to include observations of tensile strength, yield-point, and elongation. These tests have now been carried out on cast and wrought alloys for heating periods up to 720 days. The results for the wrought alloys were as follows.

As in the tests mentioned above, the material was in the form of extruded bars of 20 mm. diameter. The method and conditions of testing were the same as those used previously and were, briefly described, as follows: The apparatus for tests at elevated temperatures is shown in Fig. 7. For the tests at elevated temperature the test-piece shown in Fig. 8, having threaded ends, was used. This was heated without application of stress at the temperature of testing for 20 minutes and then pulled at a rate of testing, which resulted in fracture after 2–5 minutes according to the elongation of the material used. The gauge-length was $\frac{11{\cdot}3\sqrt{F}}{2}$. ($F$ = section.)

In the tension tests at room temperature the same test-piece was used, but the ends were not threaded. The test-pieces were broken in 1–2 minutes. The gauge-length was $11{\cdot}3\sqrt{F}$, and the test results recorded were the average of three.

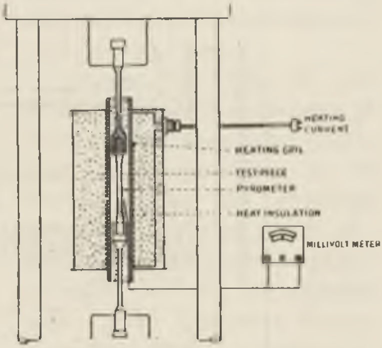

Fig. 7.—Equipment for Hot Tensile Tests.

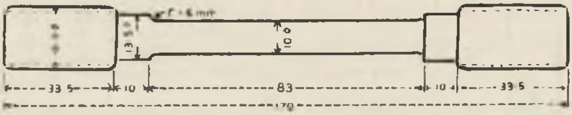

Fig. 8.—Test-Bar for Hot Tensile Tests.

Table I gives the composition and properties of the materials used. The results of the individual tests are not given on account of their very large number.

Strength Tests at Room Temperature.

The various materials—pure aluminium, Peraluman 2, Anticorodal B, Avional D, "Y" alloy, and "R.R. 59" alloy—after heating at 75°, 100°, 130°, 160°, 200°, 250°, and 300° C. for periods up to 2 years, showed different behaviour according to their previous treatment and composition.

Pure Aluminium (Figs. 9 and 10), the original material of which was half hard, shows no decrease at temperatures up at 160° C. At higher temperatures, complete softening occurs by reason of recrystallization. At 250° C., however, the lowest recorded strength properties are

TABLE I.

<table>
<tr><th rowspan="2">Material.</th><th rowspan="2">mm. φ.</th><th colspan="7">Analysis.</th><th colspan="8">Original Mechanical Properties.</th></tr>
<tr><th>Fe.</th><th>Si.</th><th>Mn.</th><th>Mg.</th><th>Cu.</th><th>Ni.</th><th>C.</th><th colspan="2">Yield-Point (σ 0·2), kg./mm.².</th><th colspan="2">Tensile Strength σ_B, kg./mm.².</th><th colspan="2">Elongation δ, Per Cent. l = 200 mm.</th><th colspan="2">Brinell Hardness, kg./mm.².</th></tr>
<tr><td>Aluminium 99·0 wire</td><td>2·5</td><td>0·35</td><td>0·25</td><td>...</td><td>...</td><td>0·02</td><td>...</td><td>...</td><td colspan="2">17·7</td><td colspan="2">23·0</td><td colspan="2">2·82</td><td colspan="2"></td></tr>
<tr><td>" 99·5 "</td><td>2·4</td><td>0·20</td><td>0·15</td><td>...</td><td>...</td><td>0·01</td><td>...</td><td>...</td><td colspan="2">14·1</td><td colspan="2">16·4</td><td colspan="2">1·75</td><td colspan="2"></td></tr>
<tr><td>Aldrey "</td><td>2·4</td><td>0·25</td><td>0·55</td><td>...</td><td>0·44</td><td>0·01</td><td>...</td><td>...</td><td colspan="2">27·4</td><td colspan="2">32·9</td><td colspan="2">8·3</td><td colspan="2"></td></tr>
<tr><td>Anticorodal B "</td><td>3·0</td><td>0·22</td><td>1·02</td><td>0·68</td><td>0·70</td><td>0·02</td><td>...</td><td>...</td><td colspan="2">21·4</td><td colspan="2">32·3</td><td colspan="2">13·1</td><td colspan="2"></td></tr>
<tr><td>Avional D "</td><td>2·8</td><td>0·28</td><td>0·25</td><td>0·55</td><td>0·52</td><td>3·85</td><td>...</td><td>...</td><td colspan="2">29·2</td><td colspan="2">46·6</td><td colspan="2">16·1</td><td colspan="2"></td></tr>
<tr><td>Iron "</td><td>2·4</td><td>R *</td><td>0·05</td><td>0·30</td><td>...</td><td>...</td><td>...</td><td>0·05</td><td colspan="2">79·8</td><td colspan="2">95·7</td><td colspan="2">1·5</td><td colspan="2"></td></tr>
<tr><td>Steel "</td><td>2·0</td><td>R *</td><td>0·15</td><td>0·40</td><td>...</td><td>...</td><td>...</td><td>0·20</td><td colspan="2">119·0</td><td colspan="2">147·7</td><td colspan="2">5·5</td><td colspan="2"></td></tr>
<tr><td>Nickel "</td><td>2·0</td><td>0·25</td><td>...</td><td>0·15</td><td>...</td><td>0·10</td><td>R *</td><td>...</td><td colspan="2">77·9</td><td colspan="2">88·8</td><td colspan="2">1·5</td><td colspan="2"></td></tr>
<tr><td>Copper "</td><td>2·0</td><td>...</td><td>...</td><td>...</td><td>...</td><td>99·99</td><td>...</td><td>...</td><td colspan="2">39·3</td><td colspan="2">45·9</td><td colspan="2">1·8</td><td colspan="2"></td></tr>
<tr><td></td><td></td><td></td><td></td><td></td><td></td><td></td><td></td><td>Ti.</td><td>(1)</td><td>(2)</td><td>(1)</td><td>(2)</td><td>(l = $11{\cdot}3\sqrt{F}$)
(1)</td><td>(l = $\frac{11{\cdot}3}{2}\sqrt{F}$)
(2)</td><td>(1)</td><td>(2)</td></tr>
<tr><td>Aluminium 99·5 round-bar</td><td>10·0</td><td>0·28</td><td>0·18</td><td>...</td><td>...</td><td>0·02</td><td>...</td><td>...</td><td>11·4</td><td>10·7</td><td>11·5</td><td>11·4</td><td>9·3</td><td>17·5</td><td>33·0</td><td>32·0</td></tr>
<tr><td>Peraluman 2 "</td><td>10·0</td><td>0·22</td><td>0·25</td><td>1·45</td><td>2·20</td><td>0·02</td><td>...</td><td>...</td><td>15·7</td><td>12·9</td><td>25·6</td><td>26·5</td><td>12·7</td><td>17·1</td><td>62·0</td><td>61·0</td></tr>
<tr><td>Anticorodal B "</td><td>10·0</td><td>0·25</td><td>0·98</td><td>0·65</td><td>0·72</td><td>0·01</td><td>...</td><td>...</td><td>36·8</td><td>30·5</td><td>38·8</td><td>33·9</td><td>7·6</td><td>14·7</td><td>106·0</td><td>104·0</td></tr>
<tr><td>Avional D "</td><td>10·0</td><td>0·25</td><td>0·28</td><td>0·58</td><td>0·55</td><td>3·78</td><td>...</td><td>...</td><td>34·1</td><td>32·7</td><td>47·6</td><td>46·5</td><td>13·1</td><td>15·8</td><td>112·0</td><td>111·0</td></tr>
<tr><td>"Y" Alloy "</td><td>10·0</td><td>0·22</td><td>0·25</td><td>...</td><td>1·45</td><td>3·95</td><td>1·92</td><td>...</td><td>26·0</td><td>22·9</td><td>39·9</td><td>38·9</td><td>20·1</td><td>23·7</td><td>107·0</td><td>109·0</td></tr>
<tr><td>"R.R. 59" Alloy "</td><td>10·0</td><td>1·15</td><td>0·82</td><td>...</td><td>1·49</td><td>2·35</td><td>1·32</td><td>0·06</td><td>39·2</td><td>29·2</td><td>41·7</td><td>37·1</td><td>7·3</td><td>12·7</td><td>138·0</td><td>121·0</td></tr>
</table>

* R = remainder.

(1) = Original mechanical properties of the test-pieces tested at 20° C.

(2) = Original mechanical properties of the test-pieces tested at the annealing temperatures.

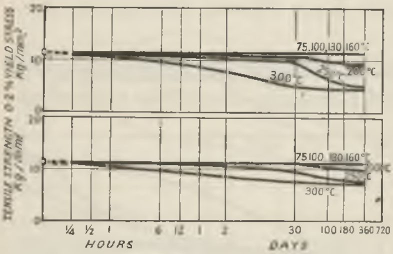

FIG. 9.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of Pure Aluminium. Tested at 20° C.

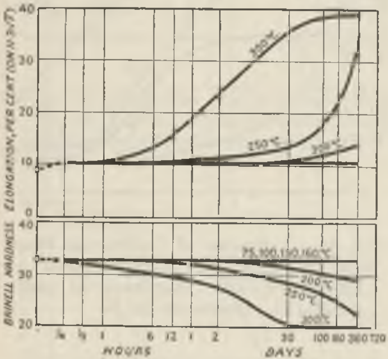

FIG. 10.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of Pure Aluminium. Tested at 20° C.

only attained after a period of 180–360 days. The degree of purity of the metal is here of great influence, since the rate of recrystallization increases greatly with degree of purity.

Peraluman 2 (Figs. 11 and 12) does not show a decrease of strength properties after treatment at temperatures up to 200° C., since the extruded material used corresponds approximately to the soft condition, in which this alloy is normally used. After treatment in the range 250°–300° C., however, a slight decrease is seen.

Anticorodal B (Figs. 13 and 14). Treatment at the lower temperatures, between 75° and 160° C., causes an increase of the yield-point and tensile strength, showing that the material had not yet undergone maximum precipitation-hardening. Treatment at 250° C. only results

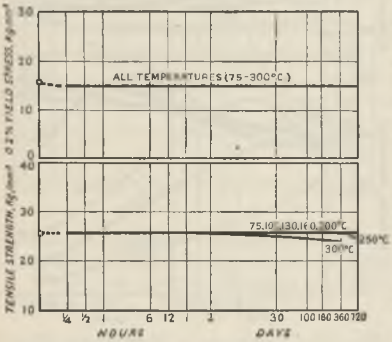

FIG. 11.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of Peraluman 2. Tested at 20° C.

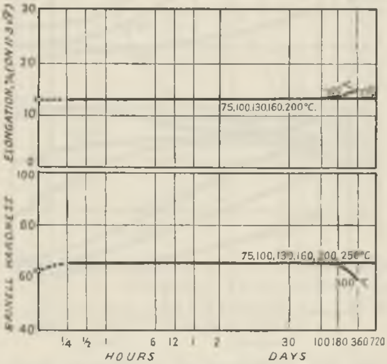

FIG. 12.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of Peraluman 2. Tested at 20° C.

in complete softening after about 360 days, whereas for treatment at 300° C., 180 days are sufficient.

Avional D (Figs. 15 and 16) shows the phenomenon, already described elsewhere,[2] of decreasing strength and increasing elongation after short-

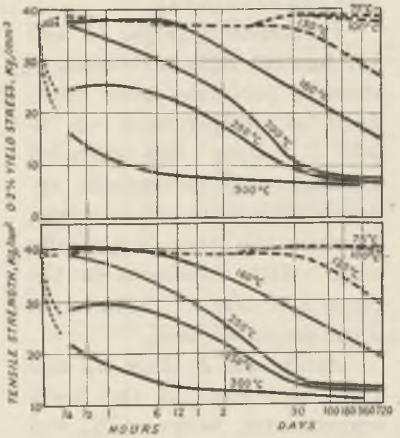

FIG. 13.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of Anticorodal B. Tested at 20° C.

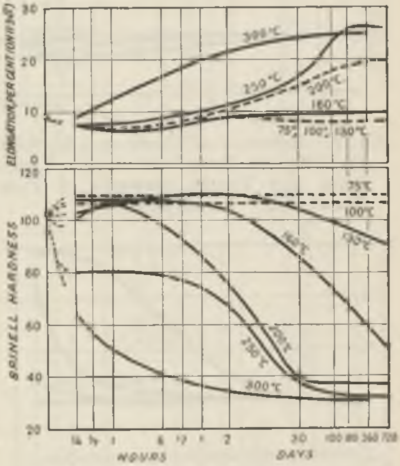

FIG. 14.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of Anticorodal B. Tested at 20° C.

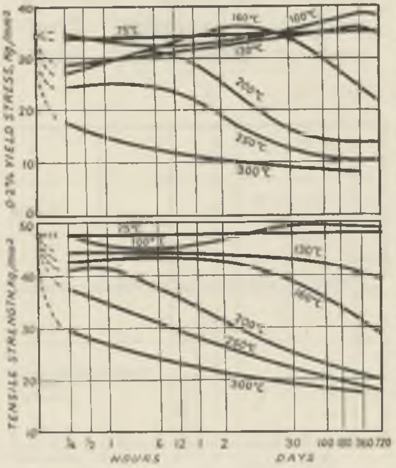

FIG. 15.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of Avional D. Tested at 20° C.

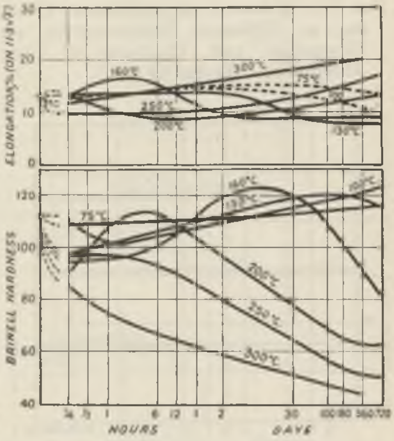

FIG. 16.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of Avional D. Tested at 20° C.

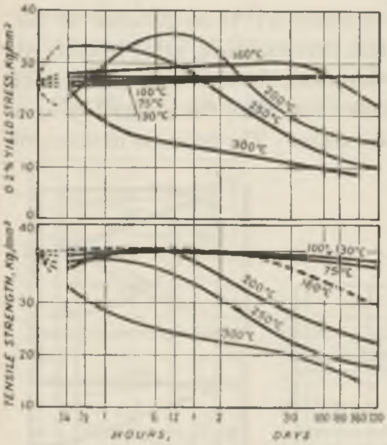

FIG. 17.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of "Y" Alloy. Tested at 20° C.

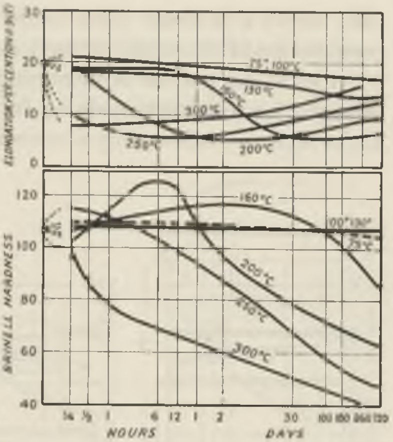

FIG. 18.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of "Y" Alloy. Tested at 20° C.

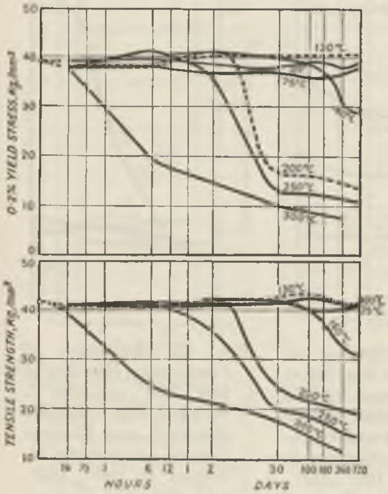

FIG. 19.—Influence of Continuous Heating at Different Temperatures on Yield-Point and Tensile Strength of "R.R. 59" Alloy. Tested at 20° C.

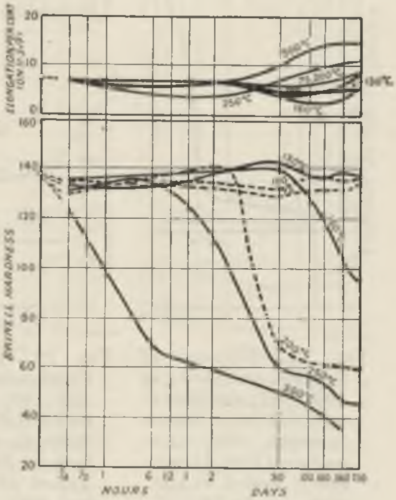

FIG. 20.—Influence of Continuous Heating at Different Temperatures on Elongation and Hardness of "R.R. 59" Alloy. Tested at 20° C.

time treatments of ¼ hr. up to 1–2 days, in which the lowest strength and highest elongation are attained the more quickly the higher the temperature of treatment. After attaining a maximum, the strength then shows a decrease after longer treatments; thus, Avional of the composition of the material used in this investigation, after heating at 75°–100° C. for periods of increasing duration, shows first a decrease in strength for heating periods up to 1 day and then for about 1 year a slight increase, followed by a second decrease. The decrease in strength occurs with an increase in elongation value and *vice versa.* Complete annealing of Avional at 300° C. requires more than 12 months.

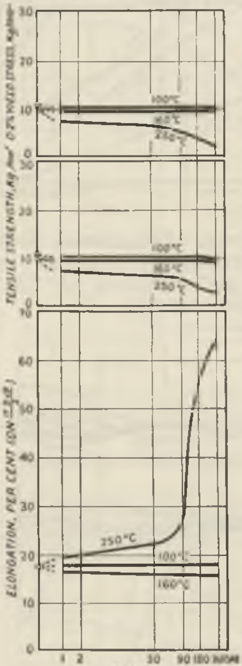

Fig. 21.—Influence of Continuous Heating at 100°, 160°, and 250° C. on Yield-Point, Tensile Strength, and Elongation of Pure Aluminium. Tested at 100°, 160°, and 250° C., Respectively.

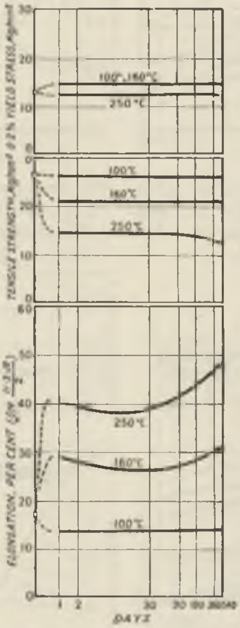

Fig. 22.—Influence of Continuous Heating at 100°, 160°, and 250° C., on Yield-Point, Tensile Strength, and Elongation of Peraluman 2. Tested at 100°, 160°, and 250° C., Respectively.

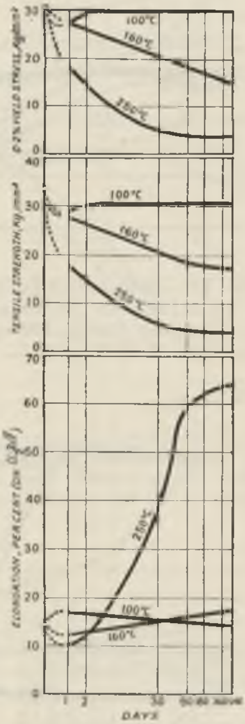

Fig. 23.—Influence of Continuous Heating at 100°, 160°, and 250° C. on Yield-Point, Tensile Strength, and Elongation of Anticorodal B. Tested at 100°, 160°, and 250° C., Respectively.

"*Y*" *Alloy* (Figs. 17 and 18) behaves similarly to Avional. After heating for $\frac{1}{4}$ hr. at 200° C., the strength commences to decrease and later increases, a maximum being attained after 12–24 hrs. by the effect of age-hardening; subsequently, the strength again decreases. The influence of heating for periods of 2 years up to 130° C. on the strength is, as in the case of Avional, still scarcely perceptible.

"*R.R. 59*" *Alloy* (Figs. 19 and 20). This alloy shows a similar behaviour to that of Avional and "Y" alloy after short-time heating, the decrease of the 0·2 per cent. yield stress and the subsequent increase being less strongly marked, however.

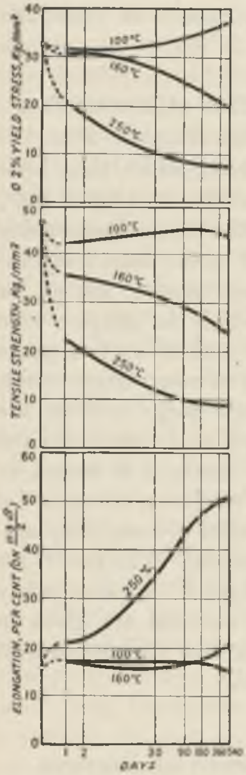

FIG. 24.—Influence of Continuous Heating at 100°, 160°, and 250° C. on Yield-Point, Tensile Strength, and Elongation of Avional D. Tested at 100°, 160°, and 250° C., Respectively.

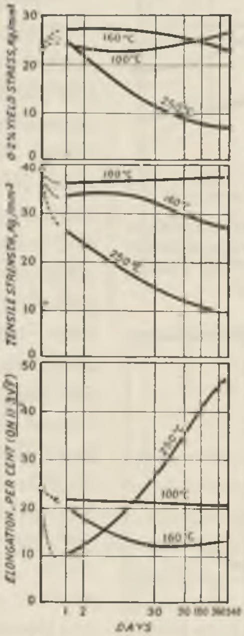

FIG. 25.—Influence of Continuous Heating at 100°, 160°, and 250° C., on Yield-Point, Tensile Strength, and Elongation of "Y" Alloy. Tested at 100°, 160°, and 250° C., Respectively.

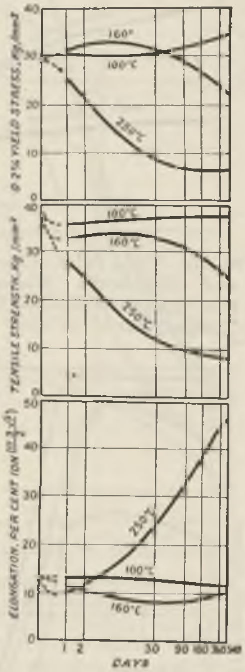

FIG. 26.—Influence of Continuous Heating at 100°, 160°, and 250° C. on Yield-Point, Tensile Strength, and Elongation of "R.R. 59" Alloy. Tested at 100°, 160°, and 250° C., Respectively.

After heating to temperatures up to 130° C. for 2 years, the strength

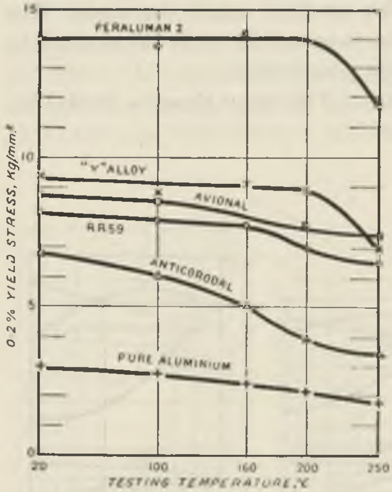

Fig. 27.—Yield-Point of Pure Aluminium and Aluminium Alloys Completely Stabilized as Depending on the Testing Temperature of 20°–250° C.

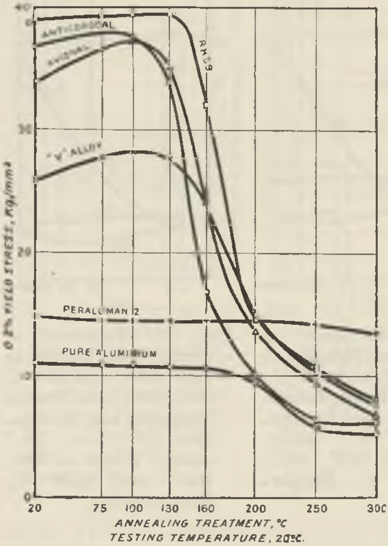

Fig. 28.—Influence of Continuous Heating for 1 Year at Temperatures up to 300° C. on the Yield-Point of Pure Aluminium and Aluminium Alloys. Tested at 20° C.

properties remain unaffected. Alloy " R.R. 59 " is exceptionally strong at elevated temperatures and differs from Avional in that heating for 1–2 years at 300° C. does not result in complete softening.

Strength Tests at Elevated Temperature (Figs. 21–26).

These tests, which have been carried out after heating at 100°, 160°, and 250° C. (the tests being carried out at the same temperatures), show results similar to those in which the samples were heated for periods up to 2 years and tested at room temperature. After a very long, or from the practical point of view, unlimited time, the material is bound to attain the minimum strength, *i.e.* the properties of completely stabilized material. The results for the yield-point of completely stabilized material are given in Fig. 27 for test temperatures of 20°–250° C.

Discussion of Results.

To get a better idea of the results of the samples tested at 20° C. and at elevated temperatures as represented by the former figures, we have produced the curves 28 and 29. We have collected the results of our investigations, carried through with different alloys in Fig. 28, showing the influence of annealing at different temperatures between

75° and 300° C. and the testing results at 20° C. on the yield-points. On Fig. 29 the yield-point tested at annealing temperature is represented. The curves for Avional and “ Y ” alloy show a marked increase in yield-point (room temperature) after heating to temperatures up to 100° C. “ R.R. 59 ” is distinctly superior to Avional and “ Y ” alloy after heating at temperatures up to 160° C. On the other hand, the yield-point of the copper-free alloy Anticorodal decreases very rapidly after 100° C., and at 160° C. is already much lower than that of alloy Avional, “ Y ” alloy, and especially “ R.R. 59.” After heating at 200° C. the yield-point of Anticorodal is even lower than that of the alloy Peraluman 2 and nearly as low as that of pure aluminium. The same effect can also be observed in Figs. 9–20, but less clearly. Further, it can be seen that even at 160° C. “ R.R. 59 ” is distinctly superior to “ Y ” alloy and Avional.

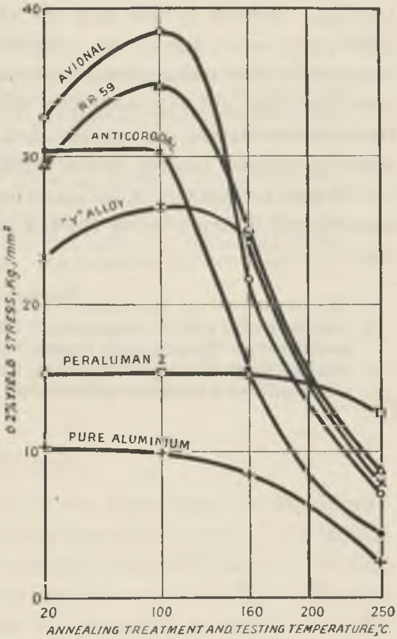

Fig. 29.—Influence of Continuous Heating for 1 Year at Temperatures up to 250° C. on the Yield-Point of Pure Aluminium and Aluminium Alloys. Tested at the Annealing Temperature.

Pure, half-hard aluminium is not affected appreciably up to 160° C. For 300° C. the differences between the alloys become very small; only Peraluman 2, which is used in the soft state and is not heat-treatable, forms an exception, as would be expected. Even after heating at 300° C. it shows no substantial decrease in strength and has a higher strength than all the other alloys.

Almost the same observation is obtained from an examination of the temperature of heating on the yield-point of the alloys heated for 1 year at 100°, 160°, and 250° C. and tested at these temperatures (Figs. 21–26 and Fig. 29). “ R.R. 59 ” is found to be distinctly superior up to 160° C. It will be noted that the 0·2 per cent. yield-point of “ R.R. 59 ” still increases up to 100° C. This suggests that the precipitation-hardening of the original material was incomplete.

Our investigations have shown that it would not be permissible to take as final strength after prolonged heating at a given temperature the results of tests of fully stabilized material tested at room temperature or at the temperature of heating after being heated to the temperature concerned. It appears to be necessary to carry out the lengthy procedure of prolonged heating tests for each alloy.

The curves included in the present paper give the strength of pure aluminium and of some wrought aluminium alloys after prolonged heating. Results of this kind are often required by engineers and the yield-point is of service as a basis for calculation. The yield-point observed in the normal tensile test (0·2 per cent. permanent extension), however, does not represent the strength for 0·2 per cent. permanent strain under loading of long duration. The matter of permissible stress under prolonged loading is one relating to the so-called " creep-limit " at different temperatures, which as indicated by results obtained to date, sometimes lies much below the 0·2 per cent. yield of the normal tensile test.

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[1] A. von Zeerleder and P. Bourgeois, "Effect of Temperatures Attained in Overhead Electric Transmission Cables," *J. Inst. Metals*, 1929, **42**, 321.

[2] A. von Zeerleder, M. Bosshard, and R. Irmann, "Warmfestigkeit und Warmhärte verschiedener Aluminiumlegierungen," *Z. Metallkunde*, 1933, **25**, 293.

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SHRINKAGE DURING THE SOLIDIFICATION OF ALUMINIUM ALLOYS.*

By VAUGHAN H. STOTT,† M.Sc.

Synopsis.

If a molten pure metal or a eutectic alloy be solidified upwards in a bottle-shaped vessel with a long narrow neck, it is possible to obtain an ingot free from cavities and to measure the shrinkage during solidification by direct observation of the descent of the liquid in the neck of the vessel. Measurements made in this way for aluminium and its eutectic with silicon are in close agreement with those of Edwards (*Chem. and Met. Eng.*, 1921, **24**, 61; 1923, **28**, 165) who used a different method. Calculations based on the data of Edwards and others, show that the shrinkage of the commercial aluminium-rich alloys not containing silicon, is usually between 6·5 and 8 per cent. The presence of silicon results in a marked lowering of the shrinkage.

It is pointed out that although these results have a bearing on the formation of pipes in castings, it must not be supposed that the size of a pipe in given circumstances is proportional to the shrinkage of the metal during solidification. For a detailed discussion of this matter the reader is referred to a paper by Edwards and Gammon (*Chem. and Met. Eng.*, 1921, **24**, 338).

Introduction.

The programme of general research of the Department of Metallurgy of the National Physical Laboratory has included a study of the factors which influence the production of castings from fluid metal. Among these factors is the change of volume which a metal or alloy undergoes during solidification. In the course of a research undertaken for the British Non-Ferrous Metals Research Association at Sheffield University, a differential gas dilatometer was devised [1] which allowed such determinations to be made under conditions which ensured equilibrium, and with an improved form of this apparatus a series of determinations was made for metals and alloys of low melting point.[2] It was desired to extend this work to aluminium and its light alloys, but the original method could not be adopted, owing to the action of molten aluminium on the silica vessels which might have been used as containers. An apparatus was constructed in which the gas-tight parts were of metal,

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† Department of Metallurgy, National Physical Laboratory.

Note to Abstractors and Other Readers.—This paper will be published, in permanent form, in the *Journal of the Institute of Metals*, Vol. LIX, 1936. Reference should accordingly be as follows: *J. Inst. Metals*, 1936, **59** (Advance copy).

but experimental difficulties were encountered in its use, and recourse was had to a simple form of dilatometer by means of which measurements could be made directly.

EXPERIMENTAL METHOD.

A clay–alumina pot, of the shape shown in Fig. 1, and containing molten metal, was cooled from the bottom. The shrinkage of the metal as it solidified caused a gradual descent of the still molten metal in the neck, which was observed until solidification was complete. Temperature measurements were made at the bottom of the pot, and the first discontinuity in the curve obtained by plotting liquid levels against temperature, indicates the beginning of solidification.

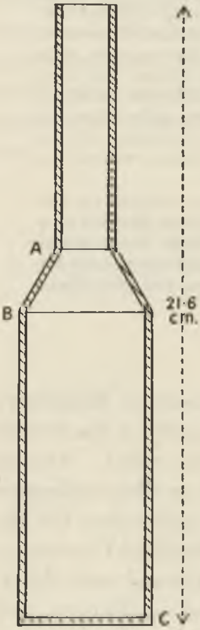

FIG. 1.

In order to ensure that solidification of the metal proceeded always upwards, a Nichrome heating coil wound on a silica tube was slipped over the neck of the pot. The pot of metal stood on the bottom of an outer thin-walled cylindrical pot some 30 cm. in height, and about 7 cm. in external diameter. The space between the two pots was filled to the height of the inner pot with calcined alumina. The outer pot was placed in a vertical Nichrome resistance furnace of which the tube was 60 cm. in length, and 7·5 cm. in bore. When in position, the point *A* (Fig. 1) of the inner pot was half-way up the furnace. With this arrangement a difference of temperature of some 15° C. was observed between thermocouples tied to the outside of the pot of metal at the points *B* and *C* of Fig. 1.

The volume of the inner pot at various levels with respect to the mean level of four points equally spaced round the top of the neck, was determined by filling with mercury to levels measured by making an electrical contact between the centre of the meniscus and a wire mounted on a micrometer.

At the high temperatures, two methods were used for the determination of the level of the liquid metal. The first method was the same as that used in calibration, except that tungsten instead of nickel wire made the contact. With pure aluminium on which the first experiments were conducted, this method was successful in spite of a certain amount of difficulty occasioned by the gradual accumulation of dross.

Very great accuracy in the measurement of the level of the liquid was fortunately not required, since a change in level of about 1 cm. was produced by a shrinkage of 1 per cent. of the volume of the metal. At the same time it was realized that the difficulties resulting from the formation of dross would be greater with alloys than with pure aluminium. A simpler, and very satisfactory method of measurement was then used. A Nichrome wire of 14 S.W.G. was bent into a hook at its upper end and was hung on the lowest point of a curved knife edge mounted on a vertical micrometer. The molten metal was so strongly illuminated as to appear quite cold, and visual inspection sufficed to determine when the lower end of the wire was in contact with the metal. For this purpose a slight swing was given to the wire while it was being lowered, and the swinging would stop immediately when the wire reached the film of oxide on the surface of the metal. The wire was then moved about, while just in contact with the scum, in such a way as to cause the meniscus to assume a position of equilibrium. In all the experiments the falling meniscus was almost flat, and concave upwards, the normal form of meniscus being observed only during an earlier stage of the experiment while the temperature of the metal was increasing. The position of the centre of the meniscus was finally determined by lowering the wire very cautiously until slight lateral movement produced a bright streak on the molten surface, showing that the film of oxide had been penetrated. The measurements were relative to the positions of four points equally spaced round the top of the neck of the pot. The mean height of the top of the pot was always measured before the beginning of the solidification and after its close, and the measurements agreed to within about 0·05 mm. It was not necessary to make corrections for changes of dimensions of *any* part of the apparatus during the solidification of the metal. In particular, it will be realized that the ratios of volumes determined by calibration with mercury at room temperature remain unchanged at all temperatures.

In order to fill the pot with metal, a number of rods of $\frac{1}{2}$ in. diameter (a loose fit in the neck of the pot) were placed in the cold pot, and reached the height B shown in Fig. 1. A central rod reached nearly to the top of the neck, which was temporarily closed with a rubber stopper to prevent ingress of alumina during packing. The pots were placed in the furnace in the afternoon, and the heating adjusted so that by the next morning a temperature of about 400° C. would be attained. The heating currents were then increased, and when all the metal was molten additional rods were dropped through the neck of the pot with the aid of tongs. These rods had been preheated to a temperature a little below the melting point in order not to risk cracking the pot. At the end of this process the pot was almost filled to the top of the neck, and it was then stirred as thoroughly as possible with a Nichrome wire to remove bubbles and scum, which were liable to be produced owing to the tenacity of the oxide films on the surfaces of the rods. The stirrer was bent so that it could reach the periphery of the bottom of the pot, since preliminary experiments showed that bubbles were liable to be formed in this position. After some experience, ingots were obtained having densities as high as 2·695 at 17° C. The density of pure aluminium has been found by the use of X-rays to be 2·699. The metal used in the experiments now described contained 0·1 per cent. iron and 0·05 per cent. silicon, which would increase its density to 2·701. Sawing through the axis disclosed no visible cavities in these castings. In the preliminary experiments stirring was less thorough, as it was feared, needlessly, that the Nichrome wire might be attacked by the metal. In these experiments small cavities were found on the surfaces of the castings, and in one case in the interior. These cavities had apparently been present in the liquid state, the measured shrinkage of the metal during solidification being unaffected by their presence. In all the experiments the liquid from which the metal rods were prepared had been freed from gas as far as possible by passing through it a current of nitrogen mixed with the vapour of carbon tetrachloride.

In calculating the results, a correction was made for the adherence of metal to the inner surface of the neck of the pot. This correction was obtained by subtracting the weight of the ingot after breaking away the pot, from the total weight of metal originally in the pot. It was assumed that no metal adhered to the pot below the neck. This was substantially exact for aluminium and the aluminium–silicon alloy of eutectic composition, but was not accurate in the case of two other alloys. The correction for aluminium was only about 1 part in 500 of the total volume. A small correction was also made for the temperature gradient in the liquid at the beginning of solidification. It was assumed that this gradient was roughly indicated by the readings of the two thermocouples. This is probably far from exact, but the correction was only about 1 part in 1000 of the total volume.

The duration of freezing was usually 2 or 3 hrs., although in one case it was prolonged to 6½ hrs. The times of the various temperature measurements were noted, so that the time–temperature curves afforded a check on the volume–temperature curves. The indications of the beginning of solidification afforded by the two methods, were in agreement.

Fig. 2 shows typical curves obtained from an experiment on aluminium, an e.m.f. of 1 mv. in a thermocouple being roughly equivalent to 100° C.

Results and Comparison with Other Work.

The results given below were obtained for aluminium containing 0·1 per cent. iron and 0·05 per cent. silicon.

Shrinkage during Solidification,
Per cent.
6·9
6·7
6·7
6·6
mean 6·7

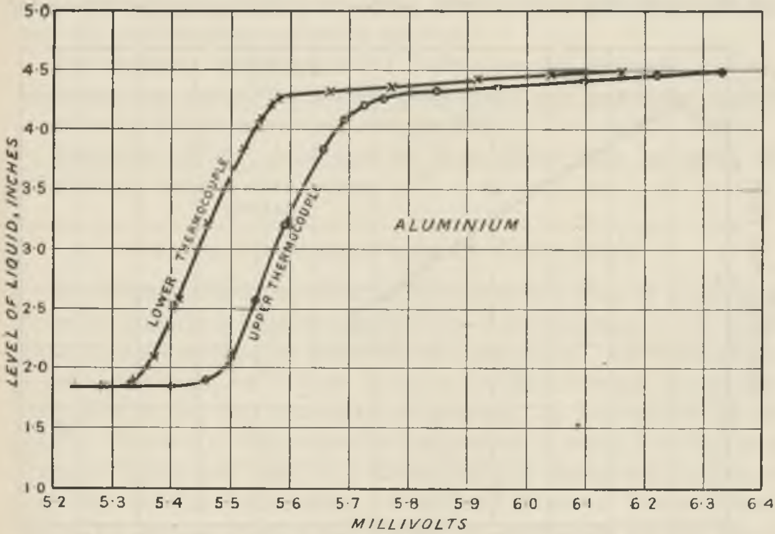

Fig. 2.

Edwards and Moorman [3] obtained the value 6·6 per cent. by calculation from the values of the densities of solid and liquid metal at the melting point. The density of the solid metal was calculated from its value at 20° C. and a formula representing its change with temperature. At the time of publication the authors were awaiting, from the Bureau of Standards (U.S.A.), the results of measurements of the thermal expansion of the particular aluminium used (99·75 per cent. aluminium). These results have since been published, and recalculation of the shrinkage yields the value 6·7 per cent. It is difficult to estimate the exactitude of the value, but the error from all sources is probably less than five units in the second significant figure. Direct measurements made earlier than those of Edwards and Moorman yielded lower values of the shrinkage, apparently owing to the formation of cavities in the solid metal.

After testing the method of measurement on aluminium, the present author determined the shrinkage of an aluminium–silicon alloy having a composition slightly on the silicon side of the eutectic. (The eutectic alloy contains 11·6 per cent. of silicon.) An alloy containing silicon 12·59, iron 0·46, and titanium 0·05 per cent., gave a shrinkage of 3·7 per cent. A duplicate determination on an alloy containing 12·75 per cent. silicon and impurities as before, gave a shrinkage of 3·4 per cent. The mean of these values has been marked with a cross on Fig. 3. The author's value for the shrinkage of pure aluminium is also indicated in the same way. The circles represent values given by Edwards,[4] and the line represents the results of calculations by the present author based on the data of Edwards.

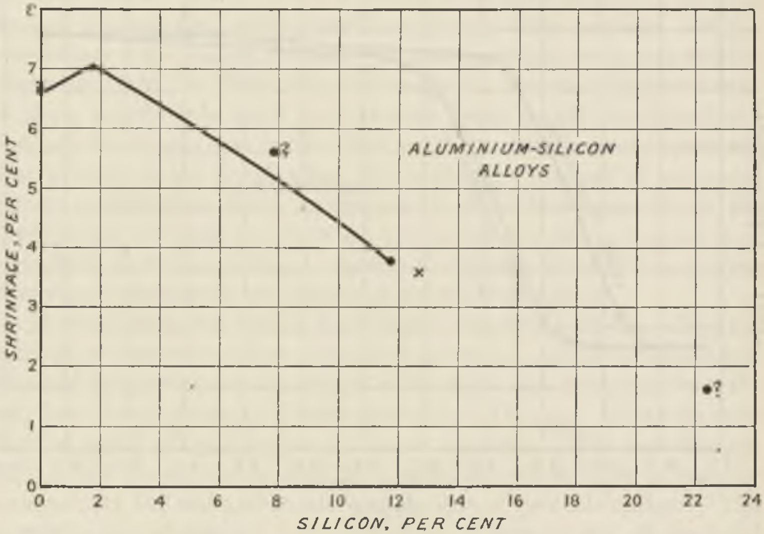

Fig. 3.

The value of the shrinkage given by Edwards corresponding with 7·81 per cent. silicon is inconsistent with both his numerical data and his graph. This discrepancy is probably due to the inadvertent substitution of the percentage expansion of the metal on melting for the corresponding shrinkage on solidification. The circle corresponding with 22·4 per cent. silicon represents a value quoted by Edwards as a good approximation based on fewer data then for his other values. The maximum of the shrinkage curve at 1·7 per cent. silicon corresponds with the maximum freezing range at this composition.

Shrinkage measurements on the two alloys known as " L5 " and " Y " were not satisfactory. Three experiments were made on the alloy "L5," and shrinkages were observed of 4·4, 4·8, and 4·4 per cent. On cutting open the first ingot, it was found that a pipe had formed, running from the centre of the bottom to the level *B* shown in Fig. 1. Similar pipes were formed in succeeding experiments. Their volumes were estimated by filling them with vacuum wax. On correcting the shrinkages in this way values of 6·1 and 5·6 per cent. were obtained. The bottom surface of each ingot, however, had a rough appearance suggestive of having been torn away from the bottom of the pot. On this assumption there would be an air space of unknown magnitude between the metal and the pot. From these considerations it is probable that the true shrinkage exceeds 6 per cent.

The apparent shrinkage of "Y"-alloy was 4·8 per cent., but the ingot was very porous, the cavities being closed, and distributed so that an estimate of their volume was not possible.

Estimates of the shrinkages of these alloys may be made by calculation.

ESTIMATES OF SHRINKAGE BY CALCULATION.

It will be assumed that the specific volume of a phase of an alloy of given composition is a continuous function of temperature for any range which does not involve the formation of a new phase. This assumption is not self-evident, for if there is any connection between the specific volume of an alloy and those of its components, the discontinuity in the specific volume of a component at a temperature at which it undergoes a change of phase may result in a discontinuity in the specific volume of the alloy at that temperature. Goodrich,[2] however, has published curves for certain alloys which verify the assumption for the liquid state for temperature ranges which include the freezing point of one of the components. Braesco [5] has published a similar verification for alloys which are solid solutions. Further evidence is afforded by the consideration that a discontinuity in volume is normally associated with an irregularity in a thermal curve, but such irregularity has probably never been observed unaccompanied by a change of phase.

Let there now be considered, at a temperature t, an alloy consisting of a single phase of specific volume v, formed from n parts by weight of a pure component of specific volume v_1, and $(1-n)$ parts by weight of a pure component of specific volume v_2.

Since v is a function of n and t, and v_1 and v_2 are functions of t, it is *always* possible to write

$$v = nv_1 + (1 - n)v_2 + f(n, t) \quad . \quad . \quad . \quad . \quad (1)$$

the function f being defined by this equation.

Let it be supposed that the form of the function f has been established by experiment for a range of temperature within which v, v_1, and v_2 are continuous. Let it further be supposed that at some temperature or temperatures discontinuity occurs in one or both of the quantities v_1 and v_2. If v is continuous at such temperatures, the function f must be discontinuous in order to satisfy equation (1); but the continuity of v implies the possibility of extrapolating the right hand side of equation (1) term by term, using for the purpose the three functions of t, namely v_1, v_2, and f, in the forms in which they appear in equation (1) for the range of temperature within which all the functions are continuous.

Such empirical extrapolation is naturally subject to uncertainty. Fortunately, the coefficients of thermal expansion with which we shall be concerned are too small for this uncertainty to be important. Moreover, the extrapolations will be simplified by the circumstance that the value of $f(n, t)$ will always be zero, although this is not relevant to the validity of the principle which has been established.

Let there now be considered a binary alloy of aluminium containing 18 per cent. zinc. According to Hanson and Gayler [6] this alloy begins to freeze at 630° C., and is completely solid at 580° C. It is necessary, therefore, to calculate the specific volume of the liquid alloy at 630° C. and of the solid alloy at 580° C., in order to find the shrinkage on solidification.

From measurements made by Edwards and Moorman,[3] the specific volume (c.c./grm.) of molten aluminium is equal to

$$0{\cdot}4198/[1 - (0{\cdot}0001142)(t - 658)]$$

From the work of Hogness [7] the specific volume of molten zinc is equal to

$$0{\cdot}15175[1 + (0{\cdot}0001472)(t - 419)]$$

Bornemann and Sauerwald [8] have shown that the volumes of the aluminium–zinc alloys may be calculated by the mixture rule, both in the liquid state, and in the solid state at room temperatures. For the solid state at higher temperatures the validity of the rule will be assumed. By applying the mixture rule to this liquid alloy at 630° C., the specific volume is found to be 0·3713.

The corresponding calculation must now be made for the solid state at 580° C.

The specific volume of liquid aluminium at its freezing point is 0·4198. With the aid of the present author's mean value of 6·72 per cent. shrinkage on solidification, the specific volume of solid aluminium at 658° C. is found to be 0·3916. The above values are for aluminium containing about 1 part in 400 of impurities. For such metal it has been shown by Souder and Hidnert[9] that

$$L_t = L_0[1 + (21{\cdot}90\,t + 0{\cdot}0120\,t^2) \times 10^{-6}]$$

where L_t is the length of a specimen at t° C., and L_0 is the corresponding length at 0° C. From this formula the ratio of the specific volume of the solid metal at any temperature to its value at 658° C. may be calculated. The latter value being known, it is found that the specific volume of solid aluminium at 580° C. is 0·3883.

It is now necessary to find the specific volume of "solid" zinc extrapolated to 580° C.

Liquid zinc at 419° C. has a specific volume of 0·15175 c.c./grm. Goodrich[2] has shown the shrinkage on solidification to be 4·5 per cent. Hence, the specific volume of solid zinc at 419° C. is 0·1449. The value at 20° C. (from X-ray measurements) is 0·1399. If the variation with temperature be linear, the specific volume of solid zinc is

$$0{\cdot}1449[1 + (0{\cdot}0000865)(t - 419)]$$

Its extrapolated value at 580° C. is 0·1469. By the mixture rule the specific volume of the solid alloy at 580° C. is 0·3449. The shrinkage on solidification is, therefore, 7·1 per cent.

It is to be observed that this figure depends mainly on the temperatures at which the alloy begins and finishes freezing, and is little influenced by inaccuracies in the assumed values of other properties of zinc involved in the calculations. Thus, if the specific volume of liquid aluminium be extrapolated to 630° C. and the percentage decrease from this value to that of the solid at 580° C. be calculated, the value 7·2 is obtained. The small influence of the zinc is due partly to its small specific volume relatively to that of aluminium, and partly to the considerable shrinkage exhibited by the metal in the free state. (The case of silicon is markedly different. Edwards is of the opinion that pure silicon expands on solidification, but there is no certain information on this point.)

The specific volume and shrinkage on solidification of copper differ little from those of zinc, and the main effect of the presence of a little copper in an aluminium–zinc alloy is to alter the range of temperature within which solidification occurs. Thus, a shrinkage of about 7·8 per cent. would be expected for an alloy containing 5 per cent. copper and 15 per cent. zinc, with a freezing range from 611° to 480° C. There is some uncertainty concerning the theoretical freezing range of the alloy "L5" containing 3 per cent. copper and 15 per cent. zinc, but the calculated shrinkage must be within the limits of 7 and 8 per cent.

In the case of "Y"-alloy, which has a freezing range from 630°

to about 493° C., and contains 4 per cent. copper, 2 per cent. nickel, and 1·5 per cent. magnesium, a shrinkage of about 8 per cent. would be expected.

Edwards [10] has shown that the shrinkage of an aluminium alloy containing 8 per cent. copper, with no other metal, is about 7 per cent. For alloys with smaller amounts of copper, down to the eutectic proportion of 5 per cent., the shrinkage would be almost the same. The shrinkages of alloys containing less than 5 per cent. copper lie between 7 per cent. and the value 6·7 per cent. found for pure aluminium.

CONCLUSIONS.

The above results are summarized in the statement that the shrinkage on solidification of aluminium and the commercial aluminium-rich alloys not containing silicon, is usually between 6·5 and 8 per cent., whilst the shrinkage of aluminium–silicon alloys containing from 11 to 13 per cent. silicon is about 3·5 per cent.

Although these results have a bearing on the formation of pipes in castings, it must not be supposed that the size of a pipe in given circumstances is proportional to the shrinkage of the metal during solidification. The theory of the formation of pipes is complex, and at present incomplete, but the general principles have been clearly stated by Edwards and Gammon [11] in a paper on the "Causes of Piping in Aluminium Ingots."

ACKNOWLEDGMENTS.

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THE COMPLEX INTERDEPENDENCE OF THE PROPERTIES OF ALLOYS AND THE INDUSTRIAL CONDITIONS OF THEIR MANUFACTURE, TESTING, AND USE.*

By MARQUIS R. DE FLEURY,† MEMBER, and DR. H. PORTIER.‡

SYNOPSIS.

Indirect factors influencing the properties of light alloys in service are classified and some methods of recognizing and controlling them are described.

I.—INTRODUCTION.

(By R. de Fleury.)

THE present study summarizes the experience of many years, directing attention to the unsuspected and troublesome factors which intervene at every stage between the production and utilization of an alloy. After a preliminary qualitative study [1] I have persisted in attempts to discover the laws governing the various factors, and to evaluate their influence. Work for the Air Ministry (of France) has given me the opportunity to apply and verify my results in this sphere of application. The bearing of these problems on the safety of aircraft is sufficient to show their importance. The following presentation may seem complicated, but it is the complication of the subject itself which is to blame, rather than its analysis. In the latter I have had the invaluable assistance of M. Portier.

GENERAL.

All metallurgists in contact with manufacture are aware of the many apparently unrelated factors which are brought to light during experiments with new alloys, new processes, or new tests. In addition to the problems involved in the pursuit of improved properties, in meeting more rigid tests, in the perfecting of manufacturing technique, or imposed by the design and dimensions of the article, either in the foundry or in use, there are the more general but less evident troubles

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* Manuscript received March 20, 1936.

† Ingénieur des Arts et Manufactures, Paris.

‡ Ingénieur des Arts et Métiers et de l'École Supérieure de Fonderie, Paris.

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which intervene unexpectedly, in a manner at once more deceiving to the founder and more dangerous to the user, since they are difficult to trace and are often aggravated by apparently logical remedies. In the interpretation of an incident of either fabrication or functioning, both direct and indirect factors should be considered; better still, they should be foreseen by the Management from a continual analysis of the facts of production.

METHOD OF CLASSIFYING FACTORS.

Observations made by one of the authors,[2] over many years, led him to classify the various factors even before defining them, according to Table I.

The direct factors are revealed by existing knowledge and experience. It is the less apparent factors that are mainly to be considered, for it is impossible to foresee what may be the indirect repercussions of a remedy applied as the result of an error in diagnosis. It is wrong to classify remedies in the same way as the factors to which they correspond, since, for example, a remedy for a direct factor (1–C) can have disastrous consequences in introducing a new indirect (2–C) or even unrecognized factor (3–C). In the closed cycle comprising the various stages in the life of an article, from its entry to the foundry as raw material to its return as scrap, a number of factors intervene, which appear to be unrelated, but which are actually so interdependent that a simple modification of one can upset or re-establish the equilibrium of the whole cycle.

TABLE I.

Influence of the Factors.	Elucidation of the Factors.		
	1. Direct (Immediate Interpretation Obvious to the Senses).	2. Indirect (Requiring the Interpretation of Observed Facts).	3. Temporarily Unrecognized (Brought Out by Hypothesis and Experiment).
A. Apparent (*i.e.* more or less deceptive) .	1–A	2–A	3–A
B. Occasional (dependent on a particular circumstance) . .	1–B	2–B	3–B
C. Fundamental . .	1–C	2–C	3–C

In the paper which follows this interdependence is studied in the particular case of light-alloy castings, although the problem is the same for all metals.[3]

II.—INFLUENCE ON THE PROPERTIES OF ALLOYS OF INDIRECT FACTORS DURING MELTING.

Experience shows that, in the best-directed foundries, the quality of the product is subject to inexplicable fluctuations which often disappear of their own accord. They are due to the intervention of indirect factors, the influence of which is small but cumulative, such as the increase in the amount of impurities in the metal by repeated remelting, and the consequent depreciation. The amount of such depreciation can be measured by the change in a given property, such as castability, elastic limit, &c., or more simply by the increase in the number of rejects. It increases with the "net output ratio," M (*i.e.* the ratio of the weight of metal charged in the furnace to the weight of good castings leaving the works). This differs from the "gross output ratio," m (the ratio of the weight of metal charged to the total weight of castings). Then if r is the proportion of scrap recharged at each melt

$$\frac{1}{M} = \frac{1}{m}(1 - r) \quad . \quad . \quad . \quad . \quad . \quad . \quad (a)$$

At each charge the weight of good castings is replaced by an addition of new metal equal to $\frac{1}{M}$ and the depreciation takes place on the complement, A, of the preceding melt.

If α is the elementary coefficient of depreciation for each melt, the depreciation, T, after n melts is given by the following expression, which forms a convergent series:

$$T_n = \frac{\alpha A[1 - A^n(1 - \alpha^n)]}{1 - A\,(1 - \alpha)} \quad . \quad . \quad . \quad . \quad . \quad (b)$$

and, after an infinite number of remelts:

$$T_\infty = \frac{\alpha\, A}{1 - A\,(1 - \alpha)}$$

or, in terms of M

$$T_\infty = \alpha(M - 1) + 1 \quad . \quad . \quad . \quad . \quad . \quad . \quad (c)$$

The depreciation is thus proportional, partly to the coefficient of depreciation, which depends mainly on direct factors (which are under control and should be favourable) and, partly, to M, which varies with the more or less insidious action of external factors, of which the influence is often unrecognized, and for which the *true* remedies often differ greatly from the *immediate* remedies.

III.—INTERDEPENDENCE OF GENERAL EXTERNAL FACTORS AND THE PROPERTIES OF ALLOYS.

Organization of Production.

If a factory be considered the current manufacture of which has a mean output of M_2, and it be supposed that important orders have to be met, for which the true output ratio, M_1 is greater than M_2 on account of foundry demands or increase in scrap percentage; a mean output ratio M' results, which is given by : [4, 5]

$$M' = \lambda M_1 + (1 - \lambda)M_2 \quad . \quad . \quad . \quad . \quad . \quad . \quad (d)$$

where λ is the percentage of the new manufacture, *e.g.* a delicate type of aluminium aero-engine cylinder block involving a high scrap percentage ($M_1 = 4$) is introduced into a current production whose output ratio has been reduced to 1·5 (M_2). Then the mean output ratio is given by equation *d* as 2·75. This increase is clearly due to the mixing of the two manufactures, and it would be useless to apply direct remedies for the loss in efficiency of the original manufacture, which had not previously given trouble. The fundamental remedy lies in the reorganization of the production on the following lines: The equilibrium of the two manufactures can be not only re-established, but even improved by casting "in echelon," *i.e.* a certain number of melts are made for the current work, and the remainder for the new work, the scrap, however, being pooled. The new manufacture will have a virtual output ratio M_1', lower than M_1, the weight of castings produced at each melt remaining equal to $\frac{1}{M_1}$. The difference, therefore, must be absorbed by the remainder of the production, each melt of which will produce a weight of castings proportional to $\frac{1}{M_2}$ and equal to $\left(\frac{1}{M_1'} - \frac{1}{M_1}\right)\frac{1}{M_2}$.

Hence $\quad \lambda\left(\frac{1}{M_1'} - \frac{1}{M_1}\right)\frac{1}{M_2}$ must be equal to $\frac{1}{M_1}(1 - \lambda)$

or

$$\frac{M_1}{M_1'} = 1 + M_2 \cdot \frac{1 - \lambda}{\lambda} \quad . \quad . \quad . \quad . \quad . \quad . \quad (e)$$

The equation (*e*) gives a value of 1·60 for M_1' in the case of the examples chosen. It shows, too, that if the proportion of the new manufacture is not too great, or if its output ratio, M_1, does not exceed a certain value, the virtual output ratio can equal unity; that is to say, all the excess metal can be used up by the original manufacture without appreciable degradation.

Influence of Specialization.

Certain firms, for economic reasons, specialize in the mass-production of a particular article. Such production is in a precarious state of stability. In the manufacture of automobile cylinder blocks the normal scrap percentage is about 12 per cent., but it periodically reaches 24 per cent., above which it is prohibitive. The reason is the depreciation of the metal with successive re-meltings by an amount which increases with the ever-increasing percentage of scrap reintroduced. The same thing is seen in aircraft work, and aluminium and magnesium castings generally, for which the output ratio is high and purity of the raw materials is important. Some firms wisely sell their scrap at low price to founders of less important castings; others shut their eyes to the problem, renewing stock and changing personnel when crises occur. It should be added that when the financial direction of a firm re-establishes its technical equilibrium by using new metal and putting excess scrap into stock, it can falsify completely the balance sheet of the firm by neglecting to allow for the depreciated quality of the scrap. Financial administrators are inclined to disclaim responsibility for the technical repercussions of their own directions.

Such are the dangers of specialization, to avoid which it is necessary to supplement the principal manufacture of high output ratio by a secondary manufacture having a lower output ratio—for example, the casting of magnesium alloy wheels for aircraft ($M_1 = 4$) together with wheels for trucks in the same material ($M_2 = 1{\cdot}5$).[6]

Influence of External Specifications.

An increase in purity does not necessarily improve quality and may, indeed, reduce it, as was the case with Duralumin during the War, when restriction of the amount of scrap to be used caused a decline in the properties of the metal by reducing the excess of silicon necessary for the formation of Mg_2Si.

A tightening of the acceptance test on the finished product, by increasing the proportion of rejects, raises the factor M, and accelerates the depreciation T.

The Effect of the Scrap Market.

The development of aircraft work requiring a high proportion of virgin metal and having, moreover, a high output ratio, the breaking up of old cars and aeroplanes, and many other sources, have created large floating stocks of scrap, the market value of which has decreased, owing to lack of demand. Later applications, such as the light alloy cylinder head, have given rise to competition, the principal cause of which is the use of such scrap by certain makers. The dangers involved need no emphasis; 10 years ago the use of aluminium for cylinder heads was jeopardized by lack of consistent soundness. In the case of pistons, also, the much superior high-silicon alloys have been replaced to some extent by very cheap alloys, which owing to progress in design give a certain amount of satisfaction to the non-discriminating buyer. The existence of large floating stocks of scrap would thus appear to prejudice the development of serious applications unless research is carried out by producers to permit the use of secondary metal without bringing it into discredit.

The Question of National Defence.

In order to assure a peace-time production, to guarantee adequate capacity for the needs of war, it is necessary to create civil outlets and to organize and distribute the production on a national plan. In the important case of magnesium alloys, for example, in Germany the problem has been attacked by insisting on the use of the material for the wheels of public vehicles.

IV.—Foundry Factors.

In the case of a pure metal or eutectic, the shrinkage is localized in internal cavities which do not usually communicate with the surface : with alloys having a solidification interval the phenomena are more complicated, and cause a succession of defects very different in appearance, but having the same cause. In both cases the problem of the foundry is to assure complete feeding of the still liquid portions of the casting from external sources. This is effected mainly by risers or denseners. The former are liable to reduce the cooling rate seriously, and to make higher casting temperatures necessary, thus reducing the mechanical properties of the alloy. Moreover, they obviously increase the output ratio, and, as a remedy for cavities, may have immediate success, whilst forming the root cause of troubles to come. It is always possible to replace a riser by a suitably placed runner, and it is equally possible to use chills, so reducing the output ratio, M.

Top pouring is to be recommended, since it eliminates at least one riser, but careful technique is required to avoid entrainment of oxide.

It is difficult to distinguish between cavities due to gas and those due to shrinkage, since they can have the same appearance, and are subject to the same laws of progress, division, and localization. In effect the imprisoned gas follows the free path of the remaining liquid metal, exactly as the shrink; they are both stopped in the same way and at the same points, when the metal becomes too viscous. Although the permeability of the sand affects the gas-hole and not the shrinkage cavity, yet the remedies for the latter are usually suitable for the former.

Iron crucibles are a possible cause of trouble in the foundry, as increase of iron content of the metal not only affects the mechanical properties but also increases the solidification interval.

The physical and chemical quality of the moulding sand also introduces many indirect factors, and such unusual occurrences as the pick-up of soot from a neighbouring railway may lead to bad castings.

The question of design is perhaps best approached by imagining the casting to be divided into a large number of thermally differing zones, each of which is considered as a separate problem. A characteristic example is a magnesium alloy aircraft wheel, with heavy rim and thin web. If the latter is made of uniform thickness this cannot be less than 5–6 mm. on account of feeding and internal stress difficulties. It is, therefore, divided into a large number of radial pyramids—see Fig. 1 (Plate I)—giving a thickness, in the thinnest parts, of only 1·5–2 mm. and an average thickness of less than 4 mm. The thick parts feed the thin parts and split up the internal stresses. A similar construction is seen in the railway door of Fig. 2 (Plate II), the average thickness of the panel being less than 3·25 mm.[7]

V.—The Relation of the Use of Alloys to Their Properties.

The Size-Factor.

As has been seen, a feeding head, in suppressing shrinkage, may cause a cavity to appear elsewhere, and castings of more than a certain size, or difficulty, will suffer from a vicious circle of defects, the remedy for one (apparent, direct factor, 1–A) constituting an occasional, temporarily unrecognized factor, 3–B, in the appearance of another. If extreme measures have to be taken for a given casting, the factor of safety (in use) may be reduced by hidden defects. Safety in large castings is more likely to result from using alloys with a small solidification interval, which involves redesigning to allow for poorer mechanical properties, *e.g.* the large monobloc castings in Alpax for marine work, which cannot be safely made in other alloys.

When Alpax was first used in France for cylinder blocks considerable difficulties with porous barrels were experienced, for which the quality of the metal was held responsible (apparent, direct factor, 1–A). The real cause of the trouble was the design (fundamental factor, 3–C), and German founders, using the same alloy, had no trouble, because they designed their blocks to take inserted barrels.

Large monobloc castings for severe duty cannot be made in high-strength alloys having a large solidification interval without prohibitive risks and difficulties. For these alloys monobloc design must be abandoned. Alpax, on the other hand, is unsuitable for designs which require the assembly of a large number of small units, on account of its poor resistance to high local stresses, particularly at right angles to the assembly bolts.

Thermodynamic Factors.

Breakdowns of motors are sometimes attributed to faulty material, when the basic cause is to be found in one of the numerous thermodynamic factors of its functioning.[8] Systematic piston seizure, for example, has at times been traced to the use of inferior fuel. Again, corrosion of magnesium-alloy petrol tanks by additions of tetraethyl lead to the fuel, resulted in decomposition of this anti-detonant, so that, until a chemical remedy was found, the metal of the tank was accused of reducing the possible compression ratio.

On the other hand, the thermodynamic factors are, themselves, affected by the appearance of new alloys. The use of beryllium–copper alloys of high thermal conductivity, for exhaust valves, has enabled these to work at lower temperature, and so increased the life of the engine.

Influence of Modulus of Elasticity.

The strength and elastic limit of light alloys has been increased, whilst the modulus of elasticity has remained almost unchanged. The secondary factor of deformation consequently governs the factor of safety to an increasing extent. For a casting of given design there is a particular value of the elastic limit of the material, which it is useless to exceed.[9]

The importance of the modulus of elasticity was shown in the case of aero engines, having their cylinders in line. These were subject to systematic breaking of crank-shafts, due to deformation of the aluminium-alloy crank-cases. When the latter were made more rigid the trouble ceased, but it reappeared when they were cast from the same patterns in magnesium alloy of different modulus of elasticity.

Effects of Use.

In many constructions the flexibility and natural period of vibration [10] vary with time. Particularly in the case of assembled units, where stress concentrations occur at the joints, the fatigue limit of the material is all important. It must be emphasized that it is not only the parts transmitting power which are subject to fatigue, but also the whole supporting structure—especially when this is not of monobloc design.

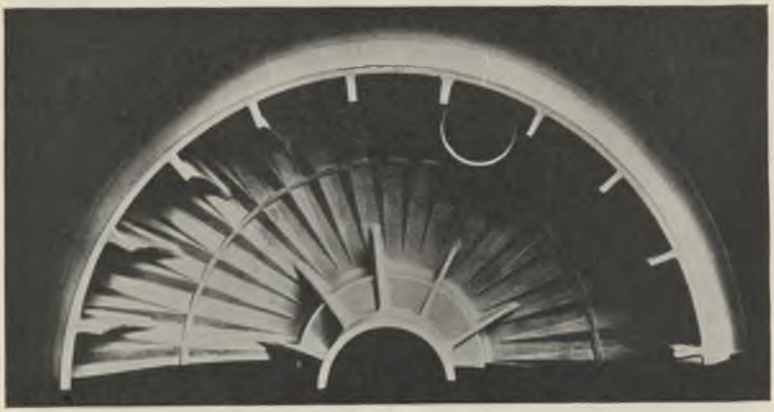

FIG. 1.—Cast Aeroplane Wheel, the Web of which is Composed of Pyramids on a Very Thin Plate, so as to Ensure Adequate Feeding, and to Control and Reduce Internal Stresses.

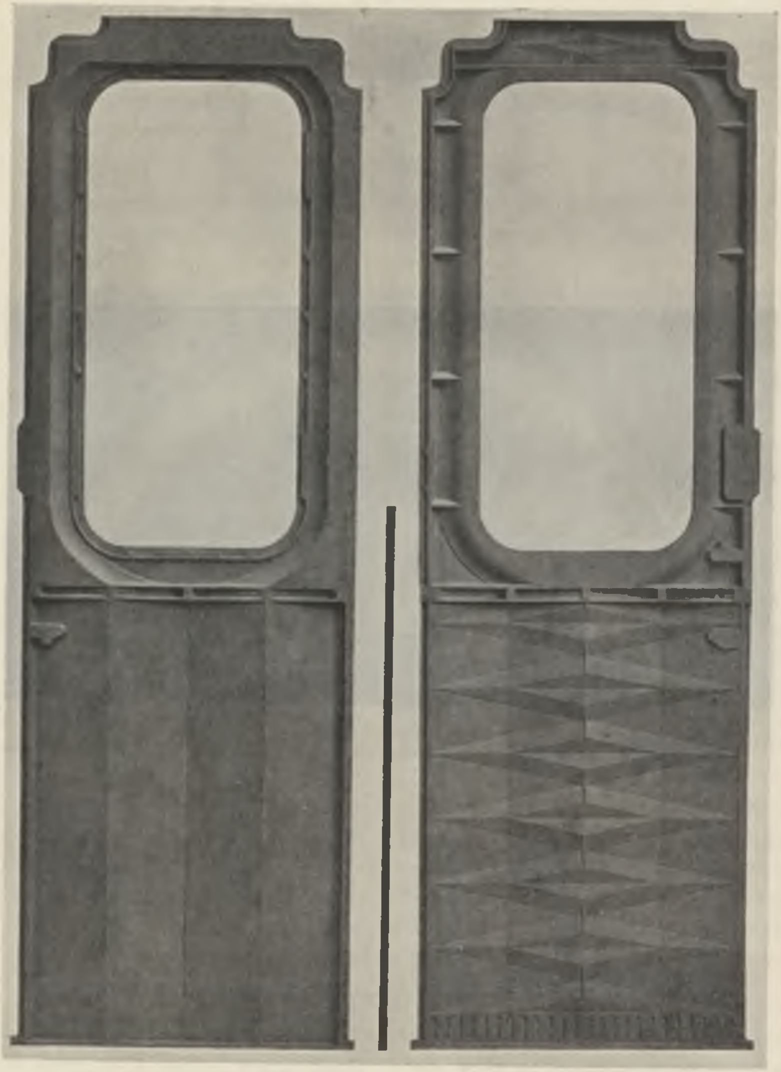

FIG. 2.—Cast Railway Door, the Panel of which is Composed of Pyramids on a Very Thin Plate, in Order to Reduce Founding Difficulties and Increase Rigidity.

CONCLUSIONS.

Some actual examples are given of a number of external factors which, by their subtle and often unsuspected interdependence, have been able to upset the manufacture and functioning of a machine. There are many others.

Metallurgists are warned against the unconsidered application of new alloys, and engineers against hasty condemnation of the materials with which they are supplied. In spite of careful forethought metallurgists and engineers will encounter unsuitable applications which will throw a prematurely advertised material into disfavour. This is the explanation of the amazing set-backs to be remembered in the history of the most logical new developments.

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METALLURGICAL ABSTRACTS

(GENERAL AND NON-FERROUS)

Volume 3 JULY 1936 Part 7

I.—PROPERTIES OF METALS

(Continued from pp. 193–197.)

***Colloid Chemical Phenomena in Metals. I.—Gases in Aluminium.** J. A. Kliachko (*Kolloid-Zeit.*, 1935, **73**, (11), 226–236; and (summary) *Light Metals Research*, 1935, **4**, (19), 314–316; *Sci. Abs.*, 1936, [A], **39**, 9).—The system gas–metal is studied as a problem in colloid chemistry. The general laws which are valid for a gas–metal system are briefly summarized, followed by a description of results of earlier workers on the interaction of gas and aluminium and its alloys. The amount of gas which dissolves in aluminium (chiefly hydrogen) differs according to whether the determination is made by analysis after extraction by heat or by the equilibrium method. Suggestions are made to account for this discrepancy. The origin of the gas is shown to be due to the reaction between the metal and the water vapour adsorbed in its pores.—S. G.

Study of the Manufacture, Properties, and Uses of Refined Aluminium. Robert Gadeau (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 85–90; and *J. Four élect.*, 1936, **45**, (1), 17–22).—[In French.] See *Met. Abs.*, this vol., p. 69.—S. G.

***On the Atomic Weight of Cadmium.** O. Hönigschmid and R. Schlee (*Z. anorg. Chem.*, 1936, **227**, (2), 184–192).—Analysis of the bromide and chloride obtained from metal which had been fractionally distilled *in vacuo* gave the value Cd = 112·41.—A. R. P.

***On the Passivity of Chromium. IV.—The Electrolytic Behaviour of Chromium Amalgam.** Erich Müller (*Z. physikal. Chem.*, 1936, [A], **176**, (4), 273–288).—Compact chromium remains permanently active in *M*-hydrochloric acid at 20° C., but chromium amalgam under the same conditions undergoes a continuous cycle of changes from the active to the passive condition alternately. Both the compact metal and the amalgam are normally passive in *M*-sulphuric and *M*-perchloric acids, but the former can be made permanently and the latter transiently active by feeble cathodic polarization. The amalgam can be rendered active in *M*-sulphuric acid merely by shaking, and the return to the passive state is accompanied by periodic variations, a phenomenon which is not observed with chromium powder. Strong cathodic polarization renders the amalgam passive in all three acids; the activation potential is lower for the amalgam than for the compact metal.—B. Bl.

Thermo-E.m.f., Peltier Effect, and Photo-E.m.f. in the Cell Copper–Cuprous Oxide–Copper. G. Mönch (*Z. tech. Physik*, 1935, **16**, (11), 361–363).—See *Met. Abs.*, this vol., p. 31.—J. S. G. T.

***Peltier Effect in the Element Copper–Cuprous Oxide–Copper.** G. Mönch (*Z. Physik*, 1936, **100**, (5/6), 321–325).—Cf. abstract above. The Peltier effect at a copper–cuprous oxide interface, is, in accordance with theory, found experimentally to be proportional to the absolute temperature of the interface.—J. T.

The Properties and Uses of Lead. R. S. Russell (*Met. Ind.* (*Lond.*), 1936, **48**, (21), 585–590).—From a series of articles in *Modern Engineer*; see *Met. Abs.*, this vol., p. 193.—J. E. N.

***The Effect of Absorbed Hydrogen on the Magnetic Susceptibility of Manganese.** Mary A. Wheeler (*Phys. Rev.*, 1936, [ii], **49**, (8), 642).—Abstract of a

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* Denotes a paper describing the results of original research.

† Denotes a first-class critical review.

paper read before the American Physical Society. Manganese which had been purified by distillation was heated in hydrogen at various temperatures. The susceptibility at room temperature was decreased to $9\cdot28(10)^{-6}$ from a normal value of $9\cdot60(10)^{-6}$ when the absorption took place at 400° C., a temperature at which the α form is stable. When the hydrogen was absorbed in the β form at 850° C. the susceptibility became $9\cdot36(10)^{-6}$. The decreased susceptibility can be due to the adsorption of diamagnetic hydrogen. When the hydrogen was absorbed by the γ form at 1220° C. or by the melted manganese, however, the susceptibility was increased to $10\cdot42(10)^{-6}$ and $10\cdot00(10)^{-6}$. While α- and β-manganese have complicated structures, γ-manganese is a simple face-centred tetragonal form. Thus in γ-manganese the hydrogen is probably inserted interstitially and increases the lattice constant. This could cause a narrowing of the energy bands of the collective electrons and an increase of the susceptibility. No trace of ferromagnetism was found.—S. G.

***Absorption of Gases by Mercury.** S. Ptizin (*Tech. Physics U.S.S.R.*, 1935, **2**, (1), 66–78; *Sci. Abs.*, 1936, [A], **39**, 25).—[In English.] The absorption of gases, whether air or organic vapours, is studied, and the question of freeing mercury from absorbed gases is also dealt with. Hydrogen distilled in a vacuum gives off but little air, 130 mm.³ of mercury giving off only 0·5 mm.³ of gas measured at 1 mm. pressure, even after being stored for some time open to the air, while cleaned mercury only absorbs atmospheric gases for a few hrs. It was also shown that much of the gas liberated comes from the decomposition of organic compounds absorbed in the mercury.—S. G.

***The Density of Mercury at 0°.** T. Batuecas and F.-L. Casado (*J. chim. phys.*, 1936, **33**, 41–50; *C. Abs.*, 1936, **30**, 3287).—About 29 kg. of carefully purified mercury from Spain was used. The density at 0° was determined by the pycnometer method. The average of 22 determinations was $d_4^0 = 13\cdot59539 \pm 0\cdot00001$ grm./ml.—S. G.

***The Diffusion of Gases Through Metals. III.—The Degassing of Nickel and the Diffusion of Carbon Monoxide Through Nickel.** C. J. Smithells and C. E. Ransley (*Proc. Roy. Soc.*, 1936, [A], **155**, (884), 195–212).—Apart from easily detachable surface gas, the gas evolved from commercial nickel by heating *in vacuo* or in hydrogen is mainly carbon monoxide. The measured rate of diffusion of the gas is far too small to account for the rate at which gas is evolved from the metal in the usual degassing processes. It is suggested that the carbon monoxide is generated from uncombined carbon and oxygen contained in the metal. Experiments supporting this hypothesis are described, and it is likely that the rate of evolution of carbon monoxide is controlled by the rate of diffusion of carbon in the metal.—J. S. G. T.

Note on the Curie Point of Nickel. Edmund C. Stoner (*Proc. Leeds Phil. Lit. Soc.*, 1936, **3**, (3), 127–131).—The conventional Curie point is reduced from the results of Weiss and Forrer for nickel. The value of θ (in ° C.) obtained by extrapolation of the steepest part of the σ, T curve is given closely by $\theta = 361\cdot5 + 1\cdot6 \times 10^{-3} H$. Extrapolation of the steepest part of the σ_0^2, T curve gives approximately $\theta = 357\cdot5 + 0\cdot8 \times 10^{-3} H$. The form of the σ_0, T curve giving the variation with temperature of the intrinsic magnetization in zero field (deduced by extrapolation, using the magneto-caloric results) is discussed. It is shown that the experimental results do not enable a unique Curie temperature to be deduced such as is indicated by the idealised theory. S. emphasizes the need for indicating clearly the method used in deducing Curie temperatures when these are given on the basis of experimental results; particularly in connection with possible differences in the Curie temperatures of materials of nearly the same composition.—S. G.

***The Internal Energy of Ferromagnetics [Magnetic Constants (and Specific Heat) of Nickel].** Edmund C. Stoner (*Phil. Trans. Roy. Soc.*, 1936, [A], **235**, (750), 165–193).—Theoretical. Estimates of the specific molecular field coeff.

$N\rho$ for ferromagnetics (N = the formal molecular field coeff., ρ = the density) differ widely not only for different temperatures but for the same temperature when derived from different sets of data. A detailed examination is made of the experimental results for the magnetization, magneto-caloric effect, and specific heat of nickel, and the assumptions of the calculations by the different methods are discussed. The results as a whole indicate values of $N\rho$, not varying appreciably with the specific (intrinsic) magnetization σ, of about $11{\cdot}7 \times 10^4$ above 500° C., and not less than $9{\cdot}5 \times 10^4$ and $7{\cdot}4 \times 10^4$ at 350° C. and room temperature, respectively. Evidence is given for the existence of a magnetic "domain" structure. This usually corresponds with the mosaic structure of metals, but may sometimes be distinct and either larger or smaller. —W. H.-R.

***The Change of Magnetization of Nickel, Permalloy, and Nickel Single Crystals in Strong Fields.** Gerd Gerloff (*Z. Physik*, 1936, **99**, (9/10), 585–594).—The magnetic susceptibility of nickel and of Permalloy attains, at room temperature, a constant value in fields of strength about 4000 Oerstedt, and this value changes by less than 6% for fields of strength up to 5500 Oerstedt.—J. S. G. T.

***The Freezing Point of Palladium.** F. H. Schofield (*Proc. Roy. Soc.*, 1936, [A], **155**, (885), 301–308).—The freezing point of palladium, determined from measurements of brightness ratio, for a certain wave-length of black body radiators held at the freezing points of palladium and gold, is found to be 1554·4° ± 1° C. on the International Temperature Scale. The value recommended in the specification of that scale is 1555° C.—J. S. G. T.

***Paramagnetism of Platinum Colloidal Powders.** Noboro Takatori (*Kinzoku no Kenkyu* (*J. Study Metals*), 1936, **13**, (5), 161–172).—[In Japanese.] Measurements were made of the paramagnetic susceptibilities of platinum colloidal particles and of those which had been annealed at 800°–1000° C. It was found that the specific susceptibility of platinum colloidal particles with diameters of $2\text{–}3{\cdot}10^{-5}$ mm. decreases by 21% as compared with those which have been annealed and by 29% as compared with the massive state. The magnetic susceptibility of very minute platinum particles which were observable only with high magnifications was also studied; the susceptibility decreases with size.—S. G.

***Alkali Films of Atomic Thickness on Platinum.** Herbert Mayer (*Z. tech. Physik*, 1935, **16**, (11), 451–454).—See *Met. Abs.*, this vol., p. 32.—J. S. G. T.

***The Theoretical Constitution of Metallic Potassium.** E. Gorin (*Physikal. Z. Sowjetunion*, 1936, **9**, (4), 328–344).—[In English.] A theoretical discussion of the constitution of metallic potassium, based on the theories of Wigner, Seitz, and Hartree, and on wave mechanics is presented. The final results are that the binding energy and lattice constants are, respectively, 20·4 kg. cal. and 4·80 A., compared with the experimental values 22·6 kg. cal. and 5·20 A.—J. S. G. T.

***The Magnetochemistry of Rhenium: Metallic Rhenium and Septivalent Rhenium.** N. Perrakis and L. Capatos (*Praktika* (*Akad. Athenon*), 1934, **9**, 121–125; *C. Abs.*, 1936, **30**, 3289).—[In Greek.] Cf. *Met. Abs.*, 1934, **1**, 551. Metallic rhenium and the septivalent compounds of the element, *viz.* Re_2O_7, $KReO_4$, NH_4ReO_4, were studied for paramagnetism. Septivalent rhenium possesses a paramagnetism independent of the temperature, $\chi = 0{\cdot}04 \times 10^{-6}$ per grm. as compared with $0{\cdot}046 \times 10^{-6}$ for metallic rhenium, and $0{\cdot}04 \times 10^{-6}$ for osmium. Reduction with hydrogen gives a very pure rhenium from the ammonium salt, with a value of $\chi = 0{\cdot}037 \times 10^{-6}$ per grm.—S. G.

***The Atomic Weight of Rubidium.** E. H. Archibald, J. Gilbert Hooley, and Norman Phillips (*J. Amer. Chem. Soc.*, 1936, **58**, (1), 70–72).—Re-investigation of the atomic weight of rubidium by determination of the silver–rubidium chloride ratio gave a slightly higher value for this constant (85·482) than that accepted by the International Committee. Cf. following abstract.—S. G.

***The Atomic Weight of Rubidium.** E. H. Archibald and J. Gilbert Hooley (*J. Amer. Chem. Soc.*, 1936, **58**, (4), 618–619).—Cf. preceding abstract. The nephelometric determination of the rubidium bromide–silver ratios gave a value of 85·483 for the atomic weight of rubidium and the gravimetric estimation of the ratios rubidium bromide–silver bromide gave 85·478. When these values are combined with that obtained from the rubidium chloride analyses, the average value 85·481 is obtained.—S. G.

***Germicidal Properties of Silver in Water.** J. Just and A. Szniolis (*J. Amer. Water Works Assoc.*, 1936, **28**, (4), 492–506).—The addition of silver to water in amounts between 25 and 600 γ per litre destroys bacteria, microfauna and microflora, and acts within a short time on organisms such as daphnia and fishes. The action is identical whether the silver is added as a soluble salt or is dissolved by electrolysis by the " electrocatadynization " process. The amount of silver dissolved in this process is shown to depend not only on the current but also on the mineral content of the water. Experiments showed that *B. Coli* added to tap-water were killed within 2–3 hrs. by a dose of 100 γ of silver per litre, but that it required 5–7 hrs. when the same number of *B. Coli* were added to polluted water. Rats fed for 100 days with water containing up to 1000 γ of silver per litre showed no obvious symptoms of poisoning, but pathological changes were found in liver, kidney, and spleen. Constant use of water disinfected by silver thus cannot be considered quite innocuous. For determining (with an accuracy of 2·5 γ) small quantities of silver in water, 2–10 drops of 1 : 1 hydrochloric acid and 2 drops of 1 : 10 potassium iodide are added to 100 ml., the solution filtered through a membrane filter in a Kolkwitz filtration apparatus, and the filter moistened with H_2S water. The disc is dried and its colour compared with standards. *A bibliography* of 19 references is appended.
—J. C. C.

***Transition of a Monocrystalline Tin Sphere from the Superconductive into the Non-Superconductive State.** W. J. de Haas and O. A. Guinau (*Physica*, 1936, **3**, (3), 182–192; and *K. Onnes Lab. Leiden Comm.* No. **241a**; *Sci. Abs.*, 1936, [A], **39**, 416).—[In English.] The study of the penetration of a magnetic field into a superconductive tin cylinder (*Met. Abs.*, 1935, **2**, 277) is extended to monocrystalline spheres, for which the demagnetization factor is known. A magnetic field was applied to the sphere and its intensity increased at constant temperature until superconductivity was entirely disturbed. It was concluded that the apparent permeability of the sphere is zero so long as the applied field strength is lower than $\frac{2}{3}$ of the field which disturbs superconductivity entirely. The permeability increases linearly until it reaches the value unity when the applied field attains its critical value. The transition was also studied in a constant external field while the temperature was being increased.—S. G.

***The Thermal Conductivity of Tungsten.** Walter C. Michels and Martha Cox (*Physics*, 1936, **7**, (4), 152–155).—The theory of the heat losses from an electrically-heated wire, as given by Roberts and Kannuluik, has been extended to include a wire supported by springs. The results obtained were used in a measurement of the thermal conductivity of commercial tungsten wire in the range 78°–273° K. There are definite indications that the Wiedemann–Franz law breaks down badly in this region.—S. G.

***Thermionic Emission from Barium-Coated Tungsten.** L. N. Dobretzow and G. A. Morozow (*Physikal. Z. Sowjetunion*, 1936, **9**, (4), 352–361).—[In English.] The emission of electrons and positive ions from tungsten in a stream of barium vapour is investigated. The emission of electrons depends on temperature in the same manner as that from tungsten in alkali vapours. The presence of a positive-ion current produced by surface ionization of barium on tungsten is established. The heat of evaporation of barium atoms from tungsten is found to be 151,000 ± 10,000 cal./mol. for the range of activation (as defined by Langmuir) $\theta = 0{\cdot}3$–$0{\cdot}7$ and decreases with increasing θ.—J. S. G. T.

The Dissociation Equilibrium of Hydrogen and Its Adsorption on Tungsten. J. K. Roberts (*Proc. Cambridge Phil. Soc.*, 1936, **32**, (1), 152–157).—The various processes occurring at the surface when hydrogen is adsorbed on tungsten are considered together with the dissociation equilibrium of hydrogen in the gas phase. The form of the adsorption isotherm is deduced from the principle of detailed balancing, and is in agreement with that obtained by Fowler using a statistical method. A detailed interpretation of the experimental results now available shows that either (*a*) measurements of the rate of removal of the adsorbed film of oxygen on tungsten do not measure the rate of evaporation of oxygen atoms; or (*b*) it is not possible to obtain a general first approximation formula giving the rate of evaporation of adsorbed atoms in terms of the heat of desorption. The desorption of hydrogen from tungsten is discussed, and it is shown that the agreement between the temperature at which the film evaporates at an appreciable rate and that deduced from a desorption formula of the type mentioned in (*b*) assuming that the hydrogen evaporates as atoms must at present be regarded as a coincidence.—S. G.

***Cæsium–Oxygen Films on Tungsten.** J. H. Lees (*Phil. Mag.*, 1936, [vii], **21**, (144), 1131–1139).—The mode of formation of the oxygen layer is investigated, the effects of pressure, temperature and time of activation, time for adsorption being examined.—J. S. G. T.

***Disintegration of a Monatomic Layer of Thorium on a Tungsten Cathode in a Mercury Discharge.** N. Morgulis and M. Bernadiner (*Tech. Physics U.S.S.R.*, 1935, **2**, (4), 333–352).—[In English.]—S. G.

***The Dependence of Cathode Sputtering on Temperature [Tungsten].** N. D. Morgulis, M. P. Bernardiner, and A. M. Patiocha (*Physikal. Z. Sowjetunion*, 1936, **9**, (4), 302–316).—[In German.] It is shown experimentally that the critical potential for the sputtering of thoriated tungsten by mercury ions at 1350°–1750° K. and the amount of sputtering of pure tungsten by mercury ions at 800°–2000° K. are practically independent of the anode temperature.—J. T.

***On the Creep of Zinc.** N. N. Davidenkov and I. N. Mirolubov (*Zhurnal Tehnicheskoy Fiziki* (*J. Tech. Physics*), 1936, **6**, (1), 60–77).—[In Russian.] An apparatus is described by means of which specimens of zinc single crystals can be stretched at a constant rate of $5{\cdot}10^{-3}$ mm./hr. Stress–flow diagrams for several zinc single crystals were found to be identical with those obtained at usual rates, both as regards the limit of flow and the occurrence of discontinuities. An explanation of these discontinuities is advanced, and their occurrence at constant load is compared with that at constant rate of deformation.—N. A.

***The Free Energies and Vapour Pressures of the Alkali Metals.** A. R. Gordon (*J. Chem. Physics*, 1936, **4**, (2), 100–102).—The free energies of the diatomic alkali vapours, K_2, Na_2, and Li_2 are calculated from spectroscopic data, and the equilibrium constants for the dissociation into the monatomic form are computed. With the aid of these constants, equations are deduced from the observed vapour pressures obtained by other investigators, which give the partial pressures of atoms and molecules in the saturated vapours for temperatures up to the normal boiling point. In the cases of sodium and potassium the entropy of the solid metal is computed from the vapour pressure equation, and is compared with that predicted by the third law of thermodynamics.—S. G.

The Atomic Frequencies of the Alkali Metals. Binayendra Nath Sen (*Gazz. chim. ital.*, 1935, **65**, 907–908; *C. Abs.*, 1936, **30**, 3288).—The formula of S. for calculating the vibration frequencies of atoms (cf. *J. Indian Chem. Soc.*, 1934, **11**, 243) gave results close to the experimental values and to those obtained by other formulæ, except for sulphur and selenium. This might be explained by the allotropy of sulphur and selenium or by their photoelectric properties, especially with selenium. Corresponding calculations with the new formula of other elements with photoelectric properties, *viz.* sodium, potassium, lithium, rubidium, and cæsium, give results which agree with those obtained by other formulæ (except for lithium). Accordingly the discordances in the calculated atomic frequencies of selenium and sulphur are attributable to their allotropy rather than to their photoelectric properties, which sulphur possesses to an insignificant degree.—S. G.

***Photo-Effects of Thin Adsorbed Films of the Alkali Metals.** V. Gey and J. Truten (*Zhurnal eksperimentalnoy i teoreticheskoy Fiziki* (*J. Exper. and Theoret. Physics*), 1936, **6**, (1), 30–36).—[In Russian.] By adsorption of the vapours of alkali metals, films with a thickness of several atomic layers have been obtained. The electrical conductivity of these films obeys Ohms' law. The relative photo-current curves obtained from them do not reveal selective maxima.—N. A.

Ultimate Strength of Metals. E. N. da C. Andrade (*Science Progress*, 1936, **30**, (120), 593–610; *C. Abs.*, 1936, **30**, 3757).—A consideration of the fundamental problems of metallic strength and of yield under stress, with 28 references.—S. G.

First Report on Viscosity and Plasticity. Committee for the Study of Viscosity of the Academy of Sciences, Amsterdam (*Verhandel. Akad. Wet. Amsterdam, Afdeel. Natuurkunde*, 1st Sect., 1935, **15**, (3), 256 pp.; *C. Abs.*, 1936, **30**, 3692).—The basic mechanical relations, the experimental investigation of flow properties, the measurement of viscosity, viscosity and plasticity from a technical point of view, the plasticity of metals and other crystalline substances, &c., are discussed.—S. G.

The Deformation, Recovery, and Recrystallization Behaviour of Metals. R. H. Hobrock (*J. Aeronaut. Sci.*, 1936, **3**, (6), 191–199).—The behaviour of single-crystal structures under mechanical stress is contrasted with that of the finely polycrystalline materials in ordinary technical use. " Recovery " from work-hardening is distinguished from " temper-relief," which is defined as a heating and cooling operation intended to bring the properties of an alloy as near as possible to those of its predominating metallic constituent. This can be effected without achieving " recovery," or the relief of stresses in the lattice. The mechanism of grain-growth is discussed, and the three-dimensional recovery–recrystallization diagram of an alloy of the Duralumin type is reproduced. The annealing temperature necessary for the relief of internal stresses may come within the recrystallization range, and grain-growth may thus be set up: some means of prevention are suggested.—P. M. C. R.

***The Thermal Stress in a Strip Due to Variation of Temperature Along the Length and Through the Thickness.** (Goodier.) See p. 278.

***On the Sticking of Two Metal Surfaces to One Another in a Vacuum and the Reduction of Sticking by Various Gases.** Ragnar Holm and Bernhard Kirschstein (*Wiss. Veröff. Siemens-Werken*, 1936, **15**, (1), 122–127).—Cylinders of nickel, platinum, and graphite with a small axial hole drilled in them were threaded on stout nickel or platinum wires to give a very loose fit, and the angle at which slip occurred was determined *in vacuo* and in various gas atmospheres after previous annealing *in vacuo* to remove all trace of surface films. The results show that *in vacuo* and in an inert gas atmosphere (nitrogen or argon) severe sticking of the metals to one another occurs, so that no slip takes place even when the wire is vertical unless it is shaken; this is attributed to a kind of cold-welding by simple pressure and occurs only with perfectly clean surfaces. In hydrogen, steam, air, or the vapour of hydrocarbons slip occurs when the wire is inclined at 15°–62° to the horizontal according to the metals in contact and the atmosphere; this effect is attributed to the formation of invisible films on the metal surface which allow slip to occur. The presence of such films accounts for the contact resistance in electric switchgear, &c.
—A. R. P.

***Experiments on Two Thin Metal Films Pressed Together.** Alexander Deubner (*Naturwiss.*, 1935, **23**, 557).—The non-linear relationship between the physical properties and the thickness of the metal films cannot be explained by the fact that the real film thickness is not identical with the theoretical value derived from the weight of the deposited metal, but the phenomenon is an actual property of thin metal films.—B. Bl.

***Diffusibility of Deuterium in Metals.** T. Franzini (*Nuovo cimento*, 1936, **13**, (2), 74–78; *Sci. Abs.*, 1936, [A], **39**, 580).—Experiments were made to determine whether deuterium diffuses through metals, palladium being used as it is known that hydrogen is occluded by this metal. The presence of hydrogen and deuterium was determined spectroscopically. The work showed that occluded hydrogen is removed by an electric field, but no evidence for this was obtained in the case of deuterium.—S. G.

Adsorption Isotherms. Critical Conditions. R. H. Fowler (*Proc. Cambridge Phil. Soc.*, 1936, **32**, (1), 144–151).—S. G.

Sixth Report of the Committee on Atomic Weights of the International Union of Chemistry. G. P. Baxter, O. Hönigschmid, and P. Lebeau (*J. Amer. Chem. Soc.*, 1936, **58**, (4), 541–548).—See *Met. Abs.*, this vol., p. 144.—S. G.

*†**The Measurement of Contact Potential Difference.** C. W. Oatley (*Proc. Roy. Soc.*, 1936, [A], **155**, (885), 218–234).—Methods of measuring contact potential differences hitherto used are critically reviewed and a new method is described, dependent on the known relation between anode voltage and critical magnetic field in a magnetron, when electrons from the filament just fail to reach the anode. Results relating to the contact potential differences of molybdenum, zinc, and platinum relative to a hot tungsten filament are briefly discussed.—J. S. G. T.

***Influence of Electron Reflection on Photoelectric Emission.** W. B. Nottingham (*Phys. Rev.*, 1936, [ii], **49**, (8), 646).—Abstract of a paper read before the American Physical Society. Electrons emitted thermionically from tungsten and thoriated tungsten filaments are distributed in energy as though they suffered a reflection at the barrier given by $R(p_x) = \exp(-p_x^2/2m\omega)$, where $\omega = 3{\cdot}05 \times 10^{-13}$ erg and $p_x =$ momentum in excess of that required to go over the barrier. Du Bridge and others have shown that Fowler's photoelectric theory which assumes reflection constant (or zero) fits the experimental data. This reflection may be incorporated into the theory. The new function is nearly identical in form with the Fowler curve over the experimental range. If the reflection hypothesis be assumed to be correct, all previous determinations of photoelectric work-functions are too high by an amount given approximately by $\Delta\phi = 8 + 0{\cdot}138T - 5 \times 10^{-5}T^2$ mv. Whereas the Fowler analysis gives work-functions for clean surfaces practically independent of temperature, the new method when applied to Du Bridge's data on palladium yields a *negative* temperature coeff. of the work-function of $(4{\cdot}5 \pm 1) \times 10^{-5}$ v. per degree. This is consistent with a thermionic constant A of 60 amp./cm.^{2}/degree^2, if it be assumed that only 40% of the apparent surface emits, as is thought to be the case for pure tungsten, and includes the effect of reflection. —S. G.

On the Theory of the Photoelectromotive Force in Semi-Conductors. L. Landau and E. Lifshitz (*Physikal. Z. Sowjetunion*, 1936, **9**, (5), 477–503).—[In English.] The e.m.f. that appears in a circuit containing a semi-conductor illuminated from one side is calculated. Two cases are considered: a semi-conductor with conducting electrons and a semi-conductor with conducting electrons and " holes."—S. G.

***Action of Periodically Varying Light on a Metallic Plate.** Q. Majorana (*Atti R. Acad. Lincei* (*Roma*), 1933, **17**, 255–256; *Sci. Abs.*, 1933, [A], **36**, 795).—Deals with the action of interrupted light on a lamp-blackened disc, which may be partly due to heat, and describes a new method for compensating the photoelectric action. (See also following abstracts.)—S. G.

***Action of Light on Thin Metallic Laminæ.** Q. Majorana (*Nuovo cimento*, 1933, **10**, (4), 261–285; *Sci. Abs.*, 1934, [A], **37**, 212).—Continuing previous work (see preceding abstract) it is found that the electrical resistance of a thin lamina of platinum, silver, gold, and tin is increased when the lamina is strongly illuminated. The effect is very small with aluminium and zinc and is non-existent with sodium. Besides varying with the nature of the metal, the magnitude of the effect depends on the thickness of the lamina and the frequency of interruption of the light. If the lamina is immersed in running water the effect is unaltered, but in still water it is sensibly reduced. There is always a phase lag in the effect of about 45° when pulsating light is used; in some circumstances this may be reduced to about 35°. The experiments seem to show the existence of a new type of direct action of light on electrical resistance. Unlike the photoelectric effect, it is not manifested immediately the light is incident, nor can it be explained away merely on purely thermal grounds.—S. G.

***Photo-Resistance of Metals.** Q. Majorana (*Nuovo cimento*, 1935, **12**, (7), 409–417; *Sci. Abs.*, 1936, [A], **39**, 78).—An account is given of further investigations of the metallic photo-resistance found by M. (see preceding abstract). After reference to facts which distinguish it from a purely thermal action of light on the resistance of metals, new determinations of the phase lag ϕ between pulsating light acting on the metallic lamina and the resistance of the latter are described. It is found that for metallic laminæ fixed on glass ϕ amounts to about 45°, but may in exceptional cases be as low as 16°; whilst for laminæ free in air it may be very nearly 90°, but sometimes only 65°. Moreover, ϕ increases on decrease of the frequency of the light, and this variation is strongest for light from a mercury lamp. The results confirm the supposition of the existence of a new photoelectric effect that, in contrast with the ordinary photoelectric effect, is in its nature rather slow in its manifestation.—S. G.

***Electric Equilibrium and Permanent Thermoelectric Currents in a Metallic Conductor.** F. Odone (*Nuovo cimento*, 1935, **12**, (8), 522–530; *Sci. Abs.*, 1936, [A], **39**, 207).—The thermodynamic theory of phenomena relating to electric equilibrium and the passage of permanent currents in a metallic conductor is considered. Using the idea of an electrochemical potential, a parameter introduced by Duhem to explain the action between electric charges and material masses, is shown the existence of a potential difference between the interior and the surface of a metal, of an internal and external Volta effect. The e.m.f. of a thermoelectric current and the coeffs. of the Peltier and Thomson effects in terms of the e.m.f. of the thermocouple are calculated. It is found that the well-known theory due to Kelvin is based on contradictory hypotheses and cannot be accepted.—S. G.

***Permanent Thermoelectric Currents in Metallic Conductors.** F. Odone (*Nuovo cimento*, 1936, **13**, (1), 11–15; *Sci. Abs.*, 1936, [A], **39**, 567).—Continuing previous work (preceding abstract) it is shown that the thermal phenomena that are produced in a metallic circuit in which a permanent current is flowing do not modify the free energy in the conductor, but cause instead a modification of the energy in the surroundings, with which the conductor itself exchanges energy.—S. G.

***Further Experiments on Metallic Conduction.** Ernst Weber (*Phys. Rev.*, 1936, [ii], **49**, (8), 643).—Abstract of a paper read before the American Physical Society. The experiments previously described (*Phys. Rev.*, 1933, [ii], **44**, 318; and *Met. Abs.*, 1934, **1**, 560) have been continued. The distribution of large d.c. in a flat copper conductor in a strong magnetic field perpendicular to the flow of current was investigated, and its relation sought with the experiments of Kapitza on the increase of resistance in a magnetic field (*Note:* No details are given in the abstract).—S. G.

Incandescence Electron Emission and Electron Conduction in the Case of Solids. A. Gehrts (*Z. tech. Physik*, 1935, **16**, (11), 370–373).—See *Met. Abs.*, this vol., p. 34.—S. G.

***Does Metallic Conduction Occur in Highly Compressed Metallic Vapours?** Werner Braunbek (*Z. tech. Physik*, 1935, **16**, (11), 497–498).—See *Met. Abs.*, this vol., p. 34.—J. S. G. T.

The Change of the Electrical Resistance and of the Reflecting Power of Metallic Mirrors Condensed at Low Temperatures. R. Suhrmann and G. Barth (*Z. tech. Physik*, 1935, **16**, (11), 447–451).—See *Met. Abs.*, this vol., p. 34.—S. G.

***The [Electrical] Resistance of Very Thin Layers of Impurities in Metallic Contacts.** R. Holm and B. Kirschstein (*Z. tech. Physik*, 1935, **16**, (11), 488–494).—See *Met. Abs.*, this vol., p. 34.—J. S. G. T.

Relaxation Phenomena in the Transition from the Superconductive into the Non-Superconductive State. W. H. Keesom and P. H. van Laer (*Physica*, 1936, **3**, (3), 173–181; and *K. Onnes Lab. Leiden Comm.* No. **240c**).—[In English.] The transition from the superconductive to the non-superconductive state takes place, as revealed by experiments on tin, with a relaxation of several seconds, if it is caused by an increase of temperature of the substance, while a constant magnetic field is applied. No such relaxation (at least not to an amount of seconds) occurs if the transition is the consequence of an increase of the external magnetic field. These experimental facts are in harmony with the views that, the threshold value curve being reached, there is a weak coupling between the atomic lattice and the system of superconductive electrons, and that there is a strong coupling between the external magnetic field and the system of superconductive electrons as well as between the non-superconductive electrons and the atomic lattice.—S. G.

***An Experimental Examination of the Electrostatic Behaviour of Superconductors.** H. London (*Proc. Roy. Soc.*, 1936, [A], **155**, (884), 102–110).—It is shown experimentally that the lines of electric induction in a superconductor terminate discontinuously in surface charges and do not penetrate a thin layer of the superconductor.—J. S. G. T.

Electron Configuration in the Superconducting Metals. U. Dehlinger (*Z. tech. Physik*, 1935, **16**, (11), 498–499).—See *Met. Abs.*, this vol., p. 35.—J. T.

†Investigations in Strong Magnetic Fields. O. v. Auwers (*Naturwiss.*, 1936, **24**, (5), 65–73).—Reviews recent work of Kapitza on the magnetic behaviour of nickel, manganese, bismuth, gallium, tin, beryllium, magnesium, and tungsten in strong magnetic fields.—B. Bl.

***An Elementary Theory of the Change of Resistance in a Longitudinal Magnetic Field.** A. Sommerfeld and B. W. Bartlett (*Z. tech. Physik*, 1935, **16**, (11), 500).—See *Met. Abs.*, this vol., p. 35.—J. S. G. T.

Estimation of the Maximum Temperature in the Neighbourhood of a Cross-Sectional Constriction of an Incandescent Wire in an Inert Gas. Leopold Prásnik (*Z. Physik*, 1936, **99**, (9/10), 710–713).—The analysis already given for an incandescent wire *in vacuo* (*Z. Physik*, 1933, **86**, 253) is here extended to the case of an incandescent wire having a cross-sectional constriction and heated in an inert gas. The results are compared with data derived from the filaments of gas-filled lamps.—J. S. G. T.

***Theory of Metallic Linking.—I.** Paul Gombás (*Z. Physik*, 1936, **99**, (11/12), 729–742).—A mathematical theory of interatomic forces in metals, employing Hartree's eigen-functions, is developed and applied to derive values of the physical constants of the potassium atom. The following values of the respective constants are derived; for comparison the experimental values are given in brackets: lattice constant, 5·37 A. (5·15 A.); lattice energy, 111·1 k.cal./mol. (126·2 k.cal./mol.); heat of sublimation, 11·4 k.cal./mol. (26·5 k.cal./mol.); compressibility, $2·6 \times 10^{-11}$ cm.^{2}/dyne ($2·0 \times 10^{-11}$ cm.^{2}/dyne).
—J. S. G. T.

II.—PROPERTIES OF ALLOYS

(Continued from pp. 198–202.)

Structural Aluminium and Fatigue. F. H. Frankland: Henry D. Johnson. R. L. Templin (*Eng. News-Record*, 1936, **116**, (20), 709–710).—Cf. *Met. Abs.*, this vol., p. 133. In a letter, F. suggests that aluminium alloys are particularly susceptible to reduction of fatigue-strength by notch effects, and directs attention to their low endurance limit. Replies by J. and T. include data on the fatigue limits of "27 ST" alloy at 10 × 10^6 cycles (19,000 lb./in.^{2}) and 500 × 10^6 cycles (13,000 lb./in.^{2}). Attention is directed to the advantages of this alloy's low modulus of elasticity and to the value of precise knowledge of the properties of materials in engineering design.—J. C. C.

The Endurance Limit of Aluminium Alloys. R. Irmann (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 435–440; and *Rev. Mét.*, 1936, **33**, (4), 231–236).—[In French.] See *Met. Abs.*, this vol., p. 113.—S. G.

***Magnetic Investigation of Precipitation-Hardening [of Aluminium–Copper Alloys].** Hermann Auer (*Z. tech. Physik*, 1935, **16**, (11), 486–488).—See *Met. Abs.*, this vol., p. 36.—J. S. G. T.

†Improvement of Aluminium Alloys. M. I. Zaharova (*Metallurg* (*Metallurgist*), **1936**, (2), 92–101).—[In Russian.] A survey of the ageing of aluminium–copper alloys, aluminium–magnesium–silicon alloys, Duralumin, and aluminium–silicon alloys, and of the influence of plastic deformation on the disintegration of solid solutions.—N. A.

***Aluminium–Magnesium Alloys.** Pierre Vachet (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 175–187; also (in English) *Light Metals Research*, 1936, **4**, (23), 411–414; and *Aluminium and Non-Ferrous Rev.*, 1936, **1**, (7), 301–309).—[In French.] See *Met. Abs.*, this vol., p. 73.—S. G.

A Quaternary Aluminium–Silicon Alloy. Development and Properties of Cetal. G. Welter (*Met. Ind.* (*Lond.*), 1936, **48**, (23), 627–634).—An account of the properties of an alloy consisting essentially of copper 3, zinc 10, silicon 6·5%, aluminium remainder. Comparisons are made with the usual standard alloys (see *J. Inst. Metals*, 1926, **36**, 325).—J. E. N.

†Beryllium. Robert Gadeau (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 189–199; and *Aluminium and Non-Ferrous Rev.*, 1936, **1**, (8), 353–360.)—[In French.] See *Met. Abs.*, this vol., p. 74.—S. G.

***The Electrical Resistance of Bismuth Alloys.** N. Thompson (*Proc. Roy. Soc.*, 1936, [A], **155**, (884), 111–123).—The electrical resistances of single crystals of bismuth containing small proportions of lead, tin, germanium, selenium, and tellurium were determined at 14°–400° K. Lead, tin, and germanium dissolve in the bismuth lattice and produce an alloy having a large negative temperature coeff. of resistance parallel to the principal axis of the crystal. The effect is similar, but small, perpendicular to the axis. Selenium and tellurium dissolve in bismuth, reducing its specific resistance both parallel and perpendicular to the axis except at low temperatures. They give no negative temperature coefficient. The results are explained qualitatively in terms of Jones's theory of bismuth.—J. S. G. T.

***On the Magnetic Behaviour of the System Chromium–Sulphur.** Haakon Haraldsen and Anna Neuber (*Naturwiss.*, 1936, **24**, (18), 280).—Susceptibility measurements in the range $CrS–Cr_2S_3$ show that within a small concentration range the system becomes markedly ferromagnetic but at 165° K. an unexpected sharp decrease in the magnetism occurs such as is not found in other ferromagnetic materials at temperatures below the Curie point.—B. Bl.

***On the Lower Sulphides of Cobalt. The Equilibrium Diagram of the System Cobalt–Cobalt Sulphide.** Otto Hulsmann and Friedrich Weibke (*Z. anorg. Chem.*, 1936, **227**, (2), 113–123).—Thermal, micrographic, and X-ray examination of the system cobalt–cobalt sulphide shows that the solubility of sulphur in cobalt is almost zero at room temperature. The eutectic occurs at 874° C. and 26·4% sulphur. The β-phase (Co_4S_3) has a homogeneous range of 1·3% cobalt and is stable only above 787° C., below which it decomposes into the cobalt-γ eutectoid. The γ-phase, which has a face-centred cubic structure is formed by a peritectic reaction at 928° C. and has the composition Co_6S_5. The δ-phase is homogeneous only at the composition $CoS_{1\cdot05}$.—A. R. P.

Definite Compounds in Alloys [Copper–Aluminium]. E. Amic (*Rev. Fonderie moderne*, **1936**, (Feb. 2), 71).—R. Thews (*ibid.*, **1935**, (Apr. 25)), stated that in preparing binary master alloys of copper and aluminium, an increase in temperature of 280°–330° C. occurs, " which has not been completely explained." A. states that this exothermic reaction is due entirely to the successive formation of the intermetallic compounds $CuAl_2$ and CuAl. [It has long been proved that the heat comes from reduction of the oxide in the copper by aluminium.—*Ed.*]—J. E. N.

***The Effect of Iron Impurities on the Annealing of High Brass.** (Gibson and Doss.) See p. 274.

***The Recrystallization of Copper–Zinc Alloys with Zinc Contents up to 7 Per Cent.** Al. Kitaigorodski (*Tech. Physics U.S.S.R.*, 1936, 3, (1), 39–50; *Sci. Abs.*, 1936, [A], **39**, 579).—[In German.] Alloys were prepared containing up to 7% zinc by melting copper and zinc together (*a*) in purified nitrogen, and (*b*) under charcoal in air, with subsequent deoxidation. Recrystallization was determined on a 97–98% cold-worked strip by an X-ray method. At about 2% zinc the rate is a minimum as shown both by time and by temperature curves. The speed is not affected by 0·2–0·4% phosphorus nor by cuprous oxide which markedly inhibits recrystallization in pure copper. The oxide is apparently immediately reduced by zinc. Theoretically, recrystallization depends on the number of nuclei and rate of growth, so that it is unnecessary to assume the formation of a special constituent at 2% zinc. Theoretical investigation of recrystallization is very difficult. The practical rule (recrystallization temperature/melting point) = const. has no theoretical or experimental basis for the alloys investigated.—S. G.

***Influence of Impurities on the Properties of Magnesium Alloys.** N. W. Ageew, M. I. Zamotorin, and D. N. Shoyket (*Metallurg* (*Metallurgist*), **1936**, (3), 27–38; (4), 48–59).—[In Russian.] The influence was studied of silicon (up to 2%), iron (up to 0·18%), nitrogen (up to 1%), sodium (up to 0·98%), potassium (up to 0·72%), calcium (up to 0·99%), magnesium oxide (up to 1·0%), and chlorides (up to 0·51% Cl) on cast Elektron alloys containing aluminium 6–10, manganese 0–0·18, zinc 0 and 2·98, and copper 0 and 1·04%, and on rolled alloys containing aluminium 1·5–6, zinc 1–4·5, manganese 0–0·2, and copper 0–0·75%. Silicon causes strong absorption of gases, increase in porosity, greater tendency to segregation, increased contraction, greater hardness and brittleness, lower resistance to impact, and smaller elongation and reduction in area, but the change in mechanical properties with temperature is less the higher the % silicon, and there is a greater tendency to age-harden. Iron dissolves up to 0·2% reducing the contraction and increasing the hardness and tensile strength; it has no effect on ageing or rolling. Nitrogen reduces contraction, increases porosity, and lowers the mechanical properties at all temperatures. Sodium intensifies liquation, increases porosity, causes the formation of cracks in the liquation zone, and seriously affects the mechanical properties, the plasticity disappearing entirely with 0·9% sodium at room temperature and with only 0·3% sodium at 150°–300° C., while the tensile strength, elongation, and reduction in area decrease at high temperatures. Even as little as 0·09% sodium prevents hot extrusion of Elektron. Potassium has a similar effect to sodium, but does not impede rolling. Calcium is a good deoxidizer for magnesium and its alloys. Cast Elektron alloys even with 0·3% calcium have a clean surface free from oxide and nitrides. Calcium increases the density and, up to 1%, does not affect the mechanical properties, ageing, or rolling. Magnesia reduces liquation and shrinkage and prevents the development of transcrystallization. With 0·1% magnesia the mechanical properties are unaffected, but with 1% they all decrease.—N. A.

***Age-Hardening of Magnesium–Aluminium Alloys.** A. M. Talbot and John T. Norton (*Metals Technology*, 1936, **3**, (2), 13 pp.; *A.I.M.M.E. Tech. Publ.* **692**).—The age-hardening of a magnesium–aluminium alloy containing 9·6% aluminium was studied by measurements of hardness, dilation, and electrical resistance; lattice parameter measurements were also made by a back-reflection camera on a flat surface of a block, the same surface being used subsequently for etching and microscopic examination. All the observations are accounted for by the simple precipitation theory, and no evidence for pre-precipitation-hardening was obtained. The fact that no appreciable change in lattice parameter occurs until after considerable increase in hardness is due to the precipitation taking place very unevenly, so that the main bulk of the alloy may be unchanged whilst small areas show complete precipitation. Under such conditions X-ray diffraction pictures may show only lines due to the unchanged solid solution, but by careful control two sets of lines may be obtained, one corresponding to the precipitated and the other to the unchanged areas.
—W. H.-R.

***Stress-Relief Annealing High-Strength Monel Metal Plate.** (Kosting.) See p. 274.

The Change of Magnetization of Nickel, Permalloy, and Nickel Single Crystals in Strong Fields. (Gerloff.) See p. 239.

***Magnetic Powder Experiments on Rolled Nickel–Iron.** J. L. Snoek (*Physica*, 1936, **3**, (2), 118–124; *Sci. Abs.*, 1936, [A], **39**, 429).—[In English.] The magnetic powder patterns obtained by Bitter (*Met. Abs.* (*J. Inst. Metals*), 1933, **53**, 8) on the surface of molten unstrained ferromagnetic bodies are briefly described and their origin discussed. It is suggested that two phenomena are present—one due to inhomogeneities in the material and the other to the magnetic state of the material and the magnetization process. The latter is investigated by studying the powder figures and the magnetization curves of a 50% nickel–iron alloy in the annealed state ("isotropic" specimen) and in the unannealed condition ("anisotropic" specimen). Marked differences are observed between the two. The general properties of the Bitter patterns are described; the most important of these is that the lines are parallel to the direction of easiest magnetization.—S. G.

***Apparatus for the Determination of Length Variations of Invar Wires 24 m. Long.** P. Werkmeister (*Z. Instrumentenkunde*, 1935, **55**, (12), 499–501; *Sci. Abs.*, 1936, [A], **39**, 102).—Describes a method for determining the variations in length of 25 m. Invar wires. Three wires are individually compared with 24 m. bases set up on a suitable wall of a 3-storey building, one base being arranged on each floor. The average accuracy of comparison is between ± 0·02 and ± 0·03 mm., and it is established that in the case of the 3 wires used no changes greater than 0·1 mm. take place within 5 years. This is confirmed by standardization of the wires at the Physikalisch-Technische Reichsanstalt.—S. G.

***Photoelectric Sensitivity of Palladium–Silver Alloys Saturated with Hydrogen.** F. Krüger and W. Kallenbach (*Z. Physik*, 1936, **99**, (11/12), 743–750).—Palladium–silver alloys, saturated with hydrogen, show a principal maximum photoelectric sensitivity in the case of alloys containing 20% of silver and a lower maximum for alloys containing 60–65% of silver. These compositions correspond very closely with the composition of the alloys absorbing a principal and secondary maximum proportion of hydrogen. These results are discussed in connection with the results obtained by Schniedermann.—J. S. G. T.

***Binary Systems of Alkali Metals.—I [Rubidium–Potassium ; Rubidium–Sodium].** Carlo Goria (*Gazz. chim. ital.*, 1935, **65**, 865–870 ; *C. Abs.*, 1936, **30**, 3305).—The systems rubidium–potassium and rubidium–sodium were studied. The thermal diagram of the system rubidium–potassium indicates a limited range of solid solutions, the absence of any compounds, but the formation of a eutectic at 34° C. with 15% of potassium. The rubidium–sodium diagram indicates the absence of solid solutions and of compounds, but shows a eutectic at − 5·2° C. with 92·0% rubidium. The results confirm those of Rinck (*Met. Abs.* (*J. Inst. Metals*), 1933, **53**, 495 ; *Met. Abs.*, 1934, **1**, 75 ; 1935, **2**, 283). The technique of preparation of the alloys is described.—S. G.

†**The Mechanical Properties of Tin-Base Alloys.** D. J. Macnaughtan and B. P. Haigh (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 235–241 ; and *Rev. Mét.*, 1936, **33**, (1), 7–13).—[In French.] See *Met. Abs.*, this vol., p. 38.—S. G.

***The Effect of the Addition of Lead on the Endurance Limit of a Certain Tin-Base Bearing Alloy.** John N. Kenyon (*Amer. Soc. Test. Mat. Preprint*, **1936**, 4 pp. ; and *Met. Ind.* (*Lond.*), 1936, **49**, (2), 33-34).—Continuation of an investigation commenced several years ago to determine the effect of the addition of lead to certain tin–antimony–copper Babbitt metals (*Met. Abs.* (*J. Inst. Metals*), 1933, **53**, 495). The addition of 4% of lead has little effect on the endurance properties of a tin-base bearing alloy (tin 80, antimony 15, copper 5%) ; if lead has any effect at all it is to improve the material. A peculiar form of bearing failure was found to occur after the alloy has been in service for a long time and an explanation of this failure is offered.—S. G.

***Spectral Emissivities, Resistivity, and Thermal Expansion of Tungsten–Molybdenum Alloys.** P. N. Bossart (*Physics*, 1936, **7**, (2), 50–54).—The spectral emissivities of 3 different tungsten–molybdenum alloys were determined ; the emissivities of tungsten and molybdenum were also measured. The resistivity–temperature and thermal expansion–temperature relations are given for the alloy containing 46·5 atoms-% tungsten. The resistivity relation is compared with those of its constituents and discussed with reference to the theory of Bragg and Williams.—S. G.

***Investigations on the Wear of Metals in Sliding Friction.** Bruno Kehl and Erich Siebel (*Arch. Eisenhüttenwesen*, 1935–1936, **9**, (11), 563–570).—The wear, friction, and tendency to seize of cast bronze, leaded bronze, and white metal bearings in contact with steel shafts was determined, using pure oil and oil–emery mixtures as lubricants. Under oil the tendency to seize increases with increasing surface roughness and in pure oil the wear decreases with increasing time of test owing to the gradual increase in smoothness which occurs. With constant load the speed of rotation has no influence on the wear in oil–emery mixtures, but with increasing load the wear increases at first proportional to the load and then as seizure occurs to several hundred times the normal value. Leaded bronze had the smallest coeff. of friction of the alloys tested.—A. R. P.

***Irreversible Transformations in Solid Solutions.** J. B. Fridman (*Zhurnal Tehnicheskoy Fiziki* (*J. Tech. Physics*), 1936, **6**, (1), 78–82).—[In Russian.] In binary nickel– and manganese–iron alloys the hysteresis of the α–γ-transformation is a linear function of the concentration, the effects of the nickel and manganese being proportional to their atomic radii. In ternary nickel–chromium–iron alloys, and in the separation and redissolution of carbides in nickel–chromium steel, hysteresis increases approximately directly with the concentration. The combined effect of temperature and deformation in nickel–iron and nickel–chromium–iron alloys is to reduce the hysteresis by 17–25%.—N. A.

Equation of State and Heat of Phase Transformations in Solids. Y. A. Trotzkiy (*Zhurnal Tehnicheskoy Fiziki* (*J. Tech. Physics*), 1936, **6**, (1), 125–136).—[In Russian.] From the equation of state of a solid, and taking into account its crystalline structure, its heat of transformation (Q) is given by the expression: $Q = 9RT^2(n_2\alpha_2 - n_1\alpha_1$, where T is the temperature and α_1 and α_2 are the coeff. of expansion and n_1 and n_2 constants characterizing the crystal lattice before and after transformation.—N. A.

***Representation of Ternary Systems.** Erich Scheil (*Arch. Eisenhüttenwesen*, 1935–1936, **9**, (11), 571–573).—To facilitate the reading of ternary diagrams it is suggested that they should be accompanied by plans showing the 3- and 4-phase equilibria in the system and their relations to one another. Examples of the following systems are given *inter alia*: iron–cobalt–tungsten, iron–Fe_3C–copper, iron–chromium–carbon, iron–manganese–carbon, and iron–zirconium–carbon.—A. R. P.

III.—STRUCTURE

(Metallography; Macrography; Crystal Structure.)

(Continued from pp. 202–203.)

***On Grain-Size in Metals.** M. G. Oknov (*Metallurg* (*Metallurgist*), **1936**, (3), 3–8).—[In Russian.] From experiments on the effects of up to 1% of numerous metals on the crystallization of pure zinc, the reduction in grain-size caused by certain impurities is attributed to the formation of a disperse oxide phase (*e.g.* magnesia, alumina, or lime) which promotes the formation of numerous crystallization nuclei but this theory alone is insufficient to explain all cases of grain-size reduction; thus when mercury is added, the wide temperature interval within which separation of excess zinc takes place may be considered as one of the factors.—N. A.

***Grain Changes in α-Copper Solid Solutions in the Cast Condition Produced by Annealing.** M. P. Slavinskiy, O. S. Malhasjan, and L. R. Edelson (*Metallurg* (*Metallurgist*), **1936**, (3), 8–15).—[In Russian.] Tests on brass with 30, 37, and 40% zinc, and on phosphor-bronze with (*a*) tin 4·5, phosphorus 0·02% and (*b*) tin 6, phosphorus 0·2% showed that α-solid solutions of copper can be recrystallized in the cast state provided that the annealing is prolonged to overcome the effects of segregation. Alloys near the limit of the solid solution range recrystallize more rapidly due to volume changes in the segregated regions between the grains. Brass shows a greater grain-growth than bronze.—N. A.

***Dependence of Diffusion Velocity in Metals on Crystallographic Orientation (Anisotropic Diffusion).** W. Bugakow and N. Breschnewa (*Tech. Physics U.S.S.R.*, 1935, **2**, (5), 435–443; *Sci. Abs.*, 1936, [A], **39**, 225).—[In German.] An investigation was made of the velocities of diffusion of mercury into cadmium and into zinc single crystals in different crystallographic directions. In both cases the diffusion velocity is greatest in the direction parallel to the base plane and least perpendicular to this plane. Increase of temperature decreases the importance of the crystallographic direction for the velocity of the process, *i.e.* it decreases the degree of anisotropy of the diffusion. The diffusion velocity of copper into zinc unit crystals was also investigated and analogous results obtained.—S. G.

***Diffusion Coefficients of Metals and Grain Magnitudes.** W. Bugakow and F. Rybalko (*Tech. Physics U.S.S.R.*, 1935, **2**, (6), 617–623; *Sci. Abs.*, 1936, [A], **39**, 225).—[In German.] This investigation deals with the alteration in the diffusion coeffs. of brass on transition from single- to poly-crystals and for different grain magnitudes in the poly-crystal. The absolute value of the diffusion coeff. increases appreciably during the transformation from single- to poly-crystals. The heat of relaxation, which has been evaluated from the temperature curve of the diffusion coeff., decreases on transition from the single- to the poly-crystal, but remains constant during further diminution of the grain. An explanation of the results is based on the general theoretical presentation of diffusion phenomena. The results are compared with data obtained by electrolytic conductivity measurements.—S. G.

***Microscopic Study of Ancient Bronze and Copper [Corrosion of Ancient Bronze and Copper].** C. G. Fink and E. P. Polushkin (*Metals Technology*, 1936, **3**, (2), 28 pp.; *A.I.M.M.E. Tech. Publ.* No. **693**.—Microstructures of ancient objects of copper and bronze are described and illustrated. In general, where the metal is not corroded the structures are quite normal and can readily be explained in terms of known metallurgical operations. The microstructure of the corroded metal is described, and attempts made to interpret the corrosion process and the nature of patina. The ultimate product is always the basic carbonate, malachite, which is stable in the presence of carbon dioxide. This however, is never found in contact with the original metal; the corrosion process is very complex, and six types have been recognized, for details of which the original paper must be consulted.—W. H.-R.

†**The Structure of Some Copper Alloys.** W. Broniewski (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 219–227).—[In French.] See *Met. Abs.*, this vol., p. 76.—S. G.

***The Structure of the Electrolytical Oxide Layer on Aluminium.** E. J. W. Verwey (*Z. Krist.*, 1935, **91**, (3/4), 317–320).—[In English.] In the electrolytic oxidation of aluminium in oxalic acid solution, the oxide consists of a cubic form (Burgers, Claasen, and Zernike, *Met. Abs.* (*J. Inst. Metals*), 1932, **50**, 289) which V. terms γ'-Al_2O_3. The structure of this may be described as a simple face-centred cubic lattice of oxygen ions in which the Al^{3+} ions are distributed over all the available interstices in such a way that about 70% occupy an octahedron hole (co-ordination number 6), and 30% a tetrahedron hole (co-ordination number 4). The γ'-Al_2O_3 may be regarded as the first stage in the transition from the amorphous to the truly crystalline form.
—W. H.-R.

Electrolytic Conduction of a Solid Insulator at High Fields. The Formation of the Anodic Oxide Film on Aluminium. E. J. W. Verwey (*Physica*, 1935, **2**, (10), 1059–1063).—[In English.] The mechanism of electrolytic oxide coating of aluminium is described in terms of a current of Al^{3+}-ions penetrating into a preformed lattice of oxygen ions. This picture explains the peculiar structure of γ'-Al_2O_3 (average structure of the cations) and is in rather good agreement with considerations on the basis of lattice theory of the electrolytic conductivity of crystals at high fields (cf. Verwey, preceding abstract, and *J. Chem. Physics*, 1935, **3**, 592).—S. G.

***Investigation of Aluminium Oxides by Means of Electron Interference.** E. Belwe (*Z. Physik*, 1936, **100**, (3/4), 192–196).—Electrolytic aluminium oxide films are shown to be composed of very finely crystalline γ-Al_2O_3. The oxide produced on aluminium foil by heating in a flame is found to be composed of the same γ-oxide in the usual macrocrystalline form interspersed with thin aluminium single crystals.—J. S. G. T.

***X-Ray Diffraction Study of Liquid Sodium.** L. P. Tarasov and B. E. Warren (*J. Chem. Physics*, 1936, **4**, (4), 236–238).—S. G.

***Optical Investigation of Thin Gold Foils.** R. Smoluchowski (*Phys. Rev.*, 1936, [ii], **49**, (8), 649).—Abstract of a paper read before the American Physical Society. Measuring the light absorption in thin metal films, Smakula found consecutive maxima and minima of absorption coeff. which he indicated as connected with electronic levels (bands) in the metal lattice. Since this point of view does not seem to be satisfactory, new measurements were made on thin gold foils in order to find another possible explanation of the effect. Special care was taken to obtain reproducible foils, which were made by condensing gold vapour on thin glass plates. A special arrangement permitted the estimation of the thickness obtained. The measurements were carried out on a series of foils, varying from "colloidal" foils to good developed crystal layers, the different structures of the foils appearing in the shape of the absorption curve. The dependence of the observed maxima and minima of absorption of light in the wave-length region 4800–6800 A. on the thickness of the foils and the angle of transmission of the light beam seems to indicate that the origin of this effect is to be looked for in the grain structure of the surface of the foils, the grains themselves, however, being very different from uniform spheres as it was supposed in Mie's theory of optical effects in colloids.—S. G.

***The Diffraction of Electrons by Amalgam Films.** A. E. Aylmer, G. I. Finch, and S. Fordham (*Trans. Faraday Soc.*, 1936, **32**, (6), 864–871).—Examination of the effect of mercury vapour on gold and silver leaf by the electron diffraction method revealed a large number of amalgam structures; both metals form simple cubic amalgams with a = 17·81 and 17·94 A., respectively. Two other face-centred tetragonal silver amalgams were also identified. Copper behaves similarly to gold and silver, but the structure of none of the resulting amalgams has been identified. Palladium is only slightly attacked by mercury. The bearing of these results on electron diffraction examination in general is discussed at some length.—A. R. P.

***An Electron Diffraction Study of the Structure of Electrodeposited Metals.** G. I. Finch and C. H. Sun (*Trans. Faraday Soc.*, 1936, **32**, (6), 852–863).—Examination of the structure of thin films of various metals deposited on discs of other metals from standard types of plating baths showed that the structure of the substrate nearly always affects that of the deposit so that the orientation of the crystals in the former determines that of the crystals in the latter. In cases where this does not occur the deposits generally are very poorly adherent. When zinc is deposited on polished copper, the first layers appear to enter into solid solution, but this is not the case with a crystalline copper substrate. When the substrate consists of fairly large (50–100 A.) crystals randomly oriented the first thin films of some metals deposited on them are unoriented, but these display their normal characteristics when they become sufficiently thick; generally the effect of the substrate extends to layers up to 1000 A. thick, and is most marked on nickel deposits on gold. Iron as a substrate has no influence on the structure of nickel or copper deposits. Metal films formed by chemical displacement, *e.g.* platinum by silver or copper, are shown by electron diffraction methods to be alloys.—A. R. P.

***The Röntgenographic Study of Zinc and Cadmium Films Deposited in the Presence of Colloids.** L. Palatnik (*Trans. Faraday Soc.*, 1936, **32**, (6), 939–941).—The degree of orientation of the crystals of zinc and cadmium deposits obtained from baths containing colloids depends on the concentration of the latter and on the current density. The effective concentration of the colloid in a plating bath can be determined by comparing the degree of orientation of the deposits obtained from the bath with that of deposits obtained from a bath containing a known amount of freshly added colloid. The brightening effect of colloids in zinc and cadmium plating baths is ascribed to their promotion of the growth of crystals in a single crystallographic direction perpendicular to the surface of the cathode.—A. R. P.

Contribution to the X-Ray Examination of Light Metals. Fritz Regler (*Mitt. tech. Versuchsamtes*, 1935, **25**, 20–28).—Laue diagrams of the fracture of light metal alloys give information not only on the grain-size but also on the size and shape of the macroscopic particles of the precipitated phase and on the distribution of porosity. The full beneficial effect of the precipitated phase is exerted only when it is in such a disperse form that it cannot be detected by the Laue method; 10 photographs of characteristic structures are given and discussed.—A. R. P.

***Al $K\beta$-Bands from Al–Cu Alloys.** Sanae Yoshida (*Sci. Papers. Inst. Phys. Chem. Res. Tokyo*, 1936, **28**, (617), 245–248).—[In English.] The form of the

Al $K\beta$ emission bands was studied, using targets of pure aluminium and of aluminium–copper alloys containing from 5 to 90 wt.-% copper. With pure aluminium the width of the band agrees with that calculated for three free electrons per atom. Application of the theory of Jones (*Met. Abs.*, 1934, **1**, 237) suggests that the free electrons in aluminium must overlap from the first Brillouin zone into the second, but in the α-solid solution in copper, and also in the $\gamma'(Cu_9Al_4)$ phase the overlapping into the second zone is small.
—W. H.-R.

***Lattice Constants and Refractive Indices of Crystals of Bismuth and Antimony.** E. Rosa (*Nuovo cimento*, 1935, **12**, (7), 448–454; *Sci. Abs.*, 1936, [A], **39**, 9).—The spacings of the (111) planes of bismuth and antimony crystals were measured, using Adinolfi's method. This reduces the errors and simplifies the Bragg method. The mean values (for Cu and Fe K_α and K_β radiations) are 3·957 and 3·754 A. for bismuth and antimony, respectively, as compared with 3·92 and 3·77 found by Ogg using Pd and Rh K-radiations.—S. G.

***The Crystal Structure of CdSb.** Arne Ölander (*Z. Krist.*, 1935, **91**, (3/4), 243–247).—[In German.] The crystal structure of the compound CdSb is orthorhombic, with $a = 8·492$ A., $b = 8·320$ A., $c = 6·390$ A.; the unit cell contains 16 atoms. The symmetry belongs to one of the following space groups: $V_h^{1,5,13}$, $V^{1,2,3,4}$, or $C_{2v}^{1,2,4,7}$. Both powder and rotation photographs were taken, but the positions of the individual atoms cannot yet be determined.
—W. H.-R.

***Precision Measurement of the Lattice Constant of Very Pure Niobium.** M. C. Neuburger (*Z. Krist.*, 1936, **93**, (1/2), 158–159).—[In German.] Very pure niobium was found to have a body-centred cubic structure with $a = 3·2941 \pm 0·0003$ A. at 20° C. The closest interatomic distance is 2·852(7) A., and the ideal density is 8·56(9).—W. H.-R.

***The Crystal Structure of Gallium.** A. J. Bradley (*Z. Krist.*, 1935, **91**, (3/4), 302–316).—[In English.] The crystal structure of gallium was re-determined by the powder method, with accurate photometer measurements of the intensities of the lines. The results confirm Laves' orthorhombic structure (*Met. Abs.* (*J. Inst. Metals*), 1933, **53**, 440) except that all three axes are different in length, whereas Laves gave equal values for the a and b axes. The lattice constants are $a = 4·5167 \pm 0·0001$ A., $b = 4·5107 \pm 0·0001$ A., $c = 7·6448 \pm 0·0002$ A. The space group is V_h^{18}, and the atomic co-ordinates are $(m\ 0\ p)$, $(m + \frac{1}{2}, \frac{1}{2}, \bar{p})$, $(\bar{m} + \frac{1}{2}, \frac{1}{2}, p)$, $(\bar{m}\ 0\ \bar{p})$, $(m, \frac{1}{2}, p + \frac{1}{2})$, $m + \frac{1}{2}, 0, \bar{p} + \frac{1}{2})$, $(\bar{m} + \frac{1}{2}, 0, p + \frac{1}{2}$, $(\bar{m}, \frac{1}{2}, \bar{p} + \frac{1}{2})$, where $m = 0·0785 \pm 0·0005$ and $p = 0·1525 \pm 0·0005$ A.—W. H.-R.

Correction to the Paper "X-Ray Analysis of the Structure of the Homogeneous Phases in the Magnesium–Nickel System." E. Bachmetew (*Acta Physicochimica U.R.S.S.*, 1935, **3**, (4), 531).—[In German.] See *Met. Abs.*, 1935, **2**, 513. The number of $MgNi_2$ molecules per elementary cell is only 8 in place of 16 as previously reported.—S. G.

***X-Ray Studies on Heat-Treatment of Nickel–Iron–Aluminium Magnetic Alloys.** L. Werestschiagin and G. Kurdjumow (*Tech. Physics U.S.S.R.*, 1935, **2**, (5), 431–434; *Sci. Abs.*, 1936, [A], **39**, 326).—[In German.] The crystal structure of iron–nickel–aluminium alloys and their alteration with thermal treatment were studied by means of X-rays. A cubic body-centred lattice with a regular atomic arrangement was observed. The alteration in coercivity does not depend on the separation of dispersed particles from the supersaturated solid solutions but is governed by the reactions which occur within the solid solution before the separation of the γ-phase.—S. G.

***Lattice Distortion and Coercive Force in Single Crystals of Nickel–Iron–Aluminium.** W. G. Burgers and J. L. Snoek (*Physica*, 1935, **2**, (10), 1064–1074; *C. Abs.*, 1936, **30**, 1344).—[In English.] A report on preliminary results obtained in a correlated investigation of the ageing characteristics of alloys of the system nickel–iron–aluminium, wherein a simultaneous study of the ferromagnetic and X-ray diffraction properties is being made. Single crystals 1 mm. square and 10–20 mm. long were made of the alloy nickel 28·3, aluminium 12·6, iron 59·1%. Three methods of heat-treatment were used: (*a*) the sample is quenched and reheated to the desired temperature; (*b*) the sample is quenched to the desired temperature and held there; and (*c*) the sample is quenched at varying cooling rates. In agreement with Glocker (*Arch. Eisenhüttenwesen*, 1935, **8**, 561), it was found that lines of the γ-phase are only visible after cooling rates which are considerably slower than those required to produce the maximum coercive force (590 gauss after cooling from 1200° to 700° C. in 45 seconds) and superlattice lines of the FeAl type were always obtained in almost constant relative intensity. Line-broadening results do not agree with those of Glocker. Marked line-broadening ((310) CoK_{α} radiation) occurs wherever the coercive force differs much from zero. When the coercive force has reached its maximum value and is decreasing the usual type of line broadening occurs. Previous to this value, the photometer curves show an increasing blackening of background with no loss in definition of the doublet. The latter type of line broadening is thought to be due to the formation of regions of slightly larger and slightly smaller spacing in the matrix lattice which has not lost its coherence. A number of possible explanations are given, tentatively, for the behaviour of coercive force and line broadening. —S. G.

***On Osmiridium.—III.** O. E. Zviaginzew and B. K. Brunowsky (*Z. Krist.*, 1936, **93**, (3/4), 229–237).—[In German.] Cf. *Met. Abs.* (*J. Inst. Metals*), 1933, **53**, 352. Single crystals of osmiridium were examined by the rotation and Laue methods. They show a fibre structure (*Faserstruktur*) in which the orientation of different parts may vary by amounts up to ± 10°. The reason for this is to be found in the geological processes during and after the solidification of the naturally occurring metal.—W. H.-R.

***Precision Measurement of the Lattice Constant of Very Pure Tantalum.** M. C. Neuburger (*Z. Krist.*, 1936, **93**, (3/4), 312–313).—[In German.] Very pure tantalum, free from hydrogen, was found to have a body-centred cubic structure with $a = 3{\cdot}2959 \pm 0{\cdot}0003$ A. at 20° C. The smallest interatomic distance is $2{\cdot}854_4$ A., and the ideal density $\rho_x = 16{\cdot}655$, assuming the atomic weight to be 180·89, or 16·702 if the atomic weight is 181·4.—W. H.-R.

***Precision Measurement of the Lattice Constant of Pure Vanadium.** M. C. Neuburger (*Z. Krist.*, 1936, **93**, (3/4), 314–315).—[In German.] Very pure vanadium (free from oxygen, hydrogen, and nitrogen) was found to have a body-centred cubic structure with $a = 3{\cdot}0338 \pm 0{\cdot}0003$ A. at 25° C. The closest interatomic distance is 2·6274 A.—W. H.-R.

***Lattice Constants of Five Elements Possessing Hexagonal Structure [Zinc, Magnesium, Ruthenium, Osmium, Beryllium].** E. A. Owen, Llewelyn Pickup, and I. O. Roberts (*Z. Krist.*, 1935, **91**, (1), 70–76).—[In English.] The following lattice constants were determined by the powder method in a precision camera at 18° C.: zinc (99·9%), $a = 2{\cdot}6591 \pm 0{\cdot}0005$ A., $c = 1{\cdot}8560$; magnesium (99·9%), $a = 3{\cdot}2020 \pm 0{\cdot}0005$ A., $c = 1{\cdot}6240$; ruthenium (99·9%), $a = 2{\cdot}6987 \pm 0{\cdot}0005$ A., $c = 1{\cdot}5833$; osmium (99·8%), $a = 2{\cdot}7304 \pm 0{\cdot}0005$ A., $c = 1{\cdot}5785$; beryllium (99·8%), $a = 2{\cdot}2812 \pm 0{\cdot}0005$ A., $c = 1{\cdot}5685$. The values are in fair agreement with those found by other observers except for beryllium, for which Neuburger (*Met. Abs.* (*J. Inst. Metals*), 1932, **50**, 615) gave $a = 2{\cdot}2680$, and axial ratio 1·5847 at 20° C. [*Note by Abstractor:* It must be emphasized that, in contrast to most papers, c (as given above) is not the second lattice constant, but is the axial ratio, which is usually denoted c/a. The confusion is aggravated by the fact that O., P., and R. give their values of c in A., whereas since c is the axial ratio the values are dimensionless numbers.]—W. H.-R.

***On the Theory of Superlattice Structures in Alloys.** William Hume-Rothery and Herbert M. Powell (*Z. Krist.*, 1935, **91**, (1), 23–47).—[In English.] Theoretical. Data on superlattice structures are reviewed with special reference to those in iron–aluminium alloys. In any given crystal structure each atom has definite numbers of neighbours at fixed distances. These are called zones of neighbouring atoms, and are numbered outwards so that the closest neighbours occupy the first zone, the next closest the second zone, and so on. The following hypotheses are suggested : (1) the formation of a normal superlattice takes place so that the solute atoms occupy the highest zone possible relative to one another ; (2) when compared with a purely random arrangement, the relative stability of a superlattice decreases rapidly as its zone number increases ; (3) other things being equal the tendency to form a superlattice becomes greater with increasing difference between the atomic diameters of solvent and solute atoms. The underlying idea is that the ideal structure stable at low temperatures is that in which the strain in the lattice is relieved by the solute atoms keeping as far away from one another as possible. This theory is applied in detail to the compositions over which the different superlattice structures are stable in iron–aluminium alloys.—W. H.-R.

Solid Solutions with a Varying Number of Atoms in the Unit Cell. Gunnar Hägg (*Z. Krist.*, 1935, **91**, (2), 114–122).—[In English.] Refers almost exclusively to salts, but some points are of great general interest. H. discusses the formation of solid solutions by addition or subtraction of one kind of atom to or from a lattice. The interstitial solid solution of carbon in γ-iron is an example of a solid solution formed by addition. The solid solution in FeS containing more sulphur than required by the formula FeS is not, however, produced by the addition of sulphur atoms to the FeS lattice, but by the falling out or subtraction of iron atoms, so that there are vacant spaces in the lattice. —W. H.-R.

Condition of Electrons in Crystal Lattices. F. Hund (*Z. tech. Physik*, 1935, **16**, (11), 494–497).—Cf. *Met. Abs.*, this vol., p. 119. An approximate calculation is given of the position of the energy bands in some simple crystal lattices. —J. S. G. T.

Lattice Constants for the Year 1936 [1935]. M. C. Neuburger (*Z. Krist.*, 1936, **93**, (1/2), 1–36).—[In German.] The title of this paper is misleading, since the lattice constants included are those determined up to the end of 1935. Tables are given showing the crystal structures of the elements with lattice constants, interatomic distances, atomic volumes, and related constants together with the purity of the samples used. The interatomic distances given for α-manganese appear to be incorrect, and reference should be made to the original paper of Bradley and Thewlis (*Proc. Roy. Soc.*, 1927, [A], **115**, 456). The reader is advised to refer to the previous tables of this type for definitions of "Volumen pro Atom," "Atomraum," and "Atomvolumen," since these are easily confused (Neuburger, *Z. Krist.*, 1931, **80**, 103 ; 1933, **86**, 395).—W. H.-R.

The Application of Plane Groups to the Interpretation of Weissenberg Photographs. M. J. Buerger (*Z. Krist.*, 1935, **91**, (3/4), 255–289).—[In English.] —W. H.-R.

†Study of Metallic Crystals. E. Schmid (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 207–217).—[In French.] See *Met. Abs.*, this vol., p. 77.—S. G.

The Structure of Plastically Deformed Crystals According to Laue Patterns. —I. A. P. Komar (*Physikal. Z. Sowjetunion*, 1936, **9**, (5), 413–432).—[In English.] K. reviews briefly and critically the various conceptions of the structure of plastically deformed crystals as based on experimental data obtained by the Laue method. A graphical analytical method for calculating Laue patterns of plastically deformed crystals is proposed. Experimental are compared with calculated results.—S. G.

Ideal and Real Crystals. Discussion. —— (*Z. Krist.*, 1936, **93**, (3/4), 161–228).—[In English or German.] The papers read at the conference on ideal and real crystals (*ibid.*, 1934, **89**, 193) have already been abstracted (*Met. Abs.*, 1935, **2**, 225–226, 289–291, 344). In the present discussion, *A. Goetz*, *W. Kleber*, *H. E. Buckley*, *D. Balarew*, *E. Orowan*, *M. J. Buerger*, *A. H. Jay*, *A. Smekal*, *L. Royer*, *G. Kalb*, *M. Straumanis*, and *P. Niggli* criticize each other's papers, and reply to the criticisms. The contributions are essentially argument and counter-argument, and are unsuitable for abstracting, but reference may be made to discussion on the structure and properties of bismuth (pp. 167–173), and a modification of the dislocation theory of G. I. Taylor (pp. 188–191). —W. H.-R.

Guide for Crystal Drawing. Wallace M. McNabb and Josephine McNabb (*J. Franklin Inst.*, 1936, **221**, (4), 539–546).—A guide, for student use, is presented for the drawing of a crystal using the gnomonic projection.—S. G.

IV.—CORROSION

(Continued from pp. 203–206.)

***Additions of Inhibitors to Solutions which Corrode Aluminium.** Hans Röhrig (*Congrès internat. Mines Mét. Géol.*, *Section de Métallurgie*, 1935, **1**, 201–204).—[In French.] See *Met. Abs.*, this vol., pp. 8 and 78.—S. G.

***Bimetallic Corrosion Promoted by Copper–Aluminium Contacts.** P. Mabb (*Metallurgia*, 1936, **14**, (80), 29–31).—A number of locking-nuts, of Duralumin, some anodized and some untreated, were submitted to the salt-spray test for 16 days, half of the specimens being in close contact with brass studs. The untreated specimens gradually corroded during the first few days of the test, while the anodized specimens were entirely unaffected at the end of the 16 days' period. In the cases of the nuts in contact with brass studs, the untreated nuts were exceptionally heavily corroded around areas of contact within 3 days; the anodized nuts blackened at the junction in 1 day, and were heavily corroded there in 5 days.—J. W. D.

Sheath Corrosion at Bond Straps Prevented. Edgar F. Wolf (*Elect. World*, 1936, **106**, (9), 597–598, 645).—Sheath failures from electrolytic corrosion have been observed when copper bonding strips have been run parallel to a lead-covered cable before soldering, and have been in contact with a cement fire-proofing mixture which was applied over the sheath. To prevent electrolysis, the bonding strip should be brought out at right angles to the cable, and as a further measure an insulating coating of hot asphalt and muslin should be applied.—J. C. C.

The Dependence of the Corrosion Velocity of Copper on the Thermal and Mechanical Pretreatment of the Copper. —— du Bois (*Oberflächentechnik*, 1933, **10**, 235; *C. Abs.*, 1934, **28**, 731).—Preliminary results show that the corrosion of copper is influenced by the arrangement of the crystallites, but there is a difference between refined and electrolytic copper, and also between hard and soft electrolytic copper. X-ray tests revealed that soft annealed material has an orientated crystallite arrangement; hard material has an unorientated arrangement. Electrolytic copper showed less resistance to sulphuric acid than refined copper. Rolling produces differences in the crystallite arrangement and therefore different corrosion-resistance. It is proposed to use as a definition of the intensity of an attack by acid on a material not the % loss in weight, but the time required to corrode away 1 mm. thickness from the surface of the material by the corroding agent, and also to state previous thermal and mechanical treatment of the material.—S. G.

***Corrosive Action of Sulphur Against Metallic Copper During Vulcanization.** Shozo Satake (*Rubber Chem. Tech.*, 1936, **9**, 301–310).—See *Met. Abs.*, this vol., p. 9.—S. G.

Report of the Copper Pipe Committee. See p. 280.

***Microscopic Study of Ancient Bronze and Copper [Corrosion of Ancient Bronze and Copper].** (Fink and Polushkin.) See p. 251.

***Comparative Tests on the Attack on Hard Lead Tubes by Tap-Water.** Erich Naumann (*Gas- u. Wasserfach*, 1936, **79**, (14), 214–216).—Hard lead pipes with up to 1·5% antimony behave similarly to ordinary soft lead pipes in tap-water during the early stages of corrosion, but in water with a high content of carbon dioxide they cannot be used. In oxygenated waters the quantity of lead dissolved increases with decreasing carbonate hardness and with increasing content of free carbonic acid. Antimony cannot be detected in the water. —B. Bl.

External Factors Leading to Corrosion of the Positive Grids of the Foré-Type Lead Battery Plates Under Conditions of Formation. A. B. Shakhkeldian (*Trans. Saratov Univ.*, 1934, **11**, (2), 59–76; *C. Abs.*, 1936, **30**, 3728).—Corrosion is reduced, and better contact of the paste with the lead grid is ensured, when the surface of the grid is covered with a layer of lead sulphate; this may be achieved by treating the grids separately, or by applying sulphuric acid containing neutral sulphates together with the paste. Other factors affecting corrosion of the grids (composition of the paste, temperature, &c., are discussed).—S. G.

***The Border Effect in the Solution of Metallic Magnesium in Aqueous Solutions of Organic Acids.** Ya. V. Durdin and A. M. Markevich (*Zhurnal Obstchei Khimii* (*J. General Chem.*), 1935, **5**, 1690–1695; *C. Abs.*, 1936, **30**, 3302).—[In Russian.] When a piece of magnesium in contact with an insoluble solid such as paraffin is placed in aqueous HCO_2H, AcOH, $EtCO_2H$, or $PrCO_2H$, the rate of solution of the magnesium in the acid is much greater at the 3-phase boundary magnesium–paraffin–acid than at other points on the magnesium piece. This " border effect " was not observed with hydrochloric acid or ammonium chloride solutions, or with the organic acids if the metal was covered with a film of *iso*-BuOH. The localized increase in rate of reaction depended directly on an increase in the rate of diffusion of the acid solution with respect to the metal surface. The active acid was adsorbed on the surface of the paraffin. The effect was intensified by the stirring action of the hydrogen evolved. When the whole body of aqueous acid was stirred mechanically the effect was obscured. The absence of the border effect with hydrochloric acid and ammonium chloride and in the presence of *iso*-BuOH is attributed to the absence of any differential local diffusion rate. The reservation is made that the observed phenomenon may not be due entirely to a difference in the diffusion rate.—S. G.

***On the Influence of Impurities in Mixtures of Ammonia and Air on the Corrosion of Platinum Gauze.** I. E. Adadurov and P. D. Didenko (*Zhurnal Prikladnoi Khimii* (*J. Applied Chem.*), 1935, [B], 8, (5), 823–832).—[In Russian, with German summary.] The effect of additions of hydrogen, water vapour, oxygen, carbon monoxide, carbon dioxide, and nitrogen oxides to ammonia–air reaction mixtures on the surfaces of platinum-gauze catalysts was investigated, the corrosion of the surfaces being examined photomicrographically. It is concluded that the fundamental cause of the destruction of platinum gauzes is the change in structure of the surface. Hydrogen penetrates into the crystalline space-lattice of the metals, forming compounds which are easily sublimed. Addition, therefore, of hydrogen to the reaction mixture in order to increase the rate, as recommended by some investigators, causes increased corrosion of the gauze. Water vapour causes rapid disintegration of the gauze, and oxygen does likewise. Carbon monoxide increases the reaction temperature, and with it corrosion, while carbon dioxide and nitrogen oxides decrease it somewhat.—M. Z.

***Influence of Internal Diffusion on the Chemical Stability of Binary Solid Solutions.** V. V. Skortcheletti (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1936**, (17), 185–194).—[In Russian.] Corrosion tests on alloys of lead with up to 30 atomic-% mercury in 20% acetic acid at 20°, 13°, and 0° C. showed Tammann reaction limits at 0° C. due to restricted atomic diffusion in the solid solution at this temperature. At 20° and 13° C. no such limits occur owing to the active internal diffusion which removes the protective surface layer of noble metal.—N. A.

***Chemical Resistance of Two-Phase Alloys.** V. V. Skortcheletti (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1936**, (17), 195–208).—[In Russian.] The corrosion of chill-cast and of annealed lead–bismuth and zinc–cadmium alloys in 20% acetic acid 0·1*N*-sulphuric acid, and 5% sodium hydroxide solution follows one of three possible courses: (1) the potential set up between the constituents of the alloy is sufficient to cause prolonged galvanic action, (2) polarization rapidly occurs, (3) a passive film is formed on one of the components. Alloys of eutectic composition do not occupy a prominent position on the composition–corrosion rate curve. Non-homogeneity in normally resistant alloys does not necessarily indicate a low resistance to corrosion.—N. A.

Hot-Water Corrosion and Its Prevention. L. W. Haase (*Vom Wasser; Ein Jahrbuch für Wasserchemie u. Wasserreinigungstechnik*, 1934, **8**, 111–123; *J. Amer. Water Works Assoc.*, 1935, **27**, 1096).—Water in hot-water systems for domestic use is heated to from 50° to 85° C. and corrosion problems differ from those in boiler plant. Corrosion is promoted by electrochemical activity causing solution of metal, and is accelerated by unsuitable material, incorrect construction of plant, and wrong method of operation. The behaviour of iron, galvanized iron, copper, tinned pipes, and pipes with other protective coatings, in hot-water conditions, is exhaustively dealt with. Methods of operation, with illustrations of suggested plant, are given, and methods for deoxygenation of water, mechanically and chemically by sulphite addition, are described. —S. G.

***The Corrosion of Metals by Oil.** P. J. Haringhuizen (*Chem. Weekblad*, 1936, **33**, (22), 351–352).—Read before the Dutch Chemical Society. The corrosion of thin films of metal in oil is investigated by an optical method previously described by H. and D. A. Was (*Proc. K. Akad. Wet. Amsterdam*, 1936, **39**, 201).—L. A. O.

Soil Corrosion Report. H. W. Clark (*J. New England Water Works Assoc.*, 1935, **49**, (1), 110–112; *J. Amer. Water Works Assoc.*, 1935, **27**, 158).—Discusses the data published in *Bur. Stand. Research Paper* No. **329** (see *J. Inst. Metals*, 1928, **40**, 551) and No. **638**.—S. G.

Some Corrosion Phenomena in Mechanically Stressed Metals. C. A. Lobry de Bruyn (*Chem. Weekblad*, 1936, **33**, (22), 340–344).—Read before the Dutch Chemical Society. Corrosion is defined and the electrochemical theory and the effect of tension and deformation on corrosion are discussed. Season-cracking is described, with examples and illustrations, and the penetration of molten metals into strained metals, and the relation of corrosion to the wear of metals are discussed.—L. A. O.

Allotropy and Corrosion. Ernst Cohen (*De Ingenieur*, 1935, **50**, Mk, 13–22; *C. Abs.*, 1936, **30**, 3763).—Recent work on the allotropy of tin (*Met. Abs.*, 1935, **2**, 415) is discussed. Experiments showing the influence of nuclei on the rate of transformation of white to grey tin at − 50° C. were made by rolling, drawing, or bending tin wires at 15° and at − 80° C., respectively. The crystal defects caused by the mechanical working act as crystallization nuclei for the transition and cause a more rapid conversion of the wires deformed at temperatures below the transition temperature of 13·2° C. The more intensive the deformation the more rapidly the conversion takes place. The same crystal defects also affect corrosion-resistance, and the latter can be improved by application of principles of allotropy.—S. G.

Fundamentals of the Electrochemical Theory of Corrosion. A. H. W. Aten and H. W. Deinum (*Chem. Weekblad*, 1936, **33**, (22), 331–337).—Read before the Dutch Chemical Society. A review of the electrochemical theory of corrosion. The factors which should be considered are the molarpotentials of the reactions involved in corrosion, the influence of concentration on the electrode potentials, and the velocity of the reactions. The latter factor can be taken into account by making use of the current–potential curves for the anodic process of the dissolution of metals, and for the cathodic process of oxygen reduction or evolution of hydrogen. This method is illustrated by application to the behaviour of iron and of copper in an aerated solution of potassium chloride.—L. A. O.

Relative Value of Accelerated Corrosion and Outdoor Exposure Tests [**of Plated Coatings**]. William Blum (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (5), 8–13; discussion, 13–16).—A brief discussion of the value of the salt-spray test as an accelerated method of determining the relative merits of plated coatings as a protection from corrosion of steel.—A. R. P.

V.—PROTECTION

(Continued from pp. 206–208.)

***Oxide Films on Aluminium and Its Alloys; Their Production and Properties.** A. G. C. Gwyer and N. D. Pullen (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 243–252).—[In French.] See *Met. Abs.*, this vol., p. 80. —S. G.

Effective Corrosion Protection by Eloxal Films on Cupriferous Aluminium Alloys. Alexander Jenny (*Aluminium*, 1936, **18**, (3), 97–99).—Duralumin treated by the Eloxal process withstands the salt-spray test considerably better than the untreated alloy, and a high degree of immunity from corrosion is afforded by subsequently impregnating the Eloxal film with wax or other sealing material.—A. R. P.

On the Use of Eloxal in Construction. K. Holzwarth (*Werkstatt u. Betrieb*, 1936, **69**, (5/6), 69–71).—A discussion of the advantages and limitations of structural elements treated by the Eloxal process.—P. M. C. R.

†**Some Problems of Technical Electrometallurgy. IV.—Anodic Treatment of Aluminium.** (Masing.) See p. 264.

***The Structure of the Electrolytical Oxide Layer on Aluminium.** (Verwey.) See p. 251.

Electrolytic Conduction of a Solid Insulator at High Fields. The Formation of the Anodic Oxide Film on Aluminium. (Verwey.) See p. 251.

***Investigation of Aluminium Oxides by Means of Electron Interference.** (Belwe.) See p. 251.

***On the Action of Soda Solution Containing Sodium Silicate on Light Metals after Various Treatments.** H. Bohner (*Aluminium*, 1936, **18**, (3), 102–103).—Immersion of rolled sheet aluminium or Pantal in boiling 10% sodium carbonate solution containing 0·1% of sodium silicate results in the formation of a protective film almost immediately, provided that the metal has previously been dipped in hydrochloric acid for 2 minutes, or in hot caustic soda solution, or has been scrubbed with steel wool. After sand-blasting or pickling for longer than 2 minutes in hydrochloric acid, the protective film takes a very long time to form in the sodium carbonate solution.—A. R. P.

Aluminium Armour for Electric Cables. J. Bally (*Rev. Aluminium*, 1936, **13**, (80), 159–163).—Describes the use and advantages of aluminium ribbon for protecting electric cables; the advantages over steel wire are: (1) lightness; (2) better corrosion resistance; (3) increased transport capacity; (4) higher current intensity can be carried in cases of short-circuiting; (5) aluminium armour will act as a fuse at 650° C. in the case of very intense currents; (6) by interposing a thin ribbon of aluminium between the insulator and the armour, a hermetically sealed envelope is obtained in case of fire; (7) owing to its being non-magnetic, loss by Foucault currents is avoided in the case of aluminium; (8) owing to its higher conductivity, it is more effective in cutting down parasitic currents.—J. H. W.

***Copper Wire Tinning.** Ch. Bernhoeft (*Draht Welt*, 1936, **29**, (15), 231–233).—A detailed account is given of work the results of which have already been published (*Met. Abs.*, this vol., p. 10).—A. R. P.

Some Practical Aspects of Galvanizer Dross at High Temperatures. —— (*Iron Steel Canada*, 1936, **19**, (1), 11–12).—In the hot-dip galvanizing process compounds of iron and zinc form dross. The compound formed at about 430° C. is rhombohedral, whilst that produced at 480° C. is hexagonal pyramid. The latter has a higher melting point, and a cycle of over-heating may be set up in an effort to maintain fluidity, since higher pot temperature increases the rate of iron pick-up. The remedy lies in dilution of the bath and lower temperatures. Correct bath temperature and galvanizing conditions are indicated by the rhomohedral form of dross crystal.—J. E. N.

Electro-Galvanizing of Round Wire. Arnold Weisselberg (*Iron Age*, 1936, **137**, (21), 34–37, 106).—The history of wire-galvanizing is briefly reviewed, and the physical properties of wire before and after electro-galvanizing are tabulated. The technique of cleaning the wire and the practical operations of electro-galvanizing, the appearance of the finished wire, and the cost of the process are described. The operation advantages are that there are no zinc losses, no losses during shut down, the plant can be started up and shut down quickly, the thickness of coating can be accurately controlled, and the operation of the plant is clean and not unhealthy. The quality advantages are that the coating is very uniform, coatings of more than 4 dips can be produced, the physical properties of the wire are unaffected, the coating of pure zinc is highly ductile and adherent, the coating surface is smooth, and defects of the wire are revealed.—J. H. W.

***Metallic Cementation. IX.—Cementation of Some Metals [Iron, Nickel, and Copper] by Means of Beryllium Powder.** Tsutomu Kase (*Kinzoku no Kenkyu* (*J. Study Metals*), 1936, **13**, (5), 173–188).—[In Japanese.] The cementation of iron, nickel, and copper by powdered beryllium was investigated between 600° and 1100° C. The surface layers were studied by hardness tests, measurements of depth of penetration, microscopic examination, and chemical analysis. Experiments were also carried out on corrosion in acids and oxidation at high temperatures. Beryllium diffuses into these three metals above 600° C., the rate of diffusion increasing with increase in temperature; the rate of diffusion into iron, however, increases more or less abruptly at the A_3 point (906° C.). The relation between increase in weight of the specimen (ΔW) or the depth of penetration (P) and the absolute temperature of cementation (T) or the time required for the treatment (θ) is given by an exponential function ΔW (or P) $= ae^{-b/T}$, or ΔW (or P) $= ae^{b\theta}$, where a and b are constants variable in each case. The hardness of the cemented surfaces is remarkably greater than that of the original materials, and some of them have a sturdy resistance to atmospheric oxidation and attack by certain fused salts at high temperatures. Iron surfaces cemented with beryllium are not easily attacked by concentrated nitric acid, but, in general, the surfaces of the metals cemented with beryllium have little advantages as regards resistance to corrosion by strong acids.—S. G.

***Corrosion-Protective Value of Electrodeposited Zinc and Cadmium Coatings.** W. Blum, P. W. C. Strausser, and A. Brenner (*Met. Ind.* (*Lond.*), 1936, **48**, (21), 591–594; (22), 615–618; (23), 639–643).—Reproduced from *J. Res. Nat. Bur. Stand.*, 1936, **16**, (2), 185–212. See *Met. Abs.*, this vol., p. 122.—J. E. N.

The Development and Application of Metal Spraying. The Molten Metal Process. A. E. Phipps (*Met. Ind.* (*Lond.*), 1936, **48**, (20), 553–558).—The development of the process since its introduction by Schoop in 1909 is described. The process is limited to the low melting point metals, tin, lead, zinc, cadmium, and aluminium; its applications are discussed, including experimental work on the spraying of bearing metals.—J. E. N.

Metallizing to Prevent Corrosion and Overcome Rust. C. J. Pott (*Eng. Contract Record*, 1934, **48**, 277–278).—Discusses the process and summarizes its applications.—S. G.

Chlorinated Rubber as a Protective Paint for Aluminium. Guido Schultze (*Aluminium*, 1936, **18**, (3), 107–111); and (summary) *Light Metals Research*, 1936, **4**, (24), 451–452.—Chlorinated rubber is an effective protection against corrosion for pure aluminium, Hydronalium, and Elektron, and is more effective than resins and varnishes under strongly corrosive conditions, such as those in the chemical industry.—A. R. P.

To Make Paint Stick to Cadmium-Plated Steel. George A. Endern (*Indust. Finishing* (*U.S.A.*), 1936, **12**, (5), 22–24).—Washing in 2% acetic acid immediately after plating is recommended to remove adherent films of various salts which cause the paint to flake.—A. R. P.

Coal-Resin Paint and Its Application in the Chemical Industries. Association of the Resin with Aluminium Powder. F. Delarozière (*Rev. prod. chim.*, 1935, **38**, 710–711, 741–743; *C. Abs.*, 1936, **30**, 4021).—Incorporation of aluminium powder with coal-resin paint yields a product which is remarkably resistant to atmospheric and chemical corrosion, and is recommended for a great variety of uses. The painted surface offers a bright metallic appearance. It has a high reflective power with low emissivity. A recommended mixture contains aluminium powder 12·5, resin varnish 62·5, and diluent 25%.—S. G.

Corrosion Studies [Protection of Metals Near Sewage]. James A. Meacham (*Munic. Sanit.*, 1936, **7**, 15–17; *C. Abs.*, 1936, **30**, 3558).—Bituminous enamel applied in two coats was found to be the only paint which resisted corrosion of metal surfaces near raw sewage.—S. G.

***Object and Significance of Ground Paints.** J. van Loon (*Chem. Weekblad*, 1936, **33**, (22), 345–351).—Read before the Dutch Chemical Society. Describes tests carried out on behalf of the Dutch Corrosion Committee on several iron and lead-base pigments, to evaluate their protective qualities. Iron test-pieces covered with the materials were exposed on the roof of the laboratory in Delft. The paints were also tested for permeability to water, swelling (% water absorbed by the paint films when immersed) and for wear by 2 methods: (*a*) in a specially constructed apparatus; and (*b*) by a sand test. Results are given in a table and permeability to water is shown by curves.—L. A. O.

VI.—ELECTRODEPOSITION

(Continued from pp. 208–210.)

***Iron in Chromium Plating Baths and Its Removal.** E. Raub (*Mitt. Forschungsinst. Edelmetalle*, 1936, **10**, (1), 1–7).—Troubles due to iron begin to appear when about 5% of iron, based on the chromic acid content, has accumulated in the bath, but not until more than 10% of iron is present are they so serious that the bath must be scrapped or purified. Deposits obtained from iron-rich baths consist of iron–chromium alloys and have a matt grey colour which cannot be buffed to a satisfactory finish. The purification of a bath contaminated with iron is very difficult and, unless large volumes of solution are available, is not worth the trouble and cost; it is far better to avoid contamination by taking care that no iron parts come into contact with the solution and that iron articles to be plated are immersed in the bath with the current already switched on.—A. R. P.

Important Factors Governing Cyanide Plating [of Copper]. Jacob Hay (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (4), 31–37).—Some experiments on the throwing power of, and method of racking numerous small articles in, a copper cyanide bath are recorded. The original must be consulted for details of the results which do not lend themselves to abstracting.—A. R. P.

***On the Electrodeposition of Binary Alloys [Brass].** V. A. Sukhodskiy, V. L. Heyfez, and J. N. Chapurskiy (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1934**, (17), 209–216).—[In Russian.] The cathode and anode potentials in the tartaric acid brass plating bath were measured at various concentrations and current densities. The difference between the deposition potentials of the copper and zinc decreases with decrease in current density and with dilution, potential equalization occurring mainly at the expense of the copper potential.—N. A.

Gold and Silver Plating of Large Metal Surfaces. U. V. Baimakov (*Metallurg* (*Metallurgist*), **1936**, (2), 3–8).—[In Russian.] The methods used in gold and silver plating the Soviet emblems on the towers of the Kremlin (total surface area about 100 $m.^2$, with individual sections of 3 × 2 m.) are described in detail.—N. A.

***On [Lead-]Plating.** V. A. Sukhodskiy, V. L. Heyfez, and J. V. Wainer (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1936**, (18), 231–239).—[In Russian.] In lead cresolsulphonate solutions the cathode polarization is determined by the presence of active tars introduced with the acid. Up to a definite concentration of lead ions, evolution of hydrogen occurs and a spongy deposit is formed at high current densities; with increase in current density a greater concentration of lead is necessary for smooth deposits. Cresol- and phenolsulphonate baths give good microcrystalline deposits with 100% current efficiency.—N. A.

Nickel and Chromium Plating. Progress and Present Position. M. Ballay (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 77–83).—[In French.]—S. G.

Protection of Aluminium by Nickel. Giacomo Bionda (*Chimica*, 1935, **11**, (6/7), 237–239; *C. Abs.*, 1936, **30**, 3331).—The main difficulty in the nickel-plating of aluminium is to obtain satisfactory adherence of the nickel to the aluminium. This has been solved by two methods: (1) a French process, in which the article is cleaned with organic solvents, washed in a soda solution and then in pure water and immersed in a nearly boiling solution containing 3·5% of ferric chloride solution (*d* 1·42) and 2·5% of 22 Bé hydrochloric acid; this covers the aluminium with a very adherent film of iron, and nickel-plating can then easily be effected electrolytically with baths containing nickel sulphate or chloride and another salt such as ammonium or magnesium sulphate. (2) an American process, in which the article is treated with hydrofluoric acid and/or nitric acid, washed and nickel-plated in the usual manner. The best results are obtained by (1).—S. G.

†*A Critical Study of Bright Nickel. Lawrence E. Stout (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (5), 25–29; and *Met. Ind.* (*Lond.*), 1936, **48**, (26), 722–723).—The merits of various " brighteners " in a commercial nickel-plating bath were critically investigated; the bath contained nickel sulphate 12, nickel chloride 4, and boric acid 2 oz./U.S. gall. Glucose (1·5–2 oz./gall.) produced a marked improvement in the brightness of the deposits; malic acid produced a brilliant shiny soft deposit at low current densities, but tended to form a black slime on the cathode with slight variation in the plating conditions, and lactic acid tended to produce yellow deposits of poor lustre. None of the numerous patented brighteners tested gave an absolutely perfect plate and were highly intolerant of variations in operating conditions.—A. R. P.

The Progress of Nickel in the Course of the Past Ten Years. John F. Thompson (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 91–99).—[In French.] See *Met. Abs.*, this vol., p. 82.—S. G.

***Electroplating with the Alloy Iron–Nickel.** D. Stepanow, F. Marshak, and E. Beliakov (*Zhurnal Obstchei Khimii* (*J. General Chem.*), 1935, [A], **5**, (1), 30–33).—[In Russian.] Cf. *Met. Abs.*, 1935, **2**, 516, 626. A mixture of ferrous and nickel sulphates, with small quantities of ammonium sulphate, nickel chloride, and glue, was used as electrolyte, the anodes being strips of iron and nickel, and the cathode an etched strip of thin tinplate. The deposits are uniform, microcrystalline, adhere firmly, and are, in small thicknesses, elastic. During prolonged electrolysis, solutions low in nickel give an increasing nickel content in the deposits, and *vice versa*.—M. Z.

***On Electroplating of Metals [Tin].** V. L. Heyfez and J. V. Wainer (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1934**, (16), 165–180).—[In Russian.] The cyanide, oxalate, and sulphate-sulphonate tin-plating baths were examined. In the cyanide and oxalate baths the anodes slowly become passive with a corresponding change in the composition of the bath; the sulphate bath is free from these defects and has a good throwing power, as shown by the cathode and anode potential curves and the electrical conductivity.—N. A.

†Zinc Plating. Oliver Sizelove (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (5), 17–24).—Good bright deposits of zinc can be obtained from an acid solution containing zinc sulphate 50, aluminium sulphate 3, ammonium chloride 2, sodium acetate 2, and corn syrup 0·5 oz./U.S. gall. operated at p_H 4–4·5 with 15–25 amp./ft.², or from an alkaline bath containing zinc cyanide 8, sodium cyanide 5, sodium hydroxide 10, and mercuric cyanide 0·25 oz./U.S. gall. operated with pure zinc anodes at 25° C. with 10–20 amp./ft.². In using the alkaline bath the plated work is rinsed thoroughly then dipped for a few seconds in 0·5% nitric acid, again washed, and finally dipped in a 1·5% cyanide solution, rinsed in hot water, and dried. The behaviour and operation of various proprietary baths is also described.—A. R. P.

***Cathodic Current Distribution in Galvanic Electrolytes. I.—Investigation of Zinc Baths.** E. Mantzell (*Z. Elektrochem.*, 1936, **42**, (6), 303–315).—The current distribution in six fundamental types of zinc bath under practical conditions was determined, namely, acid zinc sulphate–ammonium sulphate; acid zinc chloride–ammonium chloride; boric acid; neutral zinc chloride–pyridine; alkaline zinc bath; and cyanide–alkali bath. In general, increasing the electrode interval to obtain uniform metal distribution is only of value when edge scattering is completely prevented, otherwise exactly the reverse occurs. There is a close reciprocal effect between polarization and bath conductivity. At the slight risk of affecting the polarization to some extent, an improvement of the current distribution can only be obtained by increasing the conductivity, either by the addition of conductivity salts or by increasing the temperature. The effect of the conductivity salt on the polarization is very small. The current distribution deteriorates with increasing current density in all baths, but an improvement is effected by reducing the zinc ion concentration, if the conductivity is kept constant. Such electrolytes are only of use if suitable precautions are taken to prevent sponginess. The effect of the acidity and of the addition of organic or inorganic reagents on the polarization and therefore on the current distribution is very small. Cyanide baths produce the best current distribution, that of alkali zinc baths being relatively poor, because the cathode potential in the alkali zinc solutions alters very little with the current density. In acid baths, the metal distribution is almost identical in form with the current distribution, because the current efficiency approaches 100% over the whole cathode surface. With the other baths, the metal distribution is more or less evened, to the greatest extent in the cyanide baths. The structure of the metal undercoat has an appreciable effect on the distribution.
—J. H. W.

Electro-Galvanizing of Round Wire. (Weisselberg.) See p. 260.

New Zinc Coating Process Developed by Graselli Chemical Co. —— (*Amer. Metal Market*, 1936, **43**, (99), 3).—A brief note. The zinc is deposited in a brilliant, dense, and ductile form. The process operates at room temperature, high current efficiencies, and high or low current densities, with a bright plating range from 15 to 100 amp./ft.^{2}.—L. A. O.

***Corrosion-Protective Value of Electrodeposited Zinc and Cadmium Coatings.** (Blum, Strausser, and Brenner.) See p. 260.

***The Röntgenographic Study of Zinc and Cadmium Films Deposited in the Presence of Colloids.** (Palatnik.) See p. 252.

The Mechanism of Electroplating. Hiram S. Lukens (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (5), 30–40; discussion, 40–44).—Recent work of L. on the distribution of current in a plating bath is described in an educational lecture.—A. R. P.

***The Application of Mesle's Chord Method in the Measurement of the Thickness of Electrodeposits.** W. Blum and A. Brenner (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (4), 14–20).—An account is given of the results of work more fully reported in *J. Res. Nat. Bur. Stand.*, 1936, **16**, (2), 171–184. See *Met. Abs.*, this vol., p. 125.—A. R. P.

Measuring the Thickness of Metal Coatings. F. C. Mesle (*Met. Ind.* (*Lond.*), 1936, **48**, (19), 544).—Abstract from *J. Res. Nat. Bur. Stand.*, 1936, **16**, (2), 171–184. See *Met. Abs.*, this vol., p. 125.—J. E. N.

Relative Value of Accelerated Corrosion and Outdoor Exposure Tests [of Plated Coatings]. (Blum.) See p. 259.

Some Peculiarities in the Formation of Metallic [Electro-]Deposits. Jean Billiter (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 101–104).—[In French.] See *Met. Abs.*, this vol., p. 84.—S. G.

***An Electron Diffraction Study of the Structure of Electrodeposited Metals.** (Finch and Sun.) See p. 252.

Metal Rectifiers for Electroplating. —— (*Elect. Rev.*, 1936, **118**, (3038), 242).—Metal rectifiers, made from copper strips oxidized on both sides, and provided with forced-air cooling, with outputs up to 12,000 amp. at 5 v., are available as a source of current for electroplating.—J. C. C.

†Some Problems of Technical Electrometallurgy. III.—Galvanotechnics. (Masing.) See abstract below.

VII.—ELECTROMETALLURGY AND ELECTROCHEMISTRY

(Other than Electrodeposition.)

(Continued from p. 210.)

†Some Problems in Technical Electrometallurgy. II.—Hydro-Electrometallurgy. III.—Galvanotechnics. IV.—Anodic Treatment of Aluminium. G. Masing (*Z. Elektrochem.*, 1936, **42**, (6), 325–327).—Discusses some of the difficulties encountered in the electrolytic extraction and purification of zinc, in electrodeposition, and in the Eloxal process of anodic oxidation of aluminium, and the means of overcoming them.—J. H. W.

Electrolytic Manufacture of Boron and Its Alloys. L. Andrieux (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 69–75; and *J. Four élect.*, 1936, **45**, (2), 55).—[In French.] See *Met. Abs.*, this vol., p. 84.—S. G.

***The Mechanism of Electrolytic Processes. I.—The Anodic Oxidation of Some Metals of the Platinum Group.** J. A. V. Butler and G. Drever (*Trans. Faraday Soc.*, 1936, **32**, (2), 427–435).—On anodic polarization of palladium, rhodium, and iridium electrodes in acid or alkaline solutions an adsorbed layer of oxygen is formed on the metal surface prior to the establishment of the oxygen overvoltage as is the case with platinum. With the two first-named metals a second stage in the oxidation is observed with the slow formation of peroxide films which are reduced cathodically at a more positive potential than the absorbed oxygen. With rhodium the peroxide layer is a monomolecular film, but with palladium it is several molecules thick. A possible explanation of the formation of these layers is suggested.—A. R. P.

Anodes for Oxidation Reactions. V. Čupr (*Chem. Listy*, 1936, **30**, (2), 27–29; (3), 43–46; *C. Abs.*, 1936, **30**, 3727).—Č. presents a detailed review of the construction and composition of anodes at whose surfaces oxidation occurs during the passage of an electric current; particular attention is devoted to a series of oxidations which take place at electrodes of iron, copper, cupric oxide, Fe_3O_4, platinum, and carbon.—S. G.

***Electro-Capillary Phenomena in Molten Salts.** S. Karpatschoff and A. Stromberg (*Z. physikal. Chem.*, 1936, [A], **176**, (3), 182–186).—The electrocapillary curves for tin, lead, cadmium, and thallium in molten alkali halide baths were determined; the original must be consulted for details of the results and methods used. The capacity of the electrolytic double layer of tin at the equilibrium potential in the electrolytes used was 90–100 μF./cm.^{2}.—B. Bl.

VIII.—REFINING

(Continued from p. 210.)

Refining of Primary Aluminium in Melting and Casting. I. E. Gorshkov (*Metallurg* (*Metallurgist*), **1936**, (1), 82–95).—[In Russian.] The main impurity which distinguishes raw from refined aluminium is sodium. Inclusions of electrolyte, alumina, carbides, nitrides, &c., appear only under abnormal conditions of electrolysis and melting. The presence of sodium in aluminium can be detected by (*a*) treatment of the metal with water containing phenolphthalein, (*b*) fusion of the bottom of the ingot and weighing of the resultant oxide film (this gives an indication of the purity of the aluminium), (*c*) measurement of the shrinkage of an ingot. The metal may be refined by remelting in furnaces with an inclined hearth.—N. A.

***Bismuth in Copper: Its Effects, Determination, and Some Experiments in Its Gaseous Elimination.** C. O. Bannister and W. M. Doyle (*Trans. Inst. Min. Met. Eng.*, 1935, **44**, 163–196; discussion, 196–226).—See *Met. Abs.*, 1935, **2**, 106.—S. G.

IX.—ANALYSIS

(Continued from pp. 211–213.)

Quantitative Spectral Analysis Under Changeable Discharge Conditions. V. Naedler (*Compt. rend. acad. sci. U.R.S.S.*, 1935, [N.S.], **4**, (1/2), 23–26; *C. Abs.*, 1936, **30**, 3351).—[In German.] Considerable difficulty is encountered in spectroscopic analysis owing to the lack of constancy of the electric discharge. Such difficulty can be overcome by establishing empirically a curve showing the effects of such variation. The method is described and illustrated in the determination of small amounts of Pt and Rh in Ag.—S. G.

***The Quantitative Spectrum Analysis of Traces.** W. Seith and A. Keil (*Z. Elektrochem.*, 1936, **42**, (6), 299–302).—In order to establish the amount of an impurity in a metal, a characteristic line of the impurity is compared with a suitable adjacent line of the metal. To render visible the faint line of the impurity, overexposure is often necessary, the metal line being then very heavy. It also frequently happens, for instance, in the case of light metals, particularly magnesium, that there is no suitable line due to the metal near to the impurity line. In this case, the comparison is made indirectly, known traces of another metal (or metals) being introduced into the metal, and the additional lines so obtained being used for comparison with those of the metal and the impurity. The technique of this method is described and illustrated.—J. H. W.

Detecting the Cobalt in the Presence of Other Cations. R. W. Woronizow (*Zhurnal Prikladnoi Khimii* (*J. Applied Chem.*, 1935, [B], **8**, (3), 555–556).—[In Russian, with German summary.] The solution to be examined is treated with HCl, decanted, and solid NH_4CNS added in slight excess of saturation. Crystals of $SnCl_2$ are then added, until reduction is complete. A mixture of amyl alcohol and acetone is then added, and the mixture well shaken: a blue coloration of the supernatant alcohol layer indicates Co.—M. Z.

***Fractional Detection of Potassium.** H. N. Potchinok (*Zhurnal Prikladnoi Khimii* (*J. Applied Chem.*), 1935, [B], 8, (3), 524–527).—[In Russian, with German summary.] A new method of detecting K in the presence of all other cations, except Rb and Cs, is based on $Na_3Co(NO_2)_6$ together with HCHO and alkali (to remove the effect of NH_4OH). Up to 0·6 mg. of K in a dilution of 1/5000 can be detected.—M. Z.

The Analysis of Aluminium in the Presence of High-Chromium and Other Associated Alloys. A Short Process. A. McMurphy (*Aluminium and Non-Ferrous Rev.*, 1936, 1, (7), 285–286).—To determine Al in magnet steels and heat-resisting materials, the sample is dissolved in HCl and, after oxidizing with HNO_3, the solution is added from a dropping funnel to boiling NaOH. The volume is made up to 500 c.c., 250 c.c. filtered off, and Al precipitated as phosphate.—J. C. C.

***Bismuth in Copper: Its Effects, Determination, and Some Experiments in Its Gaseous Elimination.** (Bannister and Doyle.) See p. 265.

***The Potentiometric Determination of Cobalt and Manganese with Potassium Ferricyanide in Steel and Alloys.** Peter Dickens and Gerd Maassen (*Arch. Eisenhüttenwesen*, 1935–1936, 9, (10), 487–498).—See *Met. Abs.*, this vol., p. 166.—A. R. P.

***The Determination of Tin in Alloys with Antimony and Lead (Antimony Less Than 2 Per Cent.).** H. F. Hourigan (*Analyst*, 1936, 61, (722), 328–333).—The method depends on the dissolution of the sample in HCl in a current of O_2-free CO_2 and titration of the resulting $SnCl_2$ with KIO_3. If much Sb is present the alloy is dissolved in HCl and $KClO_3$ and the solution reduced with Al in a CO_2 atmosphere before titration. An apparatus is described in which the reduction or dissolution of the sample can be carried out, and by means of which the antimony residue may be filtered off before the solution is titrated, all the operations being done in CO_2 free from O_2.—A. R. P.

***Gravimetric Estimation of Vanadium with Ammonium Benzoate.** F. M. Schemjakin and W. F. Tschapygin (*Zhurnal Prikladnoi Khimii* (*J. Applied Chem.*), 1935, [B], 8, (3), 536–542).—[In Russian, with German summary.] The method is based on reduction of V to tetravalent and its precipitation in the form of vanadyl benzoate, and subsequent weighing as V_2O_5. V may be precipitated in acid solution from its ores and alloys in the presence of Mg, Al, Cu, Mo, W, and Ti, but Fe and Cr must be removed first. The method is stated to be somewhat quicker, simpler, and cheaper than that of Rose and Roscoe.—M. Z.

X.—LABORATORY APPARATUS, INSTRUMENTS, &c.

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

(Continued from p. 213.)

The Use of Ball-Bearings in the Construction of Microscopes. M. Wielding (*J. Roy. Microscop. Soc.*, 1936, [III], 56, (1), 31–38).—Ball bearings in microscope construction enable heavy bodies, even when loaded eccentrically by such fittings as illuminators, to respond accurately to the micrometer movement without any fear of seizing. Examples are described of their use in fine adjustments and rotating and mechanical stages in Leitz instruments.—J. C. C.

X-Ray Interference Employing Contact Between the Source of Radiation and the Crystal. H. Seemann (*Z. tech. Physik*, 1935, 16, (11), 443–447).—A method of X-ray crystal analysis, mounting the crystal very close to the source of X-rays, *e.g.* an anti-cathode window, is described. The method presents advantages in the investigation of mosaic structures.—J. S. G. T.

Description of a New Laue Camera and the Accompanying Reflection Net (Reflexnetz). Erich Schmid (*Z. Krist.*, 1935, **91**, (1), 95–96).—[In German.] A Laue camera for the determination of the orientation of metallic crystals is described. The film is in a plane parallel to the primary beam, and with flat metal crystals the primary beam strikes the flat surface of the specimen at 45° C. Under these conditions the time of exposure is much less than in the usual arrangements.—W. H.-R.

A New High-Power X-Ray Tube for Structural Analysis. S. V. Sergeev (*Zhurnal Tehnicheskoy Fiziki* (*J. Tech. Physics*), 1936, **6**, (1), 83–93).—[In Russian.] An electron X-ray tube with revolving anode is described in detail. —N. A.

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

(Continued from pp. 213–214.)

Mechanical Tests for Engineering Materials. A. M. Roberts (*Assoc. Eng. Ship. Draughtsmen*, **1935–1936**, 89 pp.).—A comprehensive review of mechanical tests includes the tensile test, bend test, hardness, impact, transverse, and fatigue tests as well as high-temperature tests and tests on welds. The types of testing machine used in each case are described in brief detail, dimensions of test-specimens are given, and also various test data. Test procedure and the position of test-pieces are also considered with special reference to steels.—J. W. D.

Miniature Machines for Photographically Recording Mechanical Tests on Metals. P. Chevenard (*Bull. Assoc. Tech. Fond.*, 1935, **9**, (9), 308–311).—Abstract from *Bull. Soc. d'Encour. Ind. Nat.*, 1935, **134**, 59–73; see *Met. Abs.*, 1935, **2**, 392.—J. E. N.

Testing of Soft Solders. H. Hanemann (*Z. Metallkunde*, 1936, **28**, (3), 68–69).—A mechanical soldering machine is described by the aid of which it is possible to produce soldered joints in sheet metals under standardized reproducible conditions, especially as regards temperature, time of soldering, and solder consumed. A series of tests of solders and soldering conditions on brass, zinc, galvanized iron, and tinplate is planned to determine the best solders and working conditions and the strength of the resulting joints. The tests contemplated include the following: (1) two strips are soldered together with an overlap of 5 mm., and the shear strength of the joint determined in the tensile machine; (2) torsion tests are made on butt-soldered rods prepared with a standard pressure during soldering; (3) the hot strength of joints is determined by hanging a 1 kg. weight on the lower of two strips of metal soldered together and observing the temperature at which the strips part; (4) corrosion tests are made in distilled water, tap-water, artificial sea-water, 10% tartaric acid solution, and 10% citric acid solution.—A. R. P.

Working Stresses. C. Richard Soderberg (*J. Applied Mechanics*, 1935, **2**, (3), A106–108).—See *Met. Abs.*, 1935, **2**, 438.—S. G.

On the Dilatometric Method of Analyzing Materials. J. F. Afonskiy (*Soobshenia Zentralnogo Instituta Metallow* (*Communic. Central Inst. Metals*), **1935**, (18), 186–201).—[In Russian.] Describes Chevenard's apparatus and a graphic method for calculating the differential curve for determining the true coeff. of expansion of metals.—N. A.

An Apparatus for Determining the Commencement of Permanent Set in Material Used for Flat Springs [Fatigue Testing]. E. V. Walker and K. L. Beak (*J. Sci. Instruments*, 1936, **13**, (3), 96).—A note. A machine is described for determining whether material used for flat relay springs will take up a permanent set on being deflected as a cantilever for many millions of operations at stresses below the elastic limit of the material. The springs are deflected by means of revolving cams with arrangements for counting in multiples of 5000 deflections. The instrument can be stopped at any multiple of 5000 deflections when any permanent set is measured microscopically.—W. H.-R.

***Accelerated Method for the Determination of the Creep Limit at Various Temperatures.** Max Schmidt and Helmut Krainer (*Mitt. tech. Versuchsamtes*, 1935, **24**, 5–20).—Rohn's accelerated creep test is described, and the results obtained by its use on a number of steels are analyzed mathematically. Since the logarithm of the rate of creep is a linear function of the reciprocal of the absolute temperature, the creep limit of a metal at any temperature may be calculated from the rate of creep determined in a test extending over 30–48 hrs. —A. R. P.

Experience of Methods Adopted for the Systematic Long-Term Testing of Creep. P. B. Mihailov-Miheev and G. M. Denisov (*Metallurg* (*Metallurgist*), **1936**, (3), 47–60).—[In Russian.] The methods and apparatus employed at the Leningrad Metal Works are described.—N. A.

***Slow-Bend and Impact Tests of Notched-Bars at Low Temperatures.** H. F. Moore, H. B. Wishart, and S. W. Lyon (*Amer. Soc. Test. Mat. Preprint*, **1936**, 6 pp.)—Results are given of slow-bend and impact tests of various metals (steels, cold-drawn 60 : 40 brass, cold-drawn copper, Duralumin 17 ST) at temperatures from room temperature to − 40° F. The correlation between results from slow-bend and from impact tests over this range of temperature does not differ greatly from that found by Petrenko (*J. Inst. Metals*, 1925, **34**, 547) for room temperatures. The autographic slow bend testing machine which was used is described. In a flexure test of a notched-bar specimen the friction of striking edge along the specimen seems to be rather a source of error, especially for specimens of metal so ductile that after the test they are not broken clean in two. Tension tests of notched specimens, both impact and slow-pull tests, are suggested as worthy of experimental study.—S. G.

***Similarity Between the Stress–Strain Diagrams Obtained from Impact Tensile and Torsional Tests.** Mititosi Itihara (*Tech. Rep. Tôhoku Imp. Univ.*, 1936, **12**, (1), 105–118).—[In English.] The static tensile and torsion diagrams for mild steel and brass are similar within 5–7% according to Ludwik's R–γ curve. A similar analogy within the same limits is observed between the impact tensile and torsion diagrams obtained at the same specific sliding speed. —A. R. P.

General Review of Tests for Determining the Characteristics of Aluminium Castings. A. von Zeerleder (*Congrès internat. Mines, Mét. Géol., Section de Métallurgie*, 1935, **1**, 229–234; and *Rev. Mét.*, 1936, **33**, (1), 1–6).—[In French.] See *Met. Abs.*, this vol., pp. 17 and 168.—S. G.

***Variations of the Rigidity Modulus with Changes in Applied Stress.** Gladys M. Francis (*Physics*, 1936, **7**, (4), 160–162).—An attempt to determine by an interferometer method whether or not any variations occur in the modulus of rigidity similar to those which occurred in the Young's modulus in D. K. Froman's experiment (*J. Inst. Metals*, 1930, **43**, 574). Any variations of more than 0·1 of a fringe could be detected readily. No such variations occurred. —S. G.

***On the Determination of Elastic Limit in Impression Tests as an Adjunct to Hardness Measurements.** G. Tammann and W. Müller (*Z. Metallkunde*, 1936, **28**, (3), 49–54).—When a cone is impressed into a metal a ridge is formed round the impression beginning at a distance, a_0, from the axis of the impression and increasing in height to the edge thereof. The stress along the circle of diameter a_0 is thus the elastic limit (E) in the impression test, and since a_0 is proportional to the square root of the load producing the ridge it follows that $E = Q/\pi a_0^2$, where Q is the load producing the impression. The value of a_0 can be determined by making the impression on a polished surface which has been slightly roughened by drawing it one way over coarse emery and then polishing the ridge away by rubbing in the direction perpendicular to the first with a fine emery paper; the diameter of the resulting circle round the impression is then equal to $2a_0$. Mathematical methods of calculating a_0 from the height of the ridge and the depth of impression are also given. For aluminium, copper, and electrolytic iron, E increases with the degree of rolling rapidly up to 10% reduction, then more slowly, the values for the annealed metals being 5, 6, and 14 kg./mm.^{2}; after 10% reduction 8·5, 25, and 35 kg./mm.^{2}, and after 40% reduction 10, 30, and 41 kg./mm.^{2}, respectively.—A. R. P.

***On the Minimum Principles of Theory of Elasticity.** Theodor Pöschl (*Bauingenieur*, 1936, **17**, (17/18), 160–164).—Mathematical.—B. Bl.

A Simple Method for the Determination of Magnetization-Coercive Force. C. W. Davis and Max Hartenheim (*Rev. Sci. Instruments*, 1936, [N.S.], **7**, (3), 147–149).—S. G.

RADIOLOGY.

Photography and Chemical Engineering Research. Walter Clark (*Trans. Amer. Inst. Chem. Eng.*, 1934–1935, **31**, 60–81; discussion, 81–82).—Recent developments in gamma radiology and in high-power magnification and photography as applied to steel examination are described *inter alia*.—A. R. P.

What X-Rays Can Do for Industry. Victor Hicks (*Physics*, 1936, **7**, (3), 79–84).—X-ray radiography is industrially useful in the physical inspection of welded, cast, or assembled products. X-ray diffraction analysis contributes to the understanding of the details of the processes of manufacture and treatment of alloys and other materials. Both applications have led to the control and improvement of industrial processes and products.—S. G.

Use of X-Rays in Casting Control. K. V. Peredelskiy (*Aviopromishlenost* (*Aircraft Ind.*), **1936**, (2), 32–35).—[In Russian.] A review.—N. A.

[Radiology in] Autogenous Welding. R. Schmidt (*Usine*, 1936, **45**, (12), 29; *C. Abs.*, 1936, **30**, 3765).—The radio-metallography of welds and the application of X-rays for their examination is explained. A radiogram of a piece 20 mm. thick can be obtained with 100,000 v., of one 80 mm. thick with 200,000 v., and of one 140 mm. thick with 400,000 v.; the time of exposure is less than 1 minute. The limitations of the method in recognizing defects are discussed.—S. G.

***X-Ray Investigation of Fine Structure, and Its Value in the Practical Testing of Materials.** A. Karsten (*Feuerungstechnik*, 1936, **24**, (5), 78–80).—The use of X-ray analysis for the investigation of special steels and refractory materials is briefly referred to. A small X-ray apparatus suitable for the purpose is described and illustrated.—J. S. G. T.

Progress in Design of Shock-Proof X-Ray Tubes for Therapy and Industrial Röntgenography. M. J. Gross (*Amer. J. Roentgenology*, 1935, **34**, 518–522). —S. G.

XII.—TEMPERATURE MEASUREMENT AND CONTROL

(Continued from p. 170.)

On a New Combined Colour Pyrometer with Comparative Lamp. Gerhard Naeser (*Arch. Eisenhüttenwesen*, 1935, **9**, 483–485; and *Mitt. K.-W. Inst. Eisenforschung*, 1936, **18**, (2), 21–25).—All the usual types of optical pyrometers yield uncertain results since the actual black body value of most industrial hot bodies is not known with accuracy. N. describes, with reference to diagrams and photographs, a new colour-brightness pyrometer which gives readings with an accuracy of ± 10° for all hot bodies which are not selective radiators.—A. R. P.

An Automatic Potentiometer for Thermal Analysis. R. J. M. Payne (*J. Sci. Instruments*, 1936, **13**, (5), 158–161).—The thermal curve potentiometer previously described (*Met. Abs.*, this vol., p. 16) has now been made completely automatic by incorporating a photoelectric cell which operates suitable relays each time the spot of light from the galvanometer reaches the zero mark. —W. H.-R.

***On the Construction of Platinum Thermometers and the Determination of Their Basic Points.** W. H. Keesom and B. G. Dammers (*Physica*, 1935, **2**, (10), 1051–1058).—[In English.] Full report of an investigation previously reported in summary (see *Met. Abs.*, this vol., p. 87).—S. G.

***Comparison of Some Platinum Thermometers with the Helium Thermometer Between 0 and − 183° C.** W. H. Keesom and B. G. Dammers (*Physica*, 1935, **2**, (10), 1080–1090).—[In English.] Full report of an investigation previously reported in summary (see *Met. Abs.*, this vol., p. 87).—S. G.

***Determination of Fixed Points in the Low Temperature with a Hydrogen Thermometer.** Shin'ichi Aoyama and Eizô Kanda (*Bull. Chem. Soc. Japan*, 1935, **10**, (10), 472–481).—[In English.] A constant volume hydrogen gas thermometer is described; its accuracy as a low-temperature standard was investigated. The accuracy is ± 0·036° at 0° C. and ± 0·025° at − 200° C. The boiling points of liquid nitrogen and liquid oxygen and the sublimation point of solid carbon dioxide were determined with this instrument. It was also used for the calibration of a copper–Constantan thermocouple over the range 0° to − 200° C.—S. G.

***Researches on Torsion-Pointer Thermometers.** W. Grundmann (*Z. Instrumentenkunde*, 1936, **56**, (1), 26–31; *Sci. Abs.*, 1936, [A], **39**, 171).—Thermometers composed of bimetallic strips in the form of an Archimedean spiral, in which the temperature is indicated by a pointer moving round a dial, have been described (*ibid.*, 1935, **55**, 173–174). G. now describes further investigations into the performance of this type of instrument. Possible improvements are suggested, and it is concluded that a thermometer of this type can be developed into a useful instrument.—S. G.

Portable Pointer Galvanometer and Multi-Range Millivolt-Ammeter. —— (*J. Sci. Instruments*, 1936, **13**, (3), 100–101).—A note. A portable multi-range millivolt-ammeter is described briefly. The voltage ranges give full-scale deflection for 12 and 120 mv., and 1·2 and 12 v.—W. H.-R.

XIII.—FOUNDRY PRACTICE AND APPLIANCES

(Continued from pp. 214–216.)

Developments in the Foundry Industry During Recent Years. J. E. Hurst (*Metallurgia*, 1936, **14**, (80), 37–38).—Developments in the foundry during recent years are briefly reviewed, and include the production of castings of a greater degree of mechanical and metallurgical precision, a better knowledge of moulding sands and their influence on the quality of the final casting, mechanization and improvements in foundry plant, and the introduction of better and more efficient furnaces for all types of metals and alloys. Brief reference is also made to complex brasses and bronzes, nickel alloys, aluminium and magnesium light alloys, and the potential use of metals such as beryllium and lithium in the production of new alloys.—J. W. D.

Deoxidation and Degasification of Non-Ferrous Casting Alloys. I.—Deoxidation and Degasification of Nickel Silver Alloys. R. J. Keeley (*Trans. Amer. Found. Assoc.*, 1935, **43**, 251–255; discussion, 255–261; and *Met. Ind.* (*Lond.*), 1936, **48**, (8), 232–233).—The successful production of nickel silver castings, while difficult, may be satisfactorily accomplished by proper attention to detail. The two most common types of defect are porosity and tendency for dross inclusions; these can usually be attributed to faulty melting practice. K. cites instances of foundry difficulties and explains what was done to correct them. Sufficiently high pouring temperatures, proper deoxidation, and correct feeding are essential. Manganese and magnesium are recommended as deoxidizers when more than 3% lead is present, and silicon with less than 3% lead. Aluminium is undesirable in pressure castings, and silicon, magnesium, or aluminium is objectionable in rolling-mill alloys. In the discussion *T. E. Kihlgren* favours the restricted use of silicon. *Sam Tour*, on the effect of silicon on the contraction of the alloy, insisted that this is due to the degassing effect and not to the influence of silicon *per se*. The effect of sulphur in the fuel was discussed by several speakers and is being further investigated.—J. E. N.

†**Deoxidation and Degasification of Non-Ferrous Casting Alloys. II.—The "Modifying" Phenomenon and Its Probable Relation to Non-Ferrous Alloys.** C. H. Lorig and R. W. Dayton (*Trans. Amer. Found. Assoc.*, 1935, **43**, 262–268; discussion, 269–273; also *Met. Ind.* (*Lond.*), 1936, **48**, (9), 278–280; and (summary) *Light Metals Rev.*, 1936, **2**, (20), 347–348).—It is suggested that the "slag-cloud" hypothesis developed to explain graphite formation in cast iron may be applied to non-ferrous alloys. The authors review the work of various investigators in both the ferrous and non-ferrous fields and explain how the theories mentioned might be applied to non-ferrous cast alloys to explain why such alloys which are sound, reasonably free from contaminating metallic elements, and of good manufacture sometimes possess irregular if not totally dissimilar properties. The influence of nuclei on undercooling and speed of crystallization, the effect of superheating on dissolving or coalescing the nuclei, or varying their effectiveness, are discussed, and the theory is advanced that it may be possible to control the nuclei and their distribution, modifying the properties of many cast alloys when the conditions affecting each system have been individually investigated. In the discussion it is pointed out that, apart from their behaviour as nuclei for crystallization, small foreign particles may inhibit grain-growth.—J. E. N.

On the Manufacture of Aluminium Pistons for Internal Combustion Engines. Keijo Works of the Railway Board, Chosen (*Imono* (*J. Casting Inst. Japan*), 1936, **8**, (3), 145–149).—[In Japanese.] Although formerly aluminium-alloy pistons for internal combustion engines were purchased, as a result of research these are now manufactured at the Keijo Works at only one-sixth of the previous cost. Details are given of the composition of the alloy, melting operation, moulds, method of casting, and heat-treatment.—S. G.

Certain Peculiarities in the Manufacture of Aluminium Pistons. B. J. Ginzburg (*Izvestia NATI* (*Bull. Inst. Auto-Tractor Ind.*), **1935**, (2), 45–48).—[In Russian.] Review, based on data from American motor-works.—N. A.

Casting Technique of a Condenser Cover. Naval Technical Department (*Imono* (*J. Casting Inst. Japan*), 1936, **8**, (3), 141–144).—[In Japanese.] For a bronze casting, a condenser cover is one of considerable size. A standard practice has been established, based on many years of experience and research. The paper deals with patterns, moulds, sands, and melting and casting operations according to the specifications for standard practice.—S. G.

Judging the Temperature of Molten Bronze by Observation with the Naked Eye. Tsunejiro Suwa (*Imono* (*J. Casting Inst. Japan*), 1936, **8**, (3), 165–167).—[In Japanese.] With training fairly accurate estimations may be made of the temperature of molten bronze by the naked eye. S. measured the temperatures of molten bronzes at various stages and prepared a standard for the estimation of temperature by means of the naked eye.—S. G.

Recommended Procedure for Analysis of Defects in Brass and Bronze Sand-Castings. Non-Ferrous Division Committee (*Trans. Amer. Found. Assoc.*, 1935, **43**, 247–250).—Possible defects are listed under 18 headings, with a brief description of each. To establish the connection between cause and effect, the various operations in the manufacture of castings where troubles may arise are divided under 9 main headings. This is intended to serve as an outline of the problem and the proposed method of attack; later work will be directed to defining the difficulties and prescribing remedies.—J. E. N.

50 Rules for the Treatment of Graphite Melting Crucibles in the Foundry. Edmund Richard Thews (*Giesserei-Praxis*, 1936, **57**, (21/22), 230–231).—Conclusion of the article. See *Met. Abs.*, this vol., p. 215.—J. H. W.

The Drying of Moulding Sands in the Open Air. Troubles which may Arise from this Cause. M. Girault (*Rev. Fonderie moderne*, **1936**, (Jan. 1), 4–5).—Skin defects appearing on certain castings were not avoided by applying the usual remedial measures. An investigation showed the cause to be surface drying, by exposure to sunshine and air currents, of green-sand moulds in which some time elapsed between moulding and casting. The local decrease in humidity reduced the bond and strength below that required to prevent erosion of the mould by the incoming metal.—J. E. N.

Die-Casting: Present Position. Paul Bastien (*Congrès internat. Mines Mét. Géol. Section de Métallurgie*, 1935, **1**, 287–298; also *Rev. Mét.*, 1936, **33**, (1), 43–54; and (summary) *Light Metals Rev.*, 1936, **2**, (19), 329–330).—[In French.] See *Met. Abs.*, 1935, **2**, 718.—S. G.

Pressure-Casting of Light Alloys. K. V. Peredelskiy (*Aviopromishlenost* (*Aircraft Ind.*), **1936**, (1), 20–25).—[In Russian.] Review.—N. A.

Pressure Die-Castings in Brass. Herbert Chase (*Iron Age*, 1936, **137**, (19), 40–41, 103–105).—Reviews the present production possibilities of pressure die-castings, involving light-colour alloys having basic metal other than aluminium and zinc. Such alloys are Doler brass containing about 1% silicon, Brastil containing about 5% silicon, and a white nickel–brass containing nickel 5 and manganese 12%.—J. H. W.

Pressure Casting of Magnesium Alloys of the Elektron Type. K. V. Peredelskiy (*Liteinoe Delo* (*Foundry Practice*), **1936**, (1), 29–36).—[In Russian.] The following points in the pressure casting of magnesium alloys are discussed: suitable plant, selection of alloys and fluxes and their preparation, nature of rejects, solidity of castings, and their X-ray and metallographic examination. —N. A.

Die-Casting. Herbert Chase (*Machinist* (*Eur. Edn.*), 1936, **80**, (19), 364–366E).—Describes the plant and operations of a large American die-casting works producing about 230,000 lb. of zinc alloy per day, and the subsequent treatment of the castings.—J. H. W.

Alloys for Pressure Casting. G. Delcroisette (*Rev. Fonderie moderne*, **1936**, (Jan 1), 6–7).—In view of the increasing number of die-casting alloys which are being marketed attention is directed to the need for standardizing and reducing these to the minimum. Tentative specifications have been prepared in Germany for lead-, tin-, and zinc-base, and light aluminium- and magnesium-base alloys, to cover limits of chemical composition, mechanical properties, and tolerances, and these are reproduced. It is recommended that the same specifications be generally adopted.—J. E. N.

Die- and Pressure-Casting. Rudolf Schmitt (*Anz. Masch.*, 1935, **57**, (51), 42–44).—Alloys used in the processes are briefly discussed.—B. Bl.

Die-Casting Alloys. —— (*Machine moderne*, 1936, **30**, (128), 218–221).—An account of some of the principal die-casting alloys with a tabulated summary of the chemical analysis, melting-points, and mechanical properties of 25 alloys; the source of proprietary materials is indicated in several cases. —P. M. C. R.

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

(Continued from pp. 128–129.)

The Importance of Secondary Metals. Elmer W. Pehrson (*Amer. Metal Market*, 1936, **43**, (55), 5, 6; (57), 5).—Chiefly statistical.—A. R. P.

Utilizing Zinc and Lead Clippings, &c. C. C. Downie (*Machinist* (*Eur. Edn.*), 1936, **80**, (16), 263E).—The zinc and lead waste is placed in an open-hearth furnace fired with coke, and a higher temperature than that required to melt the metals is maintained. The two metals ignite, and the white fume is collected in cotton bags by suction fans. The scrap must be free from iron. The bag product is mixed with oil and passed through hydro-extractors in the usual manner for the production of white paint.—J. H. W.

XV.—FURNACES AND FUELS

(Continued from pp. 216–217.)

Heat Transfer in Continuous Reheating Furnaces. J. E. Eberhardt and H. C. Hottel (*Heat-Treat. and Forging*, 1936, **22**, (3), 144–149).—The design and construction of new furnaces has in general been on the basis of known performance of previous similar installations, although predictions of performance based on knowledge of the mechanism of heat transmission in furnaces have been of some assistance. The factors determining the rate of heat transmission in continuous reheating furnaces are considered quantitatively: data on 6 furnaces operating under widely varying conditions are analyzed, and an equation is developed which fits the data and lends itself readily to use in design calculations.—J. E. N.

The Inductive Heating of Process Equipment. Robert C. Center (*Trans. Amer. Inst. Chem. Eng.*, 1934–1935, **31**, 44–57; discussion, 57–59).—Applications of high- and medium-frequency inductive heating for vessels for carrying out chemical reactions, &c., are briefly discussed.—A. R. P.

New Developments in the Brass and Copper Industry. W. Wirt Young, Jr. (*Gas Age-Record*, 1936, **77**, (17), 433–434, 438).—The use of towns' gas for melting brass in crucible furnaces, for preheating furnace linings, tube extrusion, bright-annealing of copper and brass, and tinning is briefly referred to.
—J. S. G. T.

Two-Stage Combustion of Natural Gas for Volume Loads. George M. Parker (*Gas Age-Record*, 1936, **77**, (21), 563–566, 576).—The conversion of natural gas of calorific value about 1000 B.Th.U./ft.^{3} into a fuel-gas having substantially the same properties as producer gas, and its two-stage combustion in furnaces are briefly discussed.—J. S. G. T.

Coal and Coke: Occurrence, Testing, and Utilization. A. C. Fieldner and W. A. Selvig (*Amer. Soc. Test. Mat. Symposium on Indust. Fuels*, **1936**, Jan., 1–28). **Industrial Fuel Oils.** H. V. Hume (*ibid.*, 29–45). **Manufactured Gas.** P. T. Dashiell and F. H. Trembly, Jr. (*ibid.*, 46–60). **Liquefied Petroleum Gas.** —— (*ibid.*, 61–70).—A series of four papers reviewing the availability of raw materials, the possibility of new processes, the tests applied to fuels and their significance, the utilization of the various fuels, and the general economic aspects of the products and their utilization.—J. W. D.

XVI.—REFRACTORIES AND FURNACE MATERIALS

(Continued from pp. 172–174.)

†**New Developments and Trends in Refractories. Processes and Materials.** Louis J. Trostel (*Trans. Amer. Inst. Chem. Eng.*, 1934–1935, **31**, 473–492; discussion, 492–493; and (summary) *Chem. and Met. Eng.*, 1935, **42**, (7),

363–367).—Recent advances in the mixing, pressing, and burning of refractory bricks are critically reviewed, and an account is given of the development, manufacture, and properties of new refractory materials based on crystalline aluminium silicates and spinels. The applications of the newer refractories to electric furnaces, cement kilns, and glass melting tanks are discussed and an indication given of the trend of future developments in the refractory industry. —A. R. P.

Recent Developments in Refractories. Christopher E. Moore (*J. B'ham. Met. Soc.*, 1936, **16**, (1), 16–29; discussion, 29–39).—See *Met. Abs.*, this vol., p. 92.—S. G.

Modern Furnace Applications of Refractories and Insulations. J. L. Spence and J. W. Craig (*Iron Steel Canada*, 1936, **19**, (1), 3–9).—A general review of materials used in the construction of furnaces to resist the action of atmosphere and slag at high temperatures. The properties and applications of the common refractories and insulation are tabulated. The processes used in the manufacture of several types of refractory bricks are briefly described.—J. E. N.

Secondary Expansion of Flint Fireclays. J. O. Everhart (*Eng. Exper. Sta. News, Ohio State Univ.*, 1936, **8**, (2), 9–10).—The majority of flint fireclays possess abnormal shrinkage and expansion properties between 1200° and 1400° C.: an examination of 21 clays indicated a directional weakness in the larger clay grains, and it is suggested that adequate grinding of the raw material should eliminate or reduce the abnormal expansion.—P. M. C. R.

The Beneficiation of Ohio Fireclays. Myril C. Shaw (*Eng. Exper. Sta. News, Ohio State Univ.*, 1936, **8**, (2), 10–12).—A study of the refinement of the crude Laurentian clays by the film flotation method.—P. M. C. R.

XVII.—HEAT-TREATMENT

(Continued from p. 174.)

***Effect of Iron Impurities on the Annealing of High Brass.** W. A. Gibson and J. H. Doss (*Amer. Soc. Test. Mat. Preprint*, **1936**, 6 pp.).—The effect of 0·008–0·287% iron impurities on the annealing of "high" brass (copper 64·1–66·4%) was studied for temperatures from 600° to 1200° F. (315°–649° C). Particular attention was devoted to the effect of annealing at 800°–900° F. (426°–482° C.), and iron impurities below 0·10%. The effect of time on the annealing of brass is discussed. It is found that: (1) in high brass, having lead and iron as the only impurities, as much as 0·09% iron is not harmful to annealing; (2) in such material a great discontinuity in annealing properties occurs between 0·09 and 0·118% iron; (3) more than 0·118% iron is not harmful to sheet high brass provided that proper allowance is made in processing and use for its annealing properties, but if these allowances are not made it is very harmful.—S. G.

***Stress-Relief Annealing High-Strength Monel Metal Plate.** Peter R. Kosting (*Amer. Soc. Test. Mat. Preprint*, **1936**, 9 pp.).—The effect of temperature of annealing and time at the temperature on the annealing of ½ in. thick high-strength Monel metal plate is discussed. Plates from 4 heats were studied. All plates were hot-rolled but finished at a low temperature except some plates from one heat which were cold-rolled 10%. Extensometers of 0·0002 and 0·00002 in. sensitivity were used to determine yield-strength, 0·00% set. All plates showed a maximum increase in yield, 0·00% set, in the range 550°–600° C. Yield-strengths at higher % set reached maximum values at lower temperatures: the greater the effect of cold-work the lower was the annealing temperature at which maximum values were obtained. The value of the yield-strength, 0·00% set, is appreciably affected by internal strains. It is apparent that this Monel metal plate must be heated to 550°–600° C. to be stress-relieved.—S. G.

XVIII.—WORKING

(Continued from pp. 217–219.)

†**Photoelectric Cells in Rolling-Mills.** K. Johannsen (*Elektrotech. Z.*, 1936, **57**, (6), 150–152).—Two examples of the use of photoelectric cells in rolling-mills are described: (1) regulation of a cutter to shear sections of definite length from the rolled metal; the radiation of the rolled metal cuts out the switch mechanism by means of a cæsium photo-cell; (2) as a means of testing cold-rolled thin sheets and foils for holes exceeding 0·3 mm. in diameter the light through which affects the cell and yields a current impulse which operates a signalling device.—B. Bl.

Bakelite Bearings [for Rolling Mills]. —— (*Engineer*, 1936, **161**, (4194), 570). **Bakelized Bearings.** —— (*Electrician*, 1936, **116**, (3028), 790).—The difficulties associated with the usual types of bearings for rolling mills are briefly reviewed. Bearings consisting of cloth bonded in a form of synthetic resin are described and their characteristics are outlined. Instructions for their use are included. By the use of these bearings in certain circumstances, power has been saved to the extent of 30–50% and bearing life increased by as much as 10 times.—R. Gr.

Manufacture of Nichrome Wire. J. S. Ginzburg and A. D. Goldberg (*Soobshenia Leningradskogo Instituta Metallow* (*Communic. Inst. Metals Leningrad*), **1936**, (15), 70–78).—[In Russian.] Alloys with nickel 60, chromium 15–16, and manganese 2% are better prepared in crucible furnaces than in high-frequency furnaces without vacuum, since the latter method produces alloys with many impurities along grain-edges and therefore difficult to work. The ingots should be cleaned and forged at above 1000° C., preferably at 1200°–1220° C. prior to hot-rolling at 1150° C.—N. A.

Hot-Pressed [Extruded] Light Metal Parts. —— Reichenbach (*Aluminium*, 1936, **78**, (3), 100–102).—Machines and methods for the extrusion of aluminium alloy parts are described, and the mechanical properties of the products obtained from several alloys are tabulated.—A. R. P.

Hot-Pressing [Extruding] of Light Metals into Dies. —— Brunnkow (*Anz. Masch.*, 1935, **57**, (51), 38–40).—The properties of brass, aluminium, and magnesium alloy die-extruded articles are compared with one another.—B. Bl.

A New Lead Extruding Machine. R. E. Horley (*G.E.C. Journal*, 1936, **7**, (2), 120–126; and *Met. Ind.* (*Lond.*), 1936, **48**, (21), 577–581, 595).—A description of a new type of extrusion press designed to eliminate troubles due to pinholes, oxide inclusions, and weak seams. The article is illustrated with photomicrographs showing these various defects which are difficult to avoid with the older types of extrusion machines.—S. V. W.

Metallurgical Aspects of Deep-Drawing. III.—Desirable Qualities in Deep-Drawing Materials: Their Nature and Specification. J. D. Jevons (*Met. Ind.* (*Lond.*), 1936, **48**, (20), 563–566; (22), 607–610, 619).—See also *Met. Abs.*, this vol., pp. 175, 218. Apart from the undesirable negative properties already enumerated, the desirable positive properties are: (1) high ductility with high tenacity; (2) smooth surface before and after drawing, and (3) very small variation in thickness. The effect on (1) of impurities and variation in composition is important. The benefit of uniform grain-size on both (1) and (2) is very marked. The selection of microsections and the interpretation of the observations are discussed at length, special attention being paid to directionality and orientation of "apparent greatest width" of crystals. No physical test has yet been devised which can form the basis of a specification for metal for deep-drawing. After reviewing critically the types of specification now in use, J. concludes that, whilst most of these may reveal radically unsuitable metal, they are not sufficiently delicate to serve as reliable acceptance tests to indicate the behaviour of the metal in actual drawing operations, and suggests that correct microscopical observation coupled with Erichsen and tear length tests are at present the most valuable criteria. The closest collaboration between supplier and user is essential to establish mutually satisfactory specifications.—J. E. N.

The Use of Rubber and Wood in the Forming of Aeroplane Parts. C. O. Herb (*Machine moderne*, 1936, **30**, (128), 181–186). **The Use of Rubber and Wood as Die Materials.** — (*Machinery* (*Lond.*), 1936, **48**, (1238), 405–409).—See *Met. Abs.*, this vol., p. 175.—P. M. C. R.

Production of a Seamless Brass Ring. B. Radford (*Machinist* (*Eur. Edn.*), 1936, **80**, (16), 256–275E).—Describes the operation lay-out, tool design, and operations for the production of seamless brass rings of " difficult " design. —J. H. W.

Economical Working with Hard Metal Tools. W. Fleischhauer (*Anz. Masch.*, 1935, **57**, (51), 45–46, 59–64).—Detailed examples are given of the type of hard metal tool most satisfactory for different jobs.—B. Bl.

Metal Turning for Instrument Makers. A. S. Newman and R. S. Clay (*J. Sci. Instruments*, 1936, **13**, (4), 109–114).—A general description of lathe work and the choice of tools.—W. H.-R.

XIX.—CLEANING AND FINISHING

(Continued from p. 219.)

Pickling of Non-Ferrous Metals. Otto Vogel (*Anz. Masch.*, 1935, **57**, (51), 74–76).—Pickling solutions are recommended for aluminium and its alloys, aluminium–bronze, lead, Britannia metal, copper and its alloys, nickel, Monel metal, zinc, and tin.—B. Bl.

Inhibitors: a Physicochemical Interpretation of Their Action. E. Jimino and I. Grifoll (*Anales soc. espan, fís. quím.*, 1935, **33**, (328), 947–951 ; *C. Abs.*, 1936, **30**, 3378).—An attempt is made to explain the inhibiting action of moderators in pickling. At the start this action may be due to adsorption phenomena. Its intensity effect is connected with the operation of local cells which probably cause the actual pickling. This explains why the inhibitor should preferably be retained with greater consistency in the cathodic regions. If the latter are covered by inhibitor there will be an increase in current density. This, or other physical effects inherent with local cells, causes a large hydrogen overvoltage, which prevents evolution of hydrogen. Therefore, acid is not spent uselessly. There is no obstacle to discharge of the necessary hydrogen ions, so that atomic hydrogen may reduce the higher oxides and thus cause pickling. In addition to electrochemical action, the inhibitor reduces the surface tension of the liquid, to assist which some substances may be added. Aqueous soap causes an analogous effect. The speed of hydrogen evolution is thus reduced only by mechanical means.—S. G.

The Polishing of Metals. E. J. Dobbs (*J. B'ham. Met. Soc.*, 1936, **16**, (2), 46–53 ; discussion, 53–64).—See *Met. Abs.*, this vol., p. 131.—S. G.

Coating Polishing Wheels with Abrasive. W. E. Warner (*Machinist* (*Eur. Edn.*), 1936, **80**, (17), 275E).—Water used for soaking a mixing glue should have been distilled. Old and fresh glue should never be mixed. Old wheels should be preheated to 120° F. (49° C.), new wheels to 110° F. (43° C.). After glueing and coating with abrasive, drying for at least 5 hrs. at from 65° to 70° F. (18°–21° C.) is necessary.—J. H. W.

Latest Equipment for Finishing Departments. —— (*Indust. Finishing* (*U.S.A.*), 1935, **11**, (4), 16–20, 22, 24, 26, 30, 32, 34–36).—Developments and improvements in apparatus made during 1934 are described and reviewed; the new apparatus includes spraying pistols, paint spreaders, vapour cleaners and degreasers, and enamelling devices.—A. R. P.

Gold and Silver Bronzes [Paints]. Herman Kahn (*Indust. Finishing* (*U.S.A.*), 1935, **11**, (6), 9–12).—The composition, properties, and uses of bronze and aluminium paints are briefly discussed.—A. R. P.

Gold, Silver, and Aluminium–Bronze Powders. Otto A. Both (*Indust. Finishing* (*U.S.A.*), 1935, **11**, (9), 9–11, 46–49).—Methods of manufacture and testing of bronze powders and their uses in finishing metal and wood work are discussed.—A. R. P.

Metal Container Finishing. Henry F. Frank (*Indust. Finishing* (*U.S.A.*), 1935, **11**, (6), 18–21).—Automatic apparatus and procedure for lithographing the outside of tooth-paste tubes and for varnishing the inside of cans for preserving food are briefly described.—A. R. P.

Major Review of Developments During 1935 in New Finishing Materials for Metal, Wood, Paper, Fibre, Leather, Fabrics, Rubber, &c. —— (*Indust. Finishing* (*U.S.A.*), 1936, **12**, (3), 19–30, 32, 37–38, 40, 42, 44, 46).—S. G.

XX.—JOINING

(Continued from pp. 219–222.)

***New Materials for Glass–Metal Seals.** A. W. Hull, E. E. Burger, and L. Navias (*Phys. Rev.*, 1936, [ii], **49**, (8), 647).—Abstract of a paper read before the American Physical Society. A careful study was made of the thermal expansion of iron–nickel alloys, and a successful search carried out for a glass to match one of these alloys. This new combination has certain advantages over the iron–nickel–cobalt combination previously described (*Met. Abs.*, 1935, **2**, 68) in that the alloy is less oxidizable, less difficult to make, and less expensive; and the glass is softer and more easily sealed, while maintaining an equally low coeff. of expansion.—S. G.

Notes on the Design of Vacuum Joints in Metal Apparatus. R. M. Archer (*J. Sci. Instruments*, 1936, **13**, (5), 161–165).—A. describes the construction of joints designed for use with metallic Dewar vessels, but suitable for other vacuum apparatus. Joints between two metal tubes, a metal and glass tube, a ground flange on a glass tube, and a ground metal plate are described and illustrated, and also a strong equatorial joint between copper hemispheres, and a joint between a thin tube and a spinning. All these are strong and easy to make and open, whilst the soldering of one part is not apt to cause unsoldering of another. The technique of soldering for vacuum apparatus is discussed.
—W. H.-R.

Testing of Soft Solders. (Hanemann.) See p. 267.

Outline of the Recent Progress Made in the Welding of Aluminium and Its Alloys. Jacques Douchement (*Congrès internat. Mines Mét. Géol., Section de Métallurgie*, 1935, **1**, 411–418; also *Rev. Mét.*, 1936, **33**, (3), 189–196; and *Aluminium and Non-Ferrous Rev.*, 1936, **1**, (8), 391–396).—[In French.] See *Met. Abs.*, this vol., p. 131.—S. G.

Autogeneous Welding of Aluminium and Magnesium and Their Alloys with the Use of Special Fluxes. L. Stiavelli (*Alluminio*, 1934, **3**, (4), 195–201; *Brit. Chem. Abs.*, **1935**, [B], 808).—The flux used is a mixture of alkali chlorides and fluorides, with added materials for cleaning. Details are not given.—S. G.

The Welding of Aluminium. G. O. Hoglund (*Welding Eng.*, 1936, **21**, (2), 26–30).—A "signal contribution" is said to have been made to the welding art when a flux for welding aluminium was developed in America, in 1910. Pure aluminium and the 1·25% manganese alloy are welded by gas, metallic and carbon arc, and resistance methods. Of the heat-treated alloys, "51 S" (Al + Si + Mg) and "53 S" (Al + Si + Mg + Cr) are recommended for welding, the former being used for furniture, and the latter for architectural work, beer-barrels, and milk cans. Arc-welding is advised for thick material where distortion is to be avoided. Tables are given of electrode size and current required for different thicknesses and of machine settings for resistance welding.
—H. W. G. H.

Gas Welding Aluminium and Its Alloys. G. O. Hoglund (*Welding Ind.*, 1936, **3**, (12), 430–433).—Brief particulars are given of the properties of welds in pure aluminium and some of its alloys, but the major part of the article deals with applications in architectural, brewery, dairy, transport, and chemical work.—H. W. G. H.

The Welding of Herculoy. R. Steinmetz (*Welding Eng.*, 1936, **21**, (4), 28–31).—Herculoy (copper–silicon–tin) may be welded by carbon arc (electrode negative), metallic arc (electrode positive), oxy-acetylene, and resistance methods. For fusion welding, backing-plates are necessary, but preheating is not required even for thick material. The technique recommended is described, and full details are tabulated.—H. W. G. H.

Welding Magnesium Alloys. A. J. T. Eyles (*Indust. Gases*, 1935, **16**, (4), 198–199).—Very careful cleaning, prior to welding, is essential. For thin sheet, flanging of the edges is not recommended; for thicknesses from $\frac{1}{8}$ to $\frac{3}{8}$ in., the edges should be bevelled to an angle of 45°. Special fluxes (composition not stated) are recommended, and preheating is advised for castings. The blow-pipe flame should be carefully regulated (presumably neutral). Completed welds should be annealed, hammered at 300° C. if possible, cleaned, and pickled in potassium dichromate solution.—H. W. G. H.

Resistance Welding. L. H. Frost (*Welding Eng.*, 1936, **21**, (1), 44–47; (2), 38–40; (3), 32–33; (4), 34).—The developments in design of spot-welding machines to suit special alloys are described. Accurate adjustment of heat, pressure, and time are shown to be necessary, and some methods adopted in practice are discussed. The interdependence of electrode pressure and current is emphasized and curves are given, showing the transformer capacity, electrode pressure, and the effects of arm spacing and length, for welding various thicknesses of mild steel. Flash-welding, projection-welding, and some applications are also discussed.—H. W. G. H.

***The Thermal Stress in a Strip Due to Variation of Temperature Along the Length and Through the Thickness.** J. N. Goodier (*Physics*, 1936, **7**, (4), 156–159).—The thermal stress developed by unequal heating of a flat strip is investigated, the temperature varying along the length and through the thickness, but not across the width. The problem is reduced to one of edge tractions only, for which general solutions are already known. From the properties of these solutions it is shown that for heating local to a narrow transverse band across the width, as might occur in welding processes, there is a concentration of stress near the edge of the strip, at the hottest part, and a maximum tensile stress of magnitude $E\alpha T_{\text{max.}}$ when the temperature is constant through the thickness. The corresponding quantity is found also for temperatures varying through the thickness. For steel $E\alpha\ T_{\text{max.}}$ is about 350 $T_{\text{max.}}$ lb./in.^{2}, the temperature being measured in ° C. above that of the cool parts of the strip, and this approaches the yield-point of mild steel for a maximum temperature (increase) of the order 150° C. For a given maximum temperature the maximum stress is reduced by about 40% by keeping one face cold.—S. G.

The Toxicology of Oxy-Acetylene Welding. Zolton T. Wirtschafter and Edward D. Schwartz (*J. Indust. Hyg. Toxicol.*, 1936, **18**, (3), 158–162; *C. Abs.*, 1936, **30**, 2902).—The toxic products to which an oxy-acetylene welder may be exposed are carbon monoxide, acetylene, metal fumes, and impurities in commercial acetylene. Welders working in vats, &c., should be provided with forced ventilation.—S. G.

The Economy of Welding. C. W. Brett (*Fuel Technologist*, 1936, **11**, (128), 316–317).—Recent progress in effecting repairs by welding, including the recent process of low-temperature welding, is briefly discussed.—J. S. G. T.

Electronic Welding Timers. Paul G. Weiller (*Electronics*, 1936, **9**, (5), 26–28, 38).—Discusses the design of valve-controlled welding timers, and their characteristics.—J. C. C.

XXI.—INDUSTRIAL USES AND APPLICATIONS

(Continued from pp. 222–227.)

***A Comparison of Corrosion-Resistant Steel (18 Per Cent. Chromium–8 Per Cent. Nickel) and Aluminium Alloy (24 ST).** J. E. Sullivan (*Tech. Notes Nat. Advis. Cttee. Aeronautics*, No. **560**, 1936, 12 pp.).—Comparison of the relevant mechanical properties of stainless steel and aluminium alloy " 24 ST " indicates that the latter has the better combination of properties for fabricated aircraft structures of stiffened sheet and for columns but that steel is preferable for highly stressed fittings carrying lugs and for wires and cables where pure tension is the primary controlling factor. Modern stainless steels are not so satisfactory as high-tensile steels.—A. R. P.

The Use of Aluminium in the Chemical Industry. Francis C. Frary (*Trans. Amer. Inst. Chem. Eng.*, 1934–1935, **31**, 3–16; discussion, 17–19).—See *Met. Abs.*, 1935, **2**, 58.—A. R. P.

***Light Metal Rotors for the Molecular Ultracentrifuge.** J. Biscoe, E. G. Pickels, and Ralph W. G. Wyckoff (*Rev. Sci. Instruments*, 1936, [N.S.], **7**, (6), 246–250).—Tests were made of the range of usefulness of light metal rotors in the ultracentrifuge for molecular sedimentation. These experiments prove that Duralumin 14 ST is definitely superior to the other aluminium- and magnesium-rich alloys tested. They also indicate a shape of 7 in. diameter rotor which if made of Duralumin 14 ST can be used safely for centrifuging at speeds up to 900 r.p.s. (field at centre of cell = 215,000 times gravity). For much more intense fields special steels must be used or the rotor diameter reduced.—S. G.

***The Use of Light Metal Rotors for the Ultracentrifuge.** J. Biscoe (*Phys. Rev.*, 1936, [ii], **49**, (8), 643).—Abstract of a paper read before the American Physical Society; this appears to be published in full by Biscoe, Pickels, and Wyckoff, *Rev. Sci. Instruments*, 1936, [N.S.], **7**, (6), 246–250 (see preceding abstract).—S. G.

Light Alloys in Rotor Design. W. R. Needham (*Machinist* (*Eur. Edn.*), 1936, **80**, (17), 278E).—Briefly describes the use of light alloys in rotor construction.—J. H. W.

The Standardization of Aluminium Rails for Electric [Traction] Installations. F. Streiff (*Bull. Assoc. Suisse Élect.*, 1936, **27**, (10), 265).—A summarized account is given of the recent recommendations issued by the Standardization Bureau of the Swiss Association of Mechanical Engineers regarding the mechanical and electrical properties of aluminium alloy rails. Tables show the principal requirements for hard, half-hard, and fully-annealed material which has received mechanical treatment by different methods.—P. M. C. R.

Light Metal Wheels for Road Vehicles. M. Koenig (*Aluminium*, 1936, **18**, (3), 81–87).—Substitution of aluminium alloy segmented rims for steel rims in the wheels of heavy road vehicles effects a reduction of 15–20% in the weight of the wheels; if the whole wheel is made of aluminium the weight is 40–60% less than that of a similar steel wheel.—A. R. P.

Metal Coaches on the Algerian Railway. —— Ducluzeau (*Rev. Aluminium*, 1936, **13**, (80), 155–158).—Describes the use of Duralinox (aluminium containing 7% magnesium), Alpax, and a 1·25% manganese aluminium alloy for the construction of railway coaches in Algeria.—J. H. W.

Uses of Aluminium-Plated Aluminium–Zinc Alloys. J. F. Kesper (*Obst. Gemüse-Verwertgsind.*, 1934, **21**, 547–549; *C. Abs.*, 1936, **30**, 3385).—Aluminium–zinc alloys have a high corrosion-resistance and good mechanical and working properties. They are suitable for the preserve industry.—S. G.

Aluminium Powder in Southern Paint Formulations. Robert I. Wray (*Amer. Paint J.*, 1936, **20**, (Mar. 30), 15–17; *C. Abs.*, 1936, **30**, 4023).—Aluminium priming paints give excellent results with Southern pine and other

Southern woods. The vehicles ordinarily used in aluminium paints have sufficient body so that they do not penetrate too far into the wood and rob the primer of its binder. The flake-like particles of aluminium appear to reinforce the paint films. The vehicle should not dry too quickly. An aluminium paint for use on wood should set to touch within 6 hrs. and dry fairly hard in 24 hrs. Synthetic resin vehicles dry somewhat more rapidly—S. G.

Can Boiler Explosions Due to Low Water be Prevented? [Fusible Plugs]. Frank Russell (*Railway Age*, 1936, **100**, (12), 499–501; *C. Abs.*, 1936, **30**, 3917).—On the Southern Pacific Lines the locomotives are protected against overheated crown sheets and explosions due to low water by the multiple application of boiler drop-plugs, consisting of brass plugs held in place by rings of fusible metal, the heads of the plugs on the fire side being covered with fusible metal to ensure free movement of the plug in case of low water.—S. G.

Necessary Conditions for the Use of Copper in Locomotive Fireboxes. —— Chan (*Rev. gén. Chemins de Fer*, 1936, **55**, (5), 331–338).—The high working temperature of the firebox in modern locomotive practice (300°–350° C.) leads to excessive wear on firebox stays and to recrystallization and ageing of the plates. The improvement in elastic limit induced by the cold-working of "pure" copper (maximum total impurities 0·4%), and the effect of temperatures up to 420° C. on the hardness of such cold-worked material are shown graphically. C. discusses the possibility of increasing the temperature of recrystallization above the working limit by additions of suitable metals (silver, cobalt, &c.) or of preliminary heat-treatment of the metal in order to prevent recrystallization.—P. M. C. R.

Report of the Copper Pipe Committee. —— (*Water Works Eng.*, 1934, **86**, 437, 496–497; *J. Amer. Water Works Assoc.*, 1935, **27**, 1092).—Results are given since 1928 for 12 water-works in the Netherlands with dune, river, and ground waters. Tinned pipes must have a non-porous layer of tin and not less than 30 grm. tin per m.^{2} covered. Copper, dissolved by 2% ammonium persulphate in 30 minutes, must not be more than 0·5 mg. per 100 cm.^{2} of tinned surface and the lead content of the tin must not exceed 0·5%. It is generally considered that the small amounts of copper dissolved from copper pipes have had no bad effect on health, but the maximum is fixed at 3 mg. per litre after 16 hrs. contact; this amount causes no unpleasant taste in water, beverages, or food. In only one case did copper exceed this amount; no danger arises from copper containing 0·25–0·45% arsenic. The amount of copper dissolved is affected by the carbon dioxide content, and copper is found to be slightly less soluble in warm than in cold water. The use of untinned copper is not permissible if the p_H is 6·9 or lower, or if dissolved oxygen exceeds 3 mg./litre, but tinned copper may be used.—S. G.

Copper Tubing [for the Refrigerating Industry]. O. Z. Klopsch (*Refrigerating Eng.*, 1936, **31**, (3), 162–165; *C. Abs.*, 1936, **30**, 3759).—Copper tubes for the refrigeration industry should be free from scale and dirt, be made from deoxidized copper, be bright-annealed after the last drawing operation, be thoroughly recrystallized and have an average grain-size not exceeding 0·04 mm. in diameter at a magnification of × 85, contain 99·90% copper and a maximum of 0·10% phosphorus, and should have a tensile strength of 30,000 lb./in.^{2}. Other physical properties are also given.—S. G.

The Use of Copper Tubing for Gas Service Replacement Work. H. S. Harris (*Gas Age-Record*, 1936, **77**, (17), 427–430).—The use of copper tubing for gas service pipe replacement is described. The tubing is, in some cases, drawn through conduits.—J. S. G. T.

Copper Channel Bus-Bars. G. W. Preston (*Elect. Rev.*, 1936, **118**, (3036), 160).—Bus-bars made of copper channel sections, provided with flexible expansion joints and silver-plated contact surfaces, are used in the Boulder

Dam hydro-electric scheme and are cheaper than laminated bars. The housing and inter-phase barriers are constructed of copper sheet.—J. C. C.

[Copper–]Copper Oxide Rectifiers. R. J. Bussell (*Elect. Eng. and Merchandiser* (*Melbourne*), 1935, **12**, (8), 267–274).—Read before the Postal Electrical Society of Victoria. An account of the construction, characteristics, and uses of copper–copper oxide rectifiers.—J. C. C.

Effect of Copper on Yeast. B. Lampe (*Brennerei-Zeit.*, 1935, **52**, 6–7).—A discussion, based on the results of previous investigators.—S. G.

The Primitive Use of Gold. T. A. Rickard (*Trans. Inst. Min. Met. Eng.*, 1935, **44**, 49–76; discussion, 77–87).—See *Met. Abs.*, 1935, **2**, 735.—S. G.

Nickel Alloys in Japanese Industry. Yasushi Taji (*Engineer*, 1936, **161**, (4190), 467–468, 493–494).—A review of applications for transport and general engineering work.—R. Gr.

Alloys for Electrical Heating. Harold Silman (*Met. Ind.* (*Lond.*), 1936, **48**, (22), 603–606).—A review of the properties required by alloys to be used for heating elements in electric furnaces, radiators, &c. Apart from the mechanical, thermal, and electrical properties of the alloys themselves, the nature of the oxide films affect the life of the element when exposed to high temperatures. The advantages and disadvantages of a number of alloys are discussed with special reference to the effect of impurities, and the improved resistance to scaling and attack by certain gases, which may be conferred by small additions of added metals. 80 : 20 nickel–chromium alloys are the most widely used, while the cheaper one containing nickel 60, chromium 16, and iron 24% is finding many applications.—J. E. N.

Power Plant Auxiliaries [Monel Metal]. G. E. Scholes (*Eng. Rev.*, 1936, **49**, (9), 708–710).—The importance of Monel metal as material for boiler plant accessories is illustrated by an account of its use for ball floats, baffles for steam purifiers, soot-blower valves and seats, and jointing rings.—P. M. C. R.

Nickel- or Monel-Plated Aluminium. —— (*New Products Digest*, **1936**, May 16; and (abstract) *Light Metals Rev.*, 1936, **2**, (22), 400).—See *Met. Abs.*, this vol., p. 180.—L. A. O.

New Industrial Uses for Silver. Lawrence Addicks (*Min. Met. Soc. Amer. Bull.* No. **236**, 1936, 5–20).—Recent developments in the use of silver in industry are described with especial reference to its use in sterilizing water, beer, vinegar, fruit juices, &c., in the manufacture of chemical plant, in the preparation of bearing metals, and in modifying the annealing characteristics of copper. Aluminium alloys with up to 10% silver have a fine, bright surface, are readily machinable, and find considerable uses in the manufacture of scientific instruments. Cadmium containing 0·75–1·75% silver and a little copper is finding increasing use as a bearing metal for motor-cars. Tin is considerably hardened by addition of up to 3·5% silver, and the alloy makes satisfactory strong, corrosion-resistant pipes. A low silver–tin alloy gives a much better tinplate for dairy use than does pure tin. The use of silver commutators on motors entirely prevents brush drop.—A. R. P.

***On the Applicability of the Oligodynamic Action of Silver in the Food Industry.** V. D. Slavin (*Voprosui Pitaniya*, 1935, **4**, (4), 147–152; *C. Abs.*, 1936, **30**, 3531).—[In Russian.] No oligodynamic action of silver (bactericidal action of very small doses) was observed in the case of water, and silver dosages which were bactericidal in water were found to be stimulators of bacterial growth in organic substances of foods. The use of silver for sterilization of food products and wash waters is justified neither by theoretical nor experimental results.—S. G.

Bearing Metals for Steam Power Plants. H. N. Bassett (*Eng. Rev.*, 1936, **49**, (9), 718–719; (11), 881).—This discussion of the characteristics of bearing metals includes a consideration of necessary lubrication conditions, and of the "conformability," thermal conductivity, resistance to deformation, and composition of such alloys in relation to steam operating conditions. Reference is made to recent work on the influence of cadmium and of lead on tin-base bearing alloys.—P. M. C. R.

The Dry Tin Accumulator. Ch. Féry (*Compt. rend. Assoc. franç. Avanc. Sci.*, **1934**, 56–58).—Cf. *Met. Abs.*, 1934, **1**, 328. The advantages of a dry accumulator, using plates of tin and lead, are briefly outlined. No constructional details are given.—J. C. C.

Switchgear with Special Metal Coatings for Oil Relays. Karl Meier (*Elektrotech. Z.*, 1936, **57**, (18), 493–495).—The use of tungsten–copper contacts is described.—B. Bl.

The Use of Hard Alloys [in the Refractories Industry]. D. L. Shmukler and —— Turezkiy (*Ogneupori* (*Refractories*), **1936**, (2), 30–33).—In the manufacture of refractories in the U.S.S.R. the parts of apparatus subjected to great wear are coated with hard alloys.—N. A.

The Use of Hard Metal Alloys in the Potassium Industry. V. Bakul and A. Kolesnikov (*Kalii* (*U.S.S.R.*), **1936**, (2), 18–23; *C. Abs.*, 1936, **30**, 3385).—[In Russian.] A description of the use and care of (1) Stalinit, which is a powdered alloy composed of chromium 9·5, manganese 11·5, carbon 12·5, and iron 55·5% and which is used for electric welding on the iron surfaces of drills and borers; and of (2) Pobedit, a mixture of tungsten and cobalt powders pressed together and sintered.—S. G.

Review of the Fields in which Zinc Die-Castings are Used.—I–II. Herbert Chase (*Met. Ind.* (*Lond.*), 1936, **48**, (21), 582–584; (23), 635–637).—(I.—) Zinc-base die-castings are used in the motor-car industry for mechanical parts, and structural and decorative fittings. Their use is extending to other industries and they are now incorporated in automatic sales machines, ciné-cameras and projectors, many types of electrical instruments and meters, and such office equipment as typewriters, duplicating machines and cash registers. In all these applications the accuracy of dimensions and the smoothness of surface reduce the finishing costs. (II.—) Further applications of die-castings are described, ranging from motor-car radiator grilles to mechanical toys. Lead-base die-castings are now generally restricted to parts where high corrosion-resistance is required, but for milder forms of corrosion the zinc-base alloys are quite satisfactory.—J. E. N.

Recommended Materials for Railway Engineering. —— (*Bur. Information on Nickel*, No. **AA9**, 1936, 6 pp.).—The recommended materials included the following non-ferrous alloys: 5 nickel–copper alloys, 3 nickel–copper–zinc alloys, 3 nickel–bronzes, 1 nickel–chromium alloy, and 1 nickel–chromium–iron alloy, and their use for such parts as firebox plates, stay bolts, safety valves and seats, feed-pump liners, electric heater elements, and rolling stock and decorative fittings both in the form of castings and extruded sections.—J. W. D.

†**Economy of Domestic Materials in German Locomotive Construction with Special Reference to the Question of Bearings.** R. P. Wagner and H. Muethen (*Glasers Annalen*, 1936, **118**, (4), 31–38; (6), 59–69; (10), 103–104).—Possible German materials for use in locomotive construction in place of imported metals are discussed, with special reference to the use of aluminium alloys.—B. Bl.

Service Pipe Friction Losses. James E. Gibson (*Water Works Eng.*, 1935, **88**, (9), 467; *J. Amer. Water Works Assoc.*, 1936, **28**, 283).—Friction losses in ft. of head per 100 ft. length are given for various types of house service connections. Diagrams are given for lead, copper, and galvanized pipe at flow rates of from 5 to 50 gall./minute and for loss of head up to 180 ft. per 100 ft. of pipe, the meter in each case being set at the pavement curb. Service line and meter account for the majority of head loss. The superiority in this respect of lead to copper and to galvanized iron is manifest. The standard "K" copper service tubing is, however, an ideal material for the purpose because of its sustained delivery, non-corrodibility, low cost, durability, ease of installation, and salvage value.—S. G.

The Formation of Metallic Blacks on Thin Foils by Evaporation. R. V. Jones and B. V. Rollin (*J. Sci. Instruments*, 1936, **12**, (4), 130–131).—A note. Heating of the receiver sometimes prevents the formation of black metallic deposits by evaporation. This may sometimes be prevented by the use of atmospheres of hydrogen or helium since these have relatively high thermal conductivities.—W. H.-R.

Metallic Insulation. J. T. Nichols (*Ice and Cold Storage*, 1935, **38**, 177–178; *C. Abs.*, 1936, **30**, 3541).—Metal surfaces which give relatively poor reflection in the visible region were found to be good reflectors of the longer wave-lengths; thus, black plate and galvanized plate are good reflectors for the longer wave-lengths. Relative reflectivities are given for a number of metals and plated materials.—S. G.

Bi-Metals. G. Keinath (*Arch. tech. Messen*, 1936, **5**, (58), T55–T56).—The thermal expansions of a number of metals and alloys between 0° and 600° C. are reproduced graphically, as are the degrees of expansion over the same range of 6 types of bimetal strip. The construction and principal uses of plain and pre-stressed strips are considered, and the degree of accuracy obtainable is discussed. An illustrated account is given of the testing of bimetal strip.
—P. M. C. R.

Metal Spraying in the Welding Shop. H. B. Rice (*Welding Eng.*, 1936, **21**, (1), 34–42).—The wire-gun process is discussed. Zinc, aluminium, and stainless steel are given as among the best spray metals. Sand- or grit-blasting is necessary to provide a key for the bonding of the deposit to the surface which is to be sprayed. Where heavy deposits are required, machine tooling or knurling, prior to blasting, is advisable. Some applications described are: zinc-spraying of seams in welded galvanized material; closing of porous welds; building-up worn machine parts; spraying of bearing metals: protecting aluminium crank-cases by spraying with lead; producing phosphor–bronze wear-resisting surfaces on steel parts.—H. W. G. H.

XXII.—MISCELLANEOUS

(Continued from p. 227.)

Physical Metallurgy. I.—Recent Developments. II.—Diagnosis of Mechanical Failures. Gordon Sproule (*Canad. Inst. Min. Met. Bull.*, No. **289**, 1936, 257–264, 264–277).—(I.—) Precipitation-hardening of non-ferrous metals is briefly discussed. (II.—) The use of the microscope in detecting faulty structures in steels is described.—A. R. P.

Fifty Years' Aluminium. K. Arndt (*Elektrotech. Z.*, 1936, **57**, (8), 199–200).—Historical.—B. Bl.

The Geographical Evolution of the Aluminium Industry. Robert Pitaval (*J. Four élect.*, 1936, **45**, (2), 49–51).—J. E. N.

The Primitive Smelting of Copper and Bronze. T. A. Rickard (*Trans. Inst. Min. Met. Eng.*, 1935, **44**, 227–252; discussion, 252–254).—See *Met. Abs.*, 1935, **2**, 127.—S. G.

Improvements in the Brass Industry. D. K. Crampton (*Monthly Rev. Amer. Electroplaters' Soc.*, 1936, **23**, (4), 23–30).—A review of recent progress, with special reference to annealing furnaces.—A. R. P.

The Training of Industrial Physicists. J. A. Crowther (*J. Sci. Instruments*, 1936, **13**, (5), 141–143).—S. G.

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(Continued from pp. 227–233.)

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***American Bureau of Metal Statistics.** *Year-Book of the American Bureau of Metal Statistics.* Sixteenth Annual Issue 1935. 4to. Pp. 118. [1936.] New York: The Bureau, 33 Rector St. ($2.00.)

[The statistics issued by the American Bureau will be found most valuable because of the obvious care that is taken in their compilation. They are clearly printed on large pages on good paper. The present volume, covering the year 1935, deals in the greatest detail with copper (34 pp.), lead (22 pp.), zinc (25 pp.), and gold and silver (13 pp.); the remaining pages deal with platinum, antimony, nickel, tin, aluminium, cadmium, cobalt, brass, and bronze, and the metal consumption per person in the principal countries from 1919 to 1935, average prices of the principal metals (New York and/or St. Louis and/or London) from 1886 to 1935, and yearly average prices from 1882 to 1935 for copper, lead, zinc, and silver. For the principal metals, the particulars given include world's production and consumption, refined metal production, mine production, producers, smelting and refining plants, smelter output, stocks, deliveries, ore reserves, imports and exports, metal available for consumption, use in U.S.A., and prices.]

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***British Metal Corporation, Ltd.** *Review of Base Metal Conditions. The World Consumption of Copper, Lead, and Spelter.* 4to. Pp. 15. 1936. London: The British Metal Corporation, Ltd. (Gratis.)

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***Clark, R. E. D.** *The Detection and Colorimetric Determination of Tin by Means of Substituted 1:2-Dimercaptobenzenes. A Specific Reagent for Tin.* (Technical Publications of the International Tin Research and Development Council, Series A, No. 40.) Roy. 8vo. Pp. 242–245. 1936. London: International Tin Research and Development Council, 378 Strand, W.C.2. (Gratis.)

[Reprinted from *Analyst*, 1936, **61**, 242–245; see *Met. Abs.*, this vol., p. 212.]

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Emich, Frederick. *Microchemical Laboratory Manual*, with a Section on *Spot Analysis*, by **Fritz Feigl.** Translated by Frank Schneider. Med. 8vo. Pp. 196, with 88 figures. London: Chapman and Hall, Ltd. (15*s.* net.)

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***Trillat, Jean J.** *La Diffraction des Électrons dans ses Applications.* (Actualités scientifiques et industrielles, No. 269.) Roy. 8vo. Pp. 61, with 6 plates and 13 illustrations in the text. 1935. Paris: Hermann et Cie. (18 francs.)

Viscosity and Plasticity. *First Report on Viscosity and Plasticity.* By J. M. Burgers, *et. al.* Pp. 256 + viii. Amsterdam: Committee for the Study of Viscosity, Academy of Sciences of Amsterdam. (Paper, 10 guilders; cloth, 11 guilders.)

XXIV.—BOOK REVIEWS

(Continued from pp. 234–236.)

Gmelins Handbuch der anorganischen Chemie. Achte völlig neubearbeitete Auflage. Herausgegeben von der Deutschen Chemischen Gesellschaft. Mitarbeiter: R. Glauner, M. A. W. Barnick, H. Lehl. System-Nummer 35: **Aluminium.** Teil A, Lieferung 4. Sup. Roy. 8vo. Pp. 535–682, illustrated. 1936. Berlin: Verlag Chemie G.m.b.H. (R.M. 24; ausl. pr., R.M. 18.)

This section of the "Handbuch" deals with the alloys of aluminium with a number of other elements, namely those comprising certain sub-groups of the Periodic System.

The whole of the available information on alloys containing Group VI*b* elements (sulphur–polonium), and Group III*a* elements (of which boron is the only one mentioned) is contained in half a page. 91 pages are then devoted to a description of the aluminium–silicon alloys. Their preparation, treatment, chemical and physical properties are described with a wealth of detail, and the influence of impurities is discussed at length. The alloys with elements of Group V*b* (phosphorus–bismuth) are next described, followed by those of Group I*a* (the alkali metals). The remainder of the book deals with the metals of Group II*a* (beryllium, magnesium, and the alkaline earths). 33 pages are devoted to alloys containing magnesium, and here again the treatment is extremely thorough.

It is necessary to point out that the only ternary alloys described are those whose constituents fall into the sub-groups mentioned above. Alloys containing magnesium and silicon, for example, are dealt with, but not those containing magnesium and copper, silicon and copper, &c. It is to be presumed that alloys of aluminium with the elements of the remaining subgroups will be described in a further volume.

The information given appears to be complete and up to date, and the book can be warmly recommended.—H. W. L. Phillips.

Aluminium-Legierungen. Patentsammlung. Von A. Grützner unter Mitarbeit von G. Apel. Nebst einem Markenverzeichnis bekannter Aluminiumlegierungen von A. v. Zeerleder. 2. Teil. Zugleich Anhang zu Aluminium Teil A in Gmelins Handbuch der anorganischen Chemie. Achte völlig neu bearbeitete Auflage. Herausgegeben von der Deutschen Chemischen Gesellschaft. Sup. Roy. 8vo. Pp. 343–868. 1936. Berlin: Verlag Chemie G.m.b.H. (R.M. 54; ausl. pr., R.M. 40.50.)

This volume completes the classified list of patents covering aluminium alloy compositions issued as an appendix to the volumes on aluminium in Gmelins "Handbuch," of which the first part was reviewed in *Met. Abs.*, this vol., p. 140. In addition to the patents, it contains a list of commercial aluminium alloys, under their trade names, giving their compositions and the names of their producers. As a work of reference the book should be invaluable.

—H. W. L. Phillips.

Traité Pratique de Galvanisation à Chaud. By Jean Chambran. (Bibliothèque de l'Usine.) Med. 8vo. Pp. 137, with 47 illustrations. 1936. Paris: L'Usine, 15 Rue Bleue, IXe. (Broché, 24 francs.)

The present work replaces that of C. Kluytman which was published in Paris in 1927. As its title implies, it deals primarily with the hot-dipping process of galvanizing, but a short illustrated note on sprayed zinc coatings is added as an appendix.

The book is divided into four parts. In the first part the general principles of hot-galvanizing, comprising the theoretical aspects of the pickling and galvanizing operations, a study of the steel base and of the spelter used, are dealt with in a very elementary and superficial manner; some of the author's statements in this section will certainly meet with criticism. The second part deals with the mechanical hot-galvanizing of sheets: the lay-out of a modern galvanizing plant, practical hints on the choice of apparatus, machinery, and raw materials, and detailed information regarding the many difficulties encountered in working such a plant, with suggested methods for overcoming them. The third part describes in a similar manner the practical aspects of the hand-dipping process, while machines for shearing, corrugating, bending, and roller-levelling are dealt with in the final part. Essentially a book for the man in the shop.—E. R. HOWELL.

Chromium Plating. A Text-Book for the Practical Electroplater, the Production Engineer, and the Student. By E. S. Richards. Foreword by J. B. Maclean. Second Edition, Revised. Cr. 8vo. Pp. vii + 131, with 70 illustrations. 1936. London: Charles Griffin and Co., Ltd. (4*s*. net.)

The chief modifications in this second edition are the insertion of 4 pages on metal rectifiers for supplying current in the plating shop and of 2 extra pages on barrel plating: otherwise, apart from a few minor additions, the book has not been altered. It is a book intended for the practical plater and is written presumably by a practical plater. Half of the space is devoted to plant (including sources and distribution of current) and to the preparation of work for plating (polishing and cleaning). The remainder of the book deals with copper, nickel, and chromium plating, the latter occupying about one-third of the total number of pages.

The book is chiefly valuable for its useful practical information on plant arrangement, polishing, and methods of jigging and suspending articles for chromium plating. The sections on copper, nickel, and chromium plating are disappointingly lacking in critical information: there is nothing about the control of these processes, and one is not told how to produce good-quality coatings of permanent value. Of the many points of detail which are open to criticism, two may be mentioned: some detrimental evidence of the quality of French chromium plating is cited, but there are so many important omissions of fact that the reader cannot judge its validity: it is to be regretted that an unqualified statement that " glue is sometimes added to (nickel plating) solutions to give white deposits " should have been repeated in this edition in view of the disastrous effect of this substance in concentrations exceeding about 0·1 grm. per litre. On pp. 66 and 67, sodium bisulphate, instead of bisulphite, is again given as an ingredient of a copper cyanide plating solution.—A. W. HOTHERSALL.

Le Groupe des Industries de l'Acétylène et de la Soudure Autogène—Ses Organismes Centraux. Par Lise Bloch-Sée. Med. 8vo. Pp. 176. 1935. Paris: Institut de Soudure Autogène, 32 Boulevard de la Chapelle.

This is a fascinating book. Its fascination depends in part on the remarkable ramifications of the subject, and, perhaps in greater part, on the way in which those ramifications are presented in clear perspective. A historical survey of the acetylene and welding industries in France is followed by an explanation of the growth of a central organization to guide their economic and technical progress. A detailed description is given of the origin, development, and functions of the following well-known bodies: the Chambre Syndicale de l'Acétylène, de la Soudure Autogène et des Industries qui s'y rattachent; the Office Central de l'Acétylène et de la Soudure Autogène; the Institut de Soudure Autogène; the École Supérieure de Soudure Autogène; the Société des Ingénieurs Soudeurs; and the Permanent International Commission of Acetylene, Acetylene Welding, and Allied Industries. The " powers that be " in the welding industries of this country cannot fail to find much of value in this study—to use its last words—" de la collaboration méthodique."—H. W. G. HIGNETT.

Electric Arc Welding Practice. A Handbook for Welding Engineers and Welders. By H. I. Lewenz. Demy 8vo. Pp. 126, with 103 illustrations. 1936. London: Crosby, Lockwood and Son, Ltd. (8*s*. 6*d*. net.)

This book is an elementary introduction to the subject of arc welding, the main principles of which are explained with as much clarity as the size of the book permits. In considering this size, account must be taken of excessively wide margins, which make the price less attractive than appears at first sight. The non-ferrous welder will find little of interest except a clearly written account (Chapter VI) of the stresses and deformation set up by welding. The author is so very cautious in passing rapidly, evading, or even ignoring controversial points, that it is almost a shock to find the definite statement that " For reasons which have been explained, d.c. is preferable to a.c." Although this opinion may be perfectly correct, the reasons are certainly not explained. The printing and illustrations are excellent. A remarkably useless appendix is given.—H. W. G. Hignett.

Handbook of Chemistry and Physics. A Ready-Reference Book of Chemical and Physical Data. Twentieth Edition. Edited by Charles D. Hodgman. Fcap. 8vo. Pp. xiv + 1951. 1935. Cleveland, O.: Chemical Rubber Publishing Co. ($6.00.)

The twentieth edition of this very useful book of reference differs from its immediate predecessor in many respects, the most important of which deal with organic chemistry. The section on X-ray spectra has been thoroughly revised and brought up to date and minor revisions have been made in most sections of the volume.

This book is undoubtedly one of the most useful works of its kind published in any language; it becomes more valuable with each succeeding edition. The data are accurate and well chosen. The book is well printed and bound and its form renders the finding of information extremely easy. It is a book to be recommended to chemists, physicists, and engineers, and all others whose work touches the fringes of chemistry and physics.

Soviet Science. By J. G. Crowther. Demy 8vo. Pp. x + 342, with 16 plates. 1936. London: Kegan Paul, Trench, Trubner & Co., Ltd. (12*s.* 6*d.* net.)

The title of this book is a misnomer: there is no Soviet science in the sense that there is Russian music. "Science in U.S.S.R." should be the title of the book. "Soviet" science, as revealed in this book, is the same as science everywhere else. The subject-matter of the book is almost entirely a recital of the various subjects of research in physics, chemistry, applied science, and biology undertaken in the scientific institutes established by the Soviets at Leningrad, Kharkov, Dniepropetrovsk, Moscow, and Sverdlovsk. This information has been supplied in most cases, probably in all, by the Directors of the respective Institutes, and reading it has left me with the impression that the main value of the work is propagandist in character. The first chapter is entitled "Dialectical Materialism," and opens "Little insight into the most original characteristics of the present developments in the Soviet Union can be acquired without some knowledge of the dialectical materialist philosophy." And what is dialectical materialism? "Its main ideas were propounded by Karl Marx"; the author then proceeds to extol Marx as a genius of high order, and to describe Lenin's book "Materialism and Empirio-Criticism" as perhaps the most remarkable essays on science ever written by a statesman of the highest order of genius. And listen to this (pp. 156, 157); "Authorities such as Rosenhain and Hume-Rothery have explained the necessity for a closer collaboration between metallurgy and physics. But their exhortations have not had the desirable effect, partly owing to the patent-monopolies in Western countries which create barriers between empirical industrial research and science. . . . The proper connection between industry and science can be established in a planned economy, so in the U.S.S.R. . . ." Well, my opinion, for what it is worth, is that scientific research in U.S.S.R., as revealed in this book, is over-planned, so much so indeed that the planning would appear to be the more important and essential part of the work. And with all this planning, there appear to be certain essentials, ministering to the comforts of everyday life, clamant for attention, but apparently receiving little or none, even from the scientists. Thus, *e.g.* (p. 38) the rooms set apart for scientists visiting Leningrad "are not entirely free from vermin. It (*sic*) gets in the cracks of the wall plaster near the stoves." The water-supply of Kharkov is inadequate and probably responsible for a considerable amount of minor summer ailments, and trains are quite commonly late in reaching their destinations. Here, ready to hand, are subjects for "planning" investigations, but Soviet science apparently passes them by, and bends itself to the task of tracking down the neutrino, investigating the facts of electrocapillarity, &c. Here, in this book, all this scientific activity is set out in very considerable detail.

Unless the book is propaganda, I can see no reason for its publication; if it is propaganda (and of this there is plenty of internal evidence) then the price is very high.—J. S. G. Thomas.

The Directory of Shipowners, Shipbuilders, and Marine Engineers, 1936. 34th Edition. Compiled under the direction of the Editor of *Shipbuilding and Shipping Record.* Demy 8vo. Pp. 806 + 75. 1936. London: The Directory Publishing Co., Ltd., 33 Tothill Street, S.W.1. (20*s.* net.)

This is a most useful publication to all who are interested in marine engineering. It furnishes particulars of all the leading steamship companies, ship-builders, ship-repairers, and marine engineering works throughout the world. It also gives the names of the directors and principal officials of the companies, and, in the case of shipping concerns, details of the fleets. A supplementary section includes a list of societies, institutions, trade and labour federations, and a directory of consulting marine engineers and naval architects. At the end of the volume are exhaustive indexes covering some 25,000 entries.


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